

AD-A273 486



12

CONTINUOUS-COOLING-TRANSFORMATION BEHAVIOR,
MICROSTRUCTURE, AND PROPERTIES OF
0.02C-3.5Ni-3.6Mo HIGH-STRENGTH STEELS

S DTIC
ELECTE
NOV 30 1993
A

by:

Masoud Molaei

Master of Science Thesis

This document has been approved
for public release and sale; its
distribution is unlimited.

August, 1992

93-07535



T-4223

Continuous-Cooling-Transformation Behavior, Microstructure,
and Properties of 0.02C-3.5Ni-3.6Mo High-Strength Steels

DTIC QUALITY INSPECTED 6

by

Masoud Molaei

Accession For	
NTIS CRA&I	☒
DTIC TAB	☐
Unannounced	☐
Justification	
By	
Distribution /	
Availability Codes	
Dist	Avail and/or Special
A-1	

T-4223

A thesis submitted to the Faculty and the Board of Trustees of the Colorado School of Mines in partial fulfillment of the requirements for the degree of Master of Science (Metallurgical and Materials Engineering).

Golden, Colorado

Date 8/18/92

Signed:

Masoud Molaei
Masoud Molaei

Approved:

Dr. Steven W. Thompson
Dr. Steven W. Thompson
Thesis Advisor

Golden, Colorado

Date 8/18/92

Dr. John J. Moore
Dr. John J. Moore
Department Head
Department of Metallurgical
and Materials Engineering

ABSTRACT

The continuous-cooling-transformation (CCT) behavior, microstructures, and properties of 0.02% C, 3.6% Mo, and 3.5% Ni steels with manganese contents ranging from 1 to 1.76% were investigated. The effects of cooling rate and prior austenite grain size were examined as part of the CCT work for the steel with 1.5% Mn. The CCT diagram for this steel is "flat-topped," with almost no variation in the transformation start and finish temperatures as a function of cooling rate.

A fully martensitic microstructure forms in response to cooling rates faster than about 37°C/s for a prior austenite grain size of about 7 μm . At cooling rates faster than 25°C/s, martensite forms for the 1.5% Mn steel with a prior austenite grain size of about 34 μm . At cooling rates slower than 25°C/s and 7°C/s for the prior austenite grain sizes of about 7 μm and 34 μm , respectively, the microstructure consists of acicular ferrite with micron-sized islands of martensite and austenite. At intermediate cooling rates, a mixture forms, made up of large packets of martensite and large packets of acicular ferrite. In this intermediate range of cooling rates, the volume fraction of martensite increases as the cooling rate increases.

The martensitic microstructures are characterized by a high dislocation density, some coarse twins, no detectable retained austenite, comparatively high hardness, a poor response to metallographic etching, and a good response to electrochemical thinning. Features of acicular ferrite are an equally high dislocation density, no twins, a significant volume fraction of retained austenite, comparatively low hardness, a good response to metallographic etching, and a poor response to electrochemical thinning. In addition, some small islands of austenite, between laths of acicular ferrite, transform to islands of high-carbon martensite, which frequently contain fine twins. Unlike previous investigations of low-carbon steels, granular ferrite was not observed, even at cooling rates as slow as 0.05°C/s .

For a cooling rate of about 1°C/s , yield strengths above 700 MPa (100 ksi) were obtained in the non-tempered condition. As manganese content increased, strength increased. Tempering resulted in lower strength and improved toughness. A single reaustenitizing treatment resulted in higher strength, although poorer toughness, than a double reaustenitizing treatment.

TABLE OF CONTENTS

	<u>PAGE</u>
ABSTRACT	iii
TABLE OF CONTENTS	v
LIST OF FIGURES	viii
LIST OF TABLES	xvi
ACKNOWLEDGEMENTS	xvii
1.0 INTRODUCTION	1
1.1 HSLA STEELS, A BRIEF SUMMARY	1
1.2 PURPOSE OF THE PROJECT	8
1.3 ROLE OF ALLOYING ELEMENTS IN HSLA STEEL	14
1.3.1 INTRODUCTION	14
1.3.2 CARBON	16
1.3.3 CHROMIUM	17
1.3.4 COPPER	17
1.3.5 MANGANESE	19
1.3.6 MOLYBDENUM	19
1.3.7 NICKEL	23
1.3.8 NIOBIUM	23
1.3.9 NITROGEN	24
1.3.10 SILICON	25
1.4 STRENGTHENING MECHANISMS IN HSLA STEELS	25
1.4.1 GRAIN REFINEMENT	25
1.4.2 SOLID-SOLUTION STRENGTHENING	33

1.4.3	PRECIPITATION HARDENING	37
1.4.4	TRANSFORMATION STRENGTHENING	40
1.5	MICROSTRUCTURE OF LOW-CARBON STEELS	44
2.0	EXPERIMENTAL PROCEDURES	54
2.1	AS-RECEIVED PLATES	54
2.1.1	GEOMETRY AND CHEMISTRY	54
2.1.2	GENERAL STRUCTURE	55
2.1.3	INCLUSIONS	57
2.1.4	HARDNESS TESTING	57
2.2	Ac ₁ AND Ac ₃ DETERMINATION	59
2.3	PRIOR AUSTENITE GRAIN SIZE	61
2.3.1	INTRODUCTION	61
2.3.2	HEAT TREATMENT AND METALLOGRAPHY	61
2.4	TRANSFORMATION KINETICS	66
2.4.1	GLEEBLE 1500	66
2.4.2	SAMPLE GEOMETRY	71
2.4.3	DILATOMETRY AS A FUNCTION OF HEATING AND COOLING RATE	73
2.4.4	DILATOMETRY	74
2.4.5	TESTING PROCEDURE	75
2.5	MICROHARDNESS MEASUREMENTS	78
2.6	MICROSTRUCTURAL CHARACTERIZATION USING SEM AND TEM	79
2.7	TENSILE TESTING	80

2.8	CHARPY IMPACT TESTING	83
2.9	FRACTOGRAPHY	84
3.0	RESULTS AND DISCUSSION	85
3.1	AS-RECEIVED PLATES	85
3.2	Ac_1 AND Ac_3 DETERMINATIONS	92
3.3	PRIOR AUSTENITE GRAIN SIZE	95
3.4	TRANSFORMATION KINETICS	96
3.4.1	DILATATION CURVES	99
3.4.2	CONTINUOUS-COOLING-TRANSFORMATION DIAGRAMS	101
3.5	MICROSTRUCTURAL CHARACTERIZATION	106
3.6	MECHANICAL PROPERTIES	153
3.6.1	TENSILE PROPERTIES	153
3.6.2	CHARPY IMPACT PROPERTIES	156
3.6.3	FRACTOGRAPHY	162
4.0	SUMMARY AND CONCLUSIONS	168
	REFERENCES	171
	APPENDIX: TABULATED DATA FOR TENSILE AND CHARPY IMPACT PROPERTIES	184

LIST OF FIGURES

<u>FIGURE</u>	<u>DESCRIPTION</u>	<u>PAGE</u>
1.1	Effect of 50% transformation temperature on tensile strength of bainitic steels (2)	5
1.2	Effect of carbon level and carbon equivalent (CE) on the susceptibility to heat-affected-zone (HAZ) cracking (22)	8
1.3	a. Schematic isothermal diagram indicating diffusion-dominated and shear-dominated austenitic transformation ranges. b. Isothermal transformation diagram for high-nickel, high-molybdenum alloy where f_t refers to the fraction transformed at the indicated time (12)	10
1.4	Continuous-cooling-transformation diagram for a low-carbon, copper-containing HSLA steel (37)	13
1.5	Solid solution hardening effects of the various elements as dissolved in pure iron (41)	15
1.6	The effect of carbon content and hence the pearlite content on impact-transition temperature curves of ferrite-pearlite steels (2)	17
1.7	Effect of chromium content on the single-phase austenite region for Fe-Cr-C alloys (41) . . .	18
1.8	Upper portions of CCT diagrams for steels alloyed with Mo. Nominal phase composition is 0.05 pct. C, 0.9 pct. Mn, 1.2 pct. Si and 0.5 pct. Cr (all wt. pct.). PF is polygonal ferrite and P is pearlite (52) . . .	20
1.9	Retardation of softening and secondary hardening during tempering of steels with various molybdenum levels (44)	21

LIST OF FIGURES
(CONTINUED)

<u>FIGURE</u>	<u>DESCRIPTION</u>	<u>PAGE</u>
1.10	Effect of alloying elements on the retardation of softening during tempering at 540°C (1000°F) relative to Fe-C alloys (44)	22
1.11	Effect of grain size of ferrite on yield stress and impact transition temperature of a steel containing 0.1%C, 0.5%Mn, 0.2%Si and 0.006%N (3)	26
1.12	a) Schematic representation of experimental thermal cycles in comparison with conventional heat treating cycles, referred to part of the Fe/C equilibrium phase diagram. b) Schematic representation of conventional and experimental rapid heat treating cycles (66)	31
1.13	Correlation between microalloy content and increase in yield strength. Increase in strength is due to grain refinement (σ_g) and precipitation strengthening (σ_{PR}) (77) . .	34
1.14	Effect of V, Nb, and Ti on CCT curves of a steel containing 2.5% Mn (62)	35
1.15	Yield strength versus atomic percent for various alloying elements (78)	38
2.1	Blanks cut to examine the variation in the microstructure and grain size	56
2.2	Sample used for inclusion analysis	58
2.3	Samples used for Ac_1 and Ac_3 determinations. Note that the center-line segregation was avoided	60
2.4	Wafers used for grain size studies	64

LIST OF FIGURES
(CONTINUED)

FIGURE	DESCRIPTION	PAGE
2.5	Different heat treatment schedules for grain size study	65
2.5	(continued)	66
2.6	Schematic diagrams showing the uniformity of temperature zone in: induction heated/conduction cooled (top), and electric resistance heated/conduction cooled (bottom) specimens (104)	67
2.7	Radial temperature gradient for an austenitic stainless steel subjected to electric resistance heating and axial conduction cooling. Heating rate: 4000°C/s, cooling rate: 100°C/s (104)	69
2.8	Schematic diagram showing the electric resistance heating of a sample used in the Gleeble. Uniform heating is produced across the work zone length and diameter (104)	70
2.9	Geometry of Gleeble samples. a) Cooling rates $\leq 10^{\circ}\text{C/s}$. b) (10 mm) $10^{\circ}\text{C/s} \leq \text{C.R.} \leq 25^{\circ}\text{C/s}$, (12 mm) $25^{\circ}\text{C/s} \leq \text{C.R.} \leq 75^{\circ}\text{C/s}$. c) $\text{C.R.} \geq 75^{\circ}\text{C/s}$	72
2.10	Graphical measurements of the 25, 50 and 75 percent transformation temperatures	77
2.11	Positions of the wafers cut for TEM studies relative to the position of the thermocouple (L: left, M: middle, R: right) . .	79
2.12	Orientation of tensile bars and Charpy bars relative to the rolling direction	81

LIST OF FIGURES
(CONTINUED)

<u>FIGURE</u>	<u>DESCRIPTION</u>	<u>PAGE</u>
3.1	(a) Light micrograph showing MnS inclusions. (b) and (c) SEM micrographs showing Ti- and Mn-rich inclusion clusters	88
3.2	Representative of EDS spectra for inclusions in steels A, B and C	90
3.3	The as-received microstructures of (a) steel A, (b) steel B and (c) steel C. Light micrographs	91
3.4	Dilatation curve for steel B. The specimen was heated up to 1200°C at a rate of 3°C/s	94
3.5	(a) Prior austenite grain size (about 30 μm) after one reaustenitizing treatment followed by quenching. (b) Prior austenite grain size (about 7 μm) after two reaustenitizing treatments followed by quenching	98
3.6	Plot of C-Strain versus temperature for a sample reaustenitized twice followed by cooling at 50°C/s after the second austenitizing treatment	100
3.7	CCT diagram for steel B	102
3.8	25, 50 and 75 percent transformation lines for (a) 1X and (b) 2X conditions	103
3.9	Microhardness versus cooling time between 800°C and 500°C	105
3.10	Martensitic microstructure of steel B after austenitization at 1200°C followed by quenching	107
3.11	Acicular ferrite with islands of retained austenite. (a) BF and (b) CDF images of the as-received steel B	109

LIST OF FIGURES
(CONTINUED)

<u>FIGURE</u>	<u>DESCRIPTION</u>	<u>PAGE</u>
3.12	SADP corresponding to the region shown in Figure 3.11	111
3.13	The as-received steel B showing acicular ferrite structure and the jagged edges of a lath. (a) BF and (b) CDF images	112
3.14	Microstructure of the as-received steel B. Notice the black M-A islands	115
3.15	Austenite centered-dark-field Image of the same region shown in Figure 3.14	116
3.16	Martensitic microstructures in specimens cooled at 150°C/s. (a) 1X condition and (b) 2X condition. Light micrographs	117
3.17	Fine martensitic lath structure produced after cooling at 150°C/s for the 1X condition	119
3.18	Fine martensitic lath structure produced after cooling at 150°C/s for the 1X condition	120
3.19	Twinned martensite in the sample cooled at 150°C/s, for the 1X condition	122
3.20	EDS analysis of the dark particles found on all of the TEM foils, except for the as-received condition	124
3.21	Light micrographs of the specimens cooled at 25°C/s. (a) 1X condition and (b) 2X condition	125
3.22	Microstructure of specimen cooled at 25°C/s for the 1X condition	127

LIST OF FIGURES
(CONTINUED)

<u>FIGURE</u>	<u>DESCRIPTION</u>	<u>PAGE</u>
3.23a	Acicular ferrite microstructure in specimen cooled at 25°C/s for the 2X condition, BF image	128
3.23b	Austenite centered-dark-field image of the same region shown in Figure 3.23a	129
3.24a	Martensitic microstructure in specimen cooled at 25°C/s for the 1X condition, BF image	130
3.24b	Martensite centered-dark-field image of the same region shown in Figure 3.24a	131
3.25	SADP from the region shown in figure 3.24. "M" and "m" are from two different martensite packets. "T" is the transmitted beam	133
3.26	DF image showing precipitates observed within a TEM foil. Cooled at 25°C/s, 2X condition	134
3.27	Acicular ferrite and retained austenite in specimen cooled at 25°C/s - 2X condition. (a) BF and (b) austenite CDF images	135
3.28	A pool of twinned martensite between laths of acicular ferrite in specimen cooled at 25°C/s, 2X condition	136
3.29	Acicular microstructure in a specimen cooled at 7°C/s, 1X condition. SEM micrograph	138
3.30	Light micrographs of the specimens cooled at 3.3°C/s. (a) 1X condition and (b) 2X condition	139
3.31	The edges of "bainitic" ferrite subunits. Specimen cooled at 3.3°C/s, 1X condition	141

LIST OF FIGURES
(CONTINUED)

<u>FIGURE</u>	<u>DESCRIPTION</u>	<u>PAGE</u>
3.32	Acicular ferrite and M-A islands in specimen cooled at 3.3°C/s, 1X condition	142
3.33	Acicular ferrite in specimen cooled at 3.3°C/s, 2X condition	144
3.34	Acicular ferrite in specimen cooled at 3.3°C/s, 2X condition	145
3.35	Light micrographs of the specimens cooled at 0.1°C/s. (a) 1X condition and (b) 2X condition	146
3.36	Acicular ferrite in specimen cooled at 0.1°C/s, 2X condition	148
3.37	Acicular ferrite in specimen cooled at 0.1°C/s, 2X condition. BF image	149
3.38	CDF image of the same region shown in Figure 3.37	150
3.39	Angular martensite island in specimen cooled at 0.1°C/s, 2X condition	152
3.40	(a) Tensile data for steels A, B and C (1X and 2X conditions). (b) Tensile properties of steel A before and after tempering	154
3.41	(a) Energy Absorbed versus temperature for steels A, B and C, 1X condition. (b) Energy Absorbed versus temperature for steels A, B and C, 2X condition	158
3.41	(c) Energy Absorbed versus temperature for steel A in the as-cooled and 620°C-tempered treatments. Includes data for the 1X and 2X conditions	160

LIST OF FIGURES
(CONTINUED)

<u>FIGURE</u>	<u>DESCRIPTION</u>	<u>PAGE</u>
3.42	SEM micrographs of the fracture surfaces after testing at (a) -195°C, (b) 23°C and (c) 100°C. Steel B, 1X condition	163
3.43	Fracture surfaces for steel B tested at -195°C. (a) 1X and (b) 2X conditions . .	164
3.44	Fracture surfaces for steel B tested at 23°C. (a) 1X and (b) 2X conditions . . .	165
3.45	Fracture surfaces for steel B tested at 100°C. (a) 1X and (b) 2X conditions . . .	167

LIST OF TABLES

<u>TABLE</u>	<u>DESCRIPTION</u>	<u>PAGE</u>
2.1	The specific designations and the dimensions of the as-received plates	54
3.1	Chemical Compositions of Steels	86
3.2	Hardness of the as-received steels	89
3.3	Calculated and measured austenite formation temperatures for steels A, B and C	93
3.4	Austenite grain sizes for various austenitizing times	97
3.5	Calculated and measured B_s and M_s temperatures for steels A, B and C	104

ACKNOWLEDGEMENTS

I would like to thank the Office of Naval Research (ONR) for funding this project, and for having the foresight to explore this new class of steels. I also thank the U.S. Steel Technical Center, who provided the necessary materials for my work, in addition to conducting chemical analysis on these steels.

Special thanks goes to my advisor, Dr. Steven Thompson, for sharing his knowledge and insights with me, and for his helpful discussions and support. His help in the TEM laboratory was invaluable.

I am grateful to my wife, Lesli, for her editing assistance, as well as her support, encouragement and patience.

I would also like to thank Dr. George Krauss and Dr. Chester Van Tyne for what I have learned from them at Colorado School of Mines, and for their well-considered contributions to this work.

Bob McGrew and Butch Brewer should also be credited for their suggestions and skilled help in the laboratory.

1.0 INTRODUCTION

1.1 High-Strength, Low-Alloy Steels: A Brief Summary

Low-carbon or mild steels have traditionally been used in the construction industry. For many years, ASTM A7 (now ASTM A283, Grade D) was the specification for the most commonly used structural steel for both bridges and buildings. Minimum yield strength for this steel is 230 MPa (33 ksi). Although it has the lowest unit price of any structural steel, it also has one of the highest price-to-yield-strength ratios (1).

In the early 1900's, the design of high-strength, hot-rolled steels was based on tensile strength, and there was little attention given to weldability, formability and resistance to brittle fracture (2). Riveting was the most common form of joining structural components, and the high tensile strength required for constructional purposes was achieved through employment of carbon (up to 0.3%) as an alloying element. Manganese was another major alloying element used to achieve higher strength.

During World War II, welding (rather than riveting) was introduced as a method of joining, and major structural failures in the steels then available required an

improvement in weldability and fracture resistance through reduction of carbon content (1,3). At the same time, the importance of both fracture toughness and low impact transition temperature were recognized. Similarly, the advantages of high manganese-to-carbon ratios for impact toughness were appreciated, and the significance of grain size was also established (2).

By the early 1950's, refinement of grain size by grain-refining additions, and the beneficial effects of ferrite-grain refinement on both yield stress and the ductile-to-brittle transition temperature had been demonstrated (2,3). During this period, the first steps toward achieving higher strength through grain refinement were taken by the introduction of aluminum to steels, which then combined with the nitrogen already present. Further increases in yield stress were then obtained by precipitation hardening. Vanadium, niobium and titanium were used, although the impact toughness was still not good, because the as-rolled grain size was coarse. Controlled rolling was then developed to improve the poor impact-transition properties of niobium steels, and this method turned out to be very successful. The result was that higher yield strengths of 450-525 MPa (65 to 75 ksi) along with low impact-transition temperatures, as low as -80°C , were obtained (1-3).

The development of high-strength, low-alloy (HSLA) steels was encouraged by the increasing demand for strong, tough, weldable steels for natural gas transmission lines, ships and off-shore drilling platforms (4). More recently, these steels have been used to increase the miles-per-gallon (MPG) efficiency of automobiles (5), to reduce the cost/weight ratio, to meet the notch-toughness requirements in constructional applications such as bridges, buildings and transmission towers (6-8) and for several other applications. The American Iron and Steel Institute (9) states that high-strength, low-alloy steels are "generally produced with emphasis on mechanical properties requirements rather than to chemical composition limits".

Beginning in about 1950, low-carbon bainitic steels were developed on a commercial scale simultaneously with ferrite-pearlite HSLA steels. Carbon contents in these steels ranged from 0.1 to 0.15%, and many combinations of alloying elements were used to produce bainitic structures during air cooling in a wide range of section sizes. Yield strengths in the range of 450-900 MPa (65 to 130 ksi) were achieved, depending on the carbon content and austenite transformation temperature (2). Bainite form at temperatures intermediate to those at which polygonal ferrite and martensite form. It has been shown (2) that

bainitic structures can be produced with a wide range of strengths (Figure 1.1).

Pickering (2,10,11) has summarized the aims of the development of low-carbon bainitic steels:

One goal is to produce a bainitic structure by air cooling, therefore minimizing the economic disadvantage of quenching and tempering. Through air cooling, yield strengths of 450-900 MPa (65-130 ksi) and tensile strengths of 600-1200 MPa (85-175 ksi), which are higher than those typically obtained for ferritic-pearlitic steels are also achievable. In addition, it was discovered that using a low carbon content and carbon equivalent would increase weldability and formability.

The next goals that arose were to provide a minimum variation of properties over a wide range of section sizes and to obtain good toughness with high strength levels. In low-carbon steels, the impact resistance of upper bainite is inferior to that of lower bainite, so most of the developments have been based either on lower bainite or low-carbon acicular ferrite. The formation of these acicular transformation products is beneficial in obtaining high strength levels.

Those who worked with low-carbon steels also recognized that a lower carbon content resulted in improved toughness

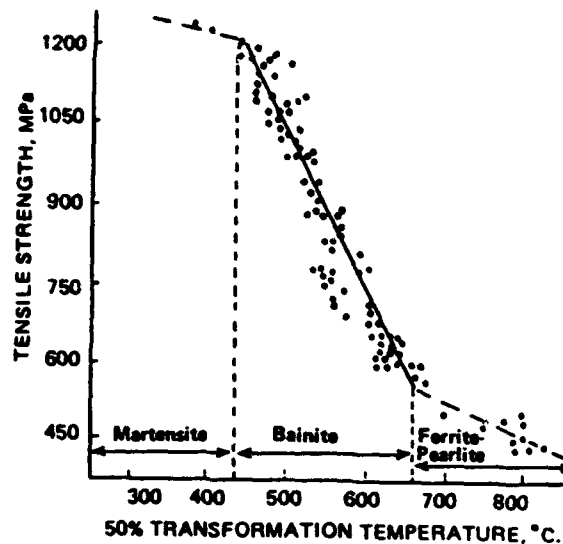


Figure 1.1 Effect of 50% transformation temperature on tensile strength of bainitic steels (2).

(1,2), compared with higher-carbon steels. An extreme example of a very low-carbon steel was the development of the 0.02% carbon, 3% nickel, 3% molybdenum, niobium-containing, aluminum-treated steels (12,13). These steels possessed a "bainite-like" microstructure with yield strengths of more than 700 MPa (100 ksi) and transition temperatures below -75°C . However, these steels are uneconomical for general structural engineering (2). During this time, other alloying combinations (Mn-Mo-Nb, Ni-Cr, Mo-B and Ni-Cu-B) were also investigated, but they showed lower strength levels (14-16).

As mentioned above, structural steels required for

critical applications such as shipbuilding, arctic pipeline, off-shore oil drilling platform and construction equipment require not only good weldability, but also high strength and notch-toughness in addition to a low impact transition temperature. High-strength steels were developed with these applications (primarily Navy shipbuilding) in mind. Such steels have yield strengths of 550 to 700 MPa (80 to 100 ksi), and are known as HY-80 and HY-100. These steels were quenched and tempered to provide good, low-temperature toughness with minimum strength levels of 80 ksi and 100 ksi, respectively (17).

To some extent, these alloys have been replaced by more economical (18,19) precipitation-hardened HSLA-80 and HSLA-100 steels that have been studied thoroughly by several investigators (17,19,20,21). The transformation behavior and mechanical properties of these HSLA steels will be compared with those of this study.

In recent years, somewhat similar alloys to those of McEvily et al. (12,13) have been studied by DeArdo and his co-workers (22), who refer to them as "ultra-low carbon bainitic" (ULCB) steels. They have investigated the possible candidacy of the ULCB steels to replace the traditionally heat-treated HTS and HY steels. The HTS steels are C-Mn steels which, like the HY steels, rely on

carbon for strength, except that the latter also contain alloying elements required for adequate hardenability. Both the HTS and HY steels exhibit base-plate properties which are adequate for their intended purposes. However, the HY steels suffer from poor weldability (i.e., susceptibility to heat-affected-zone cracking under all conditions) which is caused by both high carbon content and high carbon-equivalent values (22). Figure 1.2 shows the critical importance of low carbon contents in improving the weldability and weldment toughness in steels.

DeArdo et al. showed that ULCB steels are very attractive for heavy-section applications and do not need to be heat treated to obtain good mechanical properties. The ULCB steels obtain their high strength and high resistance to both ductile and brittle fracture through the application of specially designed alloys and advanced thermomechanical processing. DeArdo's group also suggested that these steels show a very good combination of strength and toughness in the as-hot-rolled condition, e.g., a 745 MPa (121 ksi) yield strength with a ductile-brittle transition temperature below -70°C .

Although there has been some research done to explore the isothermal transformation behavior of this unique class of ferrous alloy (12,13), the continuous cooling

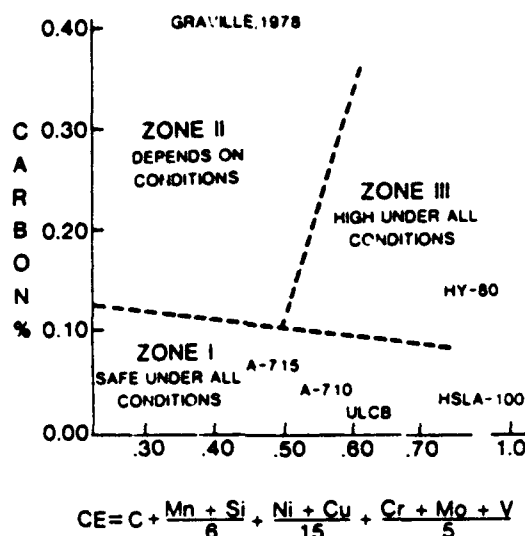


Figure 1.2 Effect of carbon level and carbon equivalent (CE) on the susceptibility to heat-affected-zone (HAZ) cracking (22).

transformation kinetics and the microstructural characterization of these steels are yet to be understood. Also uncertain is how the transformation behavior of these steels compares with low-carbon, copper-bearing alloys. The aim of this project is to address these needs, and the results of this study will be compared with the HSLA copper-bearing class of steels.

1.2 Purpose of The Project

Great demand for high-strength, tough steels with good weldability encourages metallurgists to develop new alloys.

The common ground for many of the newest HSLA steels is low carbon content. This is because carbon and/or carbides in high-temperature transformation structures, such as pearlite, are detrimental to weldability, formability, impact toughness and ductility (10). The formation of carbides can be avoided by reducing the carbon content of the steel.

As mentioned in the previous section, there is a wide range of compositions to consider for the new steels, with a wide range of mechanical properties. Although a great deal of attention has been paid to the intermediate transformation products (bainite) (10-16,23-34), more work needs to be done to obtain a better understanding of the transformation behavior of these products. In particular, the effect of chemical composition and cooling rate on austenite decomposition needs to be addressed.

The aim of this project is to explore the continuous-cooling-transformation behavior and microstructure of a unique class of ultra-low-carbon bainitic steels.

McEvily and his co-workers (12,13) studied the isothermal transformation and hardenability of similar steels.

Their isothermal transformation studies (Figure 1.3) showed that a high-temperature transformation product would form above a deep bay in the transformation diagram. Below the

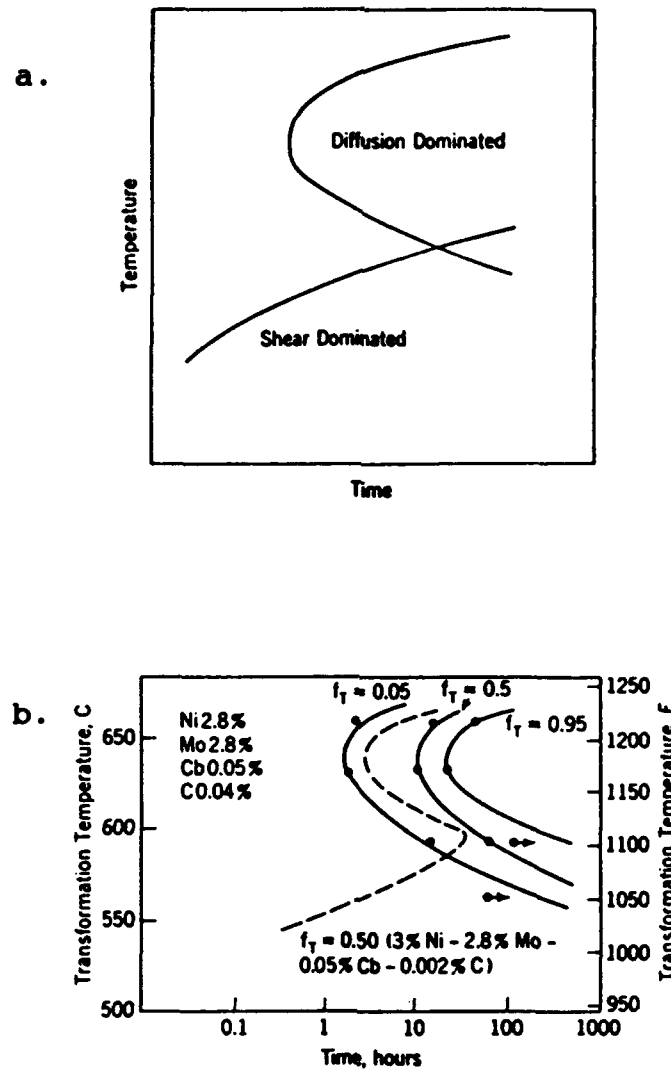


Figure 1.3 a. Schematic isothermal diagram indicating diffusion-dominated and shear-dominated austenitic transformation ranges.
b. Isothermal transformation diagram for high-nickel, high-molybdenum alloy where f_T refers to the fraction transformed at the indicated time (12).

bay, a shear-dominated transformation product would form. Based on their isothermal studies, they concluded that high strength (more than 700 MPa) and good toughness (about 240 J at room temperature) could readily be obtained upon air cooling. They suggested that these excellent properties are due to the strength of austenite and the shear transformation products.

Because of the high cost of alloying (2), no attention was paid to these steels until a few years ago, when DeArdo et al. (22) studied the structure and properties of similar steels. They placed the emphasis of their research on austenite conditioning as a way to improve the mechanical properties of these steels. Their results showed that the as-hot-rolled combination of strength and toughness was quite good, that is, a 745 MPa yield strength with a ductile-brittle transition temperature below -70°C . They also suggested that tempering will improve toughness but does not affect the strength of these steels.

Although there has been some work done to explore the transformation behavior of this unique class of steels, there is very little information available about their microstructural characterization. Even more noticeable is the absence of information about the continuous-cooling transformation of these steels.

It is very important to consider the fact that the paths followed by the austenite in the metastable field (austenite+ferrite) are largely different in the case of isothermal transformation from that of continuous-cooling transformation. Although the fundamental mechanisms of transformation are not dissimilar, the produced microstructures may be very different (35).

Recent and ongoing work at the Colorado School of Mines (20,36-38) has revealed interesting austenite phase-transformation behavior in low-carbon (0.04-0.06%) copper-containing steels. These steels contain approximately 1.5% Mn, 0.7% Cr, 0.4% Mo, and 1 to 3.5% Ni (20,37,38). Figure 1.4 shows a typical continuous cooling transformation diagram for these steels generated by Colvin (36). As shown in the diagram, acicular ferrite and granular ferrite form over a wide range of intermediate cooling rates, while polygonal ferrite and Widmanstätten ferrite form during slow cooling, and martensite forms as a result of fast cooling rates.

As mentioned before, it is desirable to achieve similar properties upon cooling over the widest possible range of rates for the production of heavy section sizes (39). In order to obtain this goal, polygonal ferrite transformation ought to be delayed, and the bainite shelf on a CCT diagram

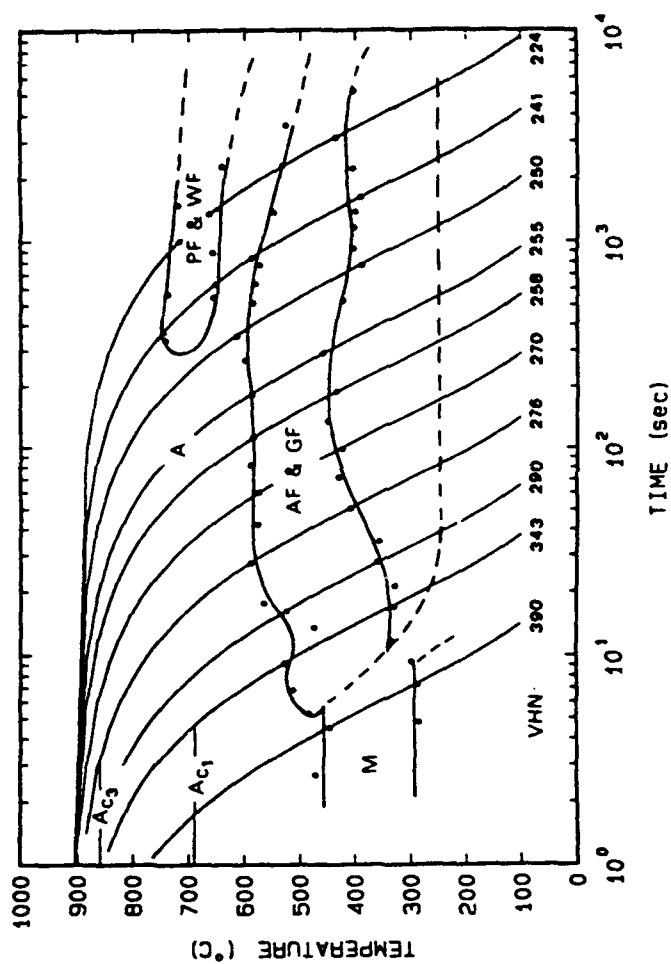


Figure 1.4 Continuous-cooling-transformation diagram for a low-carbon, copper-containing HSLA steel (37).

should be as horizontal (flat topped) as possible (36,39). The steels under investigation are assumed to be capable of these requirements (12,13,22).

1.3 Effect of Alloying Elements in HSLA Steels

1.3.1 Introduction

Plain-carbon steels are satisfactory in applications requiring moderate strength and toughness, and where mild environments are encountered. The relatively low hardenability of these steels limits the strength which can be obtained, except in fairly thin sections. However, most of the limitations of plain-carbon steels may be overcome by the use of alloying elements (40). This is true in the case of HSLA steels as well. Figure 1.5 shows the hardening effect of various alloying elements present in solid solution within iron (41). Some of the most important purposes of adding alloying elements to HSLA steels are (2):

- To increase hardenability;
- To improve strength at ordinary temperatures;
- To improve mechanical properties at either high or low temperatures;
- To improve toughness;

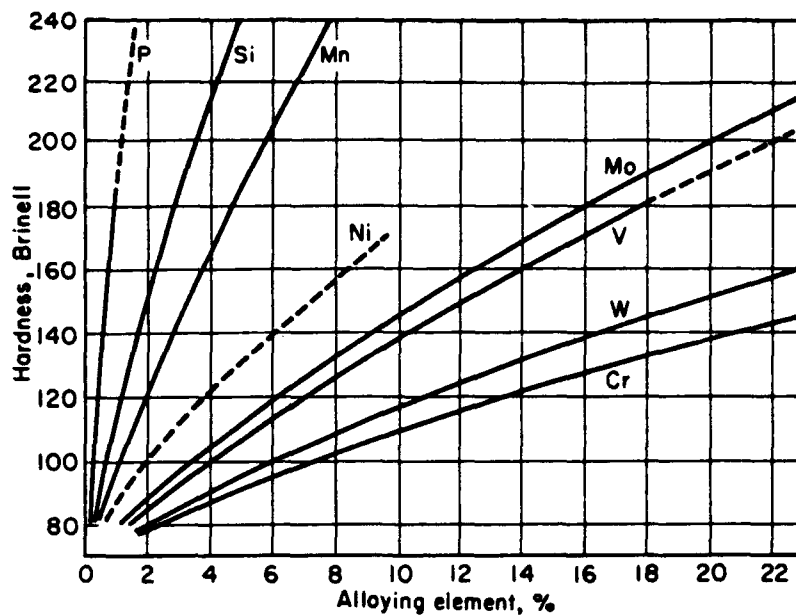


Figure 1.5 Solid solution hardening effects of the various elements as dissolved in pure iron (41).

- To improve wear resistance, and;
- To improve corrosion resistance.

Alloying elements in general act to depress the austenite-to-bainite transformation temperature, primarily through their effects on the diffusion of carbon which is necessarily associated with high-temperature transformations (e.g., pearlite) (14). The effect of some of the most popular alloying elements used in HSLA steels will be discussed briefly in this section.

1.3.2 Carbon

Carbon is the least expensive alloying element. It is present in iron as an interstitial solute atom. Carbon has a maximum solubility of 4.3% in austenite at 1148°C; by contrast, its solubility in ferrite at room temperature is essentially zero. Therefore, quenching a solid solution of carbon in face-centered-cubic (fcc) iron to a temperature below the eutectoid temperature results in the formation of body-centered-cubic (bcc) iron and the precipitation of carbon in the form of iron-carbide particles, or the formation of the non-equilibrium martensite phase which is supersaturated with carbon (11,41). Independent of microstructure type, increasing carbon content in steel produces a higher strength product.

Although increasing the carbon content of steel has positive effects on strength and hardness, a high carbon level has detrimental effects on impact strength and transition temperature. These negative effects explain the industry-wide tendency to reduce the carbon content and the carbon equivalent of HSLA steels to the lowest levels possible (2,12,39). Figure 1.6 shows the effect of carbon content on impact-transition temperature curves of ferrite-pearlite steels.

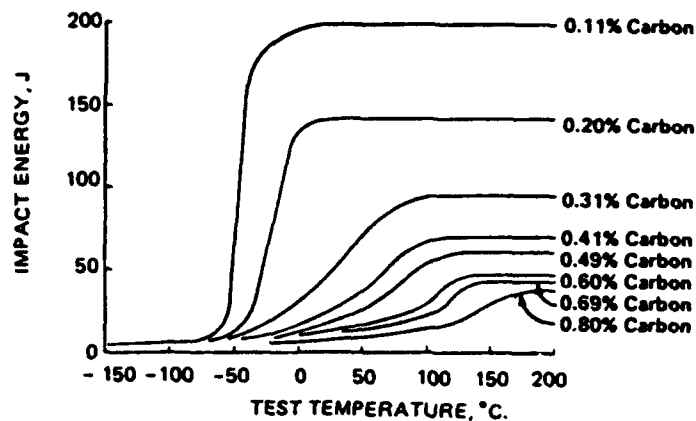


Figure 1.6 The effect of carbon content and hence the pearlite content on impact-transition temperature curves of ferrite-pearlite steels (2).

1.3.3 Chromium

Chromium, a bcc metal, has at most 20% solubility in austenite with 0.5% carbon and unlimited solubility in ferrite (41). Chromium shrinks the austenite phase field in accord with its role as a ferrite stabilizer (Figure 1.7). In addition to being a moderate carbide former, chromium strengthens the ferrite matrix and controls grain size (19). It also increases resistance to corrosion (41).

1.3.4 Copper

Copper was originally added to steels to provide resistance to atmospheric corrosion (42). Later it was

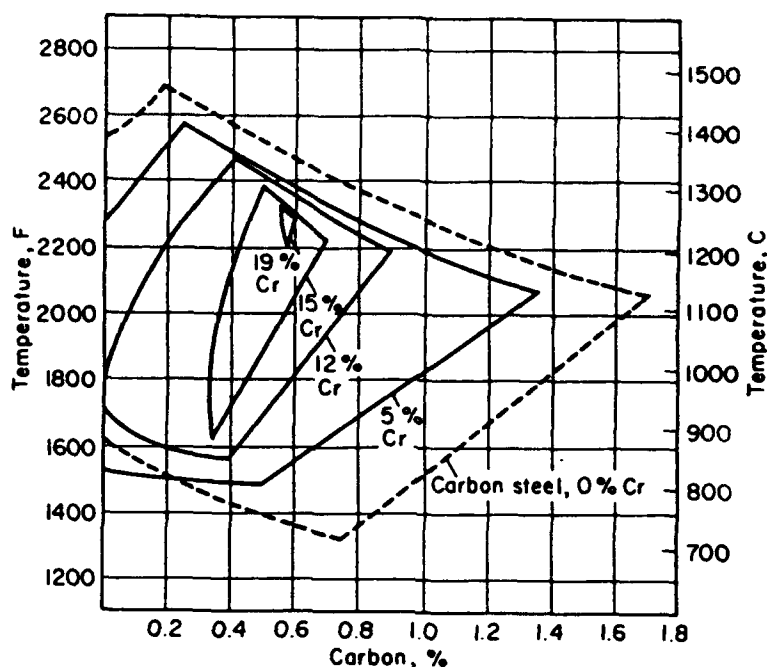


Figure 1.7 Effect of chromium content on the single-phase austenite region for Fe-Cr-C alloys (41).

found that this element could be used as a moderate solid-solution strengthener and also as a precipitation strengthener. It has fcc an structure and is an austenite stabilizer. Since its solubility in ferrite is much less than in austenite, precipitation may occur during the austenite to ferrite transformation, or in the ferrite after it forms (43).

1.3.5 Manganese

Manganese has unlimited solubility in austenite, while its solubility in ferrite reduces to 3% at room temperature (40). Manganese is a very common alloying element in steel because it is inexpensive, and because it improves the hardenability of steel by inhibiting the transformation of austenite to polygonal ferrite (12-16). Manganese also improves the strength of steel by its solid-solution strengthening effect.

1.3.6 Molybdenum

This bcc metal is a ferrite stabilizer and a strong carbide former. Its maximum solubility in ferrite is 37.5%, which lowers with decreasing temperatures (40,41). Molybdenum strongly increases hardenability of steel by delaying austenite decomposition at high temperatures (650 to 500°C), Figure 1.8. In addition, it provides significant solid-solution strengthening (12-15,45). It reduces the susceptibility of steel to temper embrittlement and increases secondary hardening, while retarding softening during tempering of steel (44,46). These effects are shown in Figures 1.9 and 1.10. Molybdenum decreases the activity

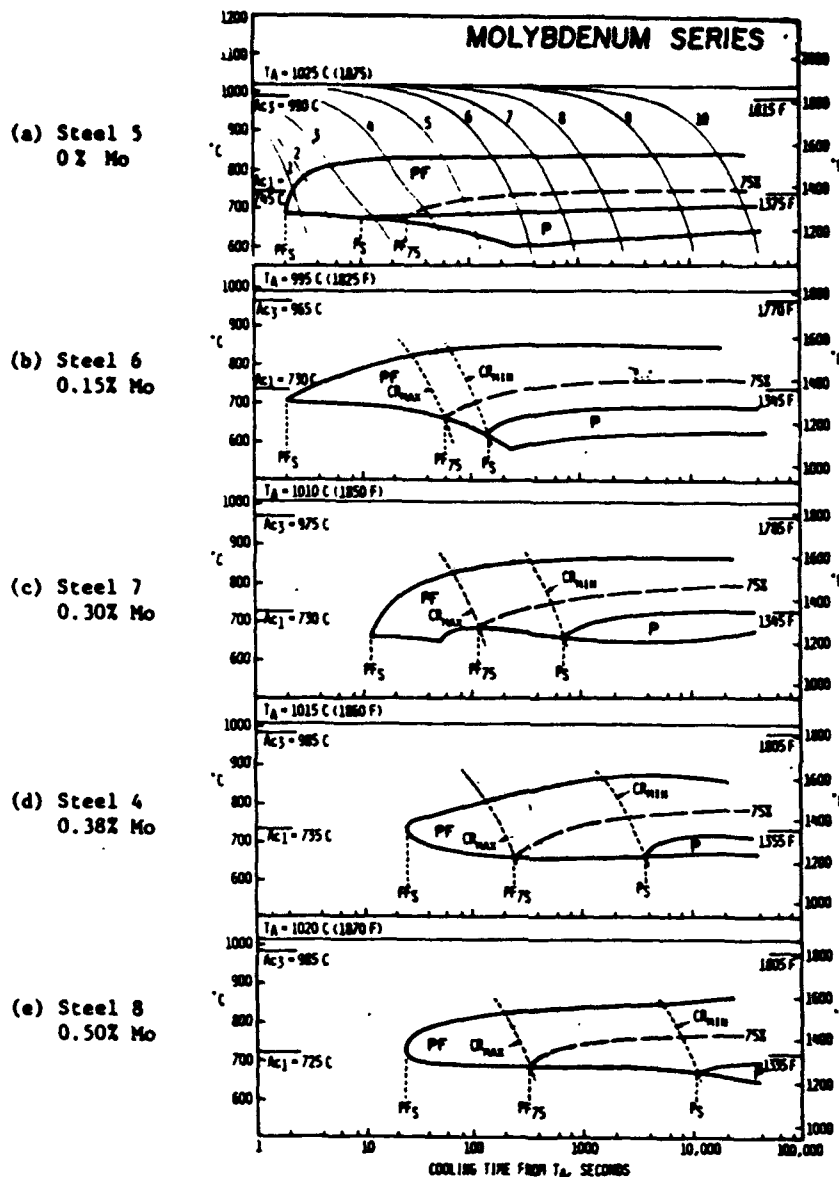


Figure 1.8 Upper portions of CCT diagrams for steels alloyed with Mo. Nominal phase composition is 0.05 pct. C, 0.9 pct. Mn, 1.2 pct. Si and 0.5 pct. Cr (all wt. pct.). PF is polygonal ferrite and P is pearlite (52).

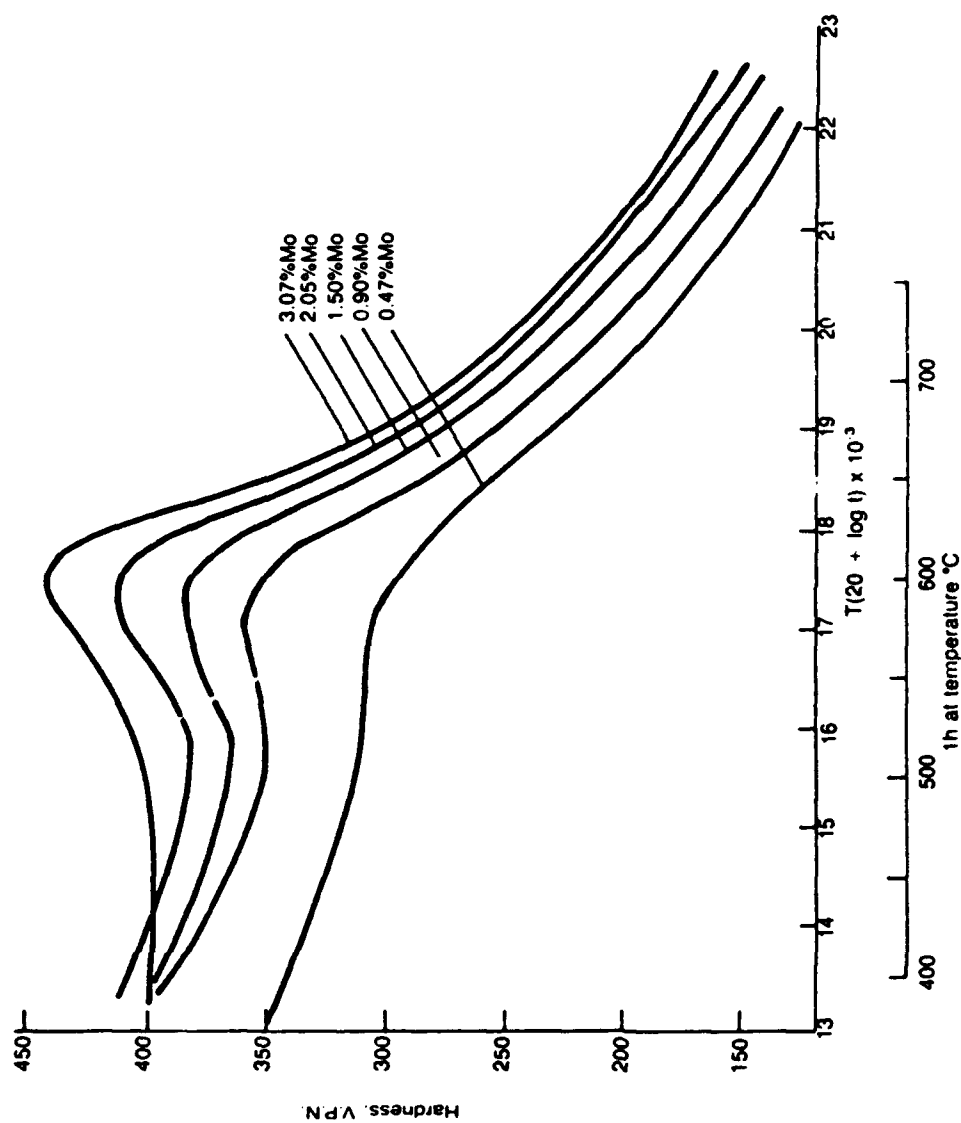


Figure 1.9 Retardation of softening and secondary hardening during tempering of steels with various molybdenum levels (44).

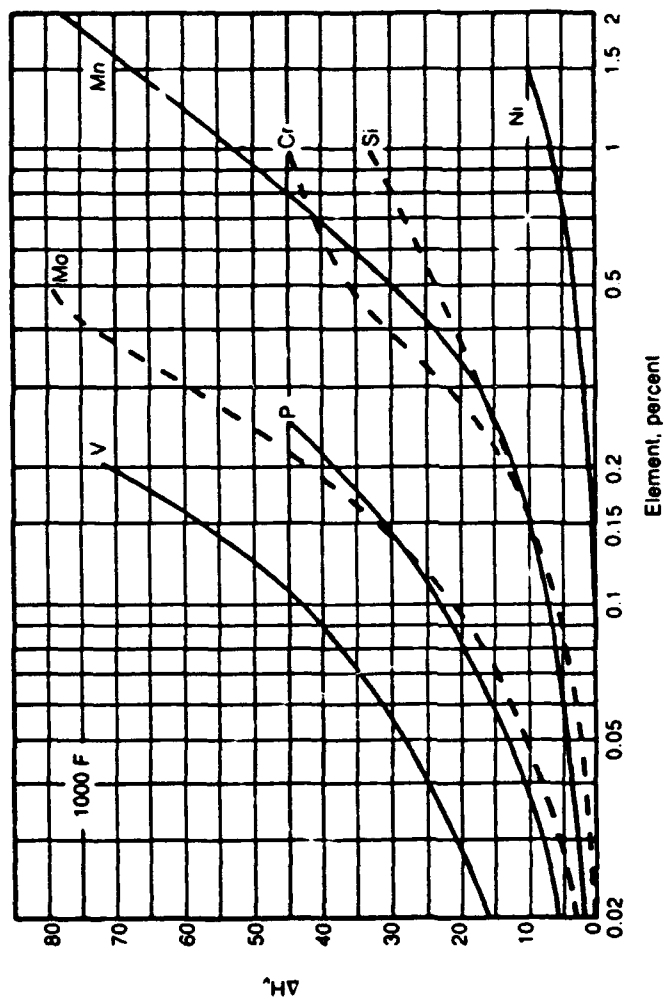


Figure 1.10 Effect of alloying elements on the retardation of softening during tempering at 540°C (1000°F) relative to Fe-C alloys (44).

coefficient of carbon in austenite (47) and in ferrite (48), therefore retarding the formation and growth of ferrite (49,50).

1.3.7 Nickel

Nickel is an austenite stabilizer with unlimited solubility in austenite and 10% solubility in ferrite at room temperature, irrespective of carbon content (40). It has an fcc structure and, like manganese, depresses austenite transformation temperatures upon cooling. It also increases the hardenability of steel primarily by delaying the proeutectoid ferrite transformation (13,44). Nickel improves the impact properties of steel by its presence in solid solution in ferrite (12,35,41). It also increases steel strength by grain refinement in addition to solid- solution hardening (51).

1.3.8 Niobium

Three important roles of Nb(C,N) precipitation in HSLA steels containing Mn, Mo and Nb are: inhibiting austenite recrystallization during rolling, precipitation strengthening and contributing to improved hardenability

(15).

Niobium is mainly used for grain refinement purposes and also precipitation strengthening (42). Precipitation of Nb(C,N) can retard or inhibit recrystallization and subsequent grain growth of austenite, which in turn will lead to grain refinement of ferrite.

Like copper, the solubility of niobium in ferrite is much less than in austenite. Therefore, particles may precipitate during transformation of austenite to ferrite or within the ferrite after it forms (42). If used in combination with manganese, niobium's solubility in austenite will increase, which in turn further lowers austenite transformation temperature and also limits the premature precipitation of niobium in austenite. The presence of nickel in conjunction with manganese may also tend to limit the premature precipitation of niobium (53).

1.3.9 Nitrogen

Nitrogen is an interstitial alloying element, and like carbon, this element is an austenite stabilizer and improves both the yield and tensile strength of steel. At the same time, nitrogen promotes brittle fracture and increases the impact transition temperature. Because of the high

electronegativity of nitrogen, nitrides usually form before carbides and are very stable.

1.3.10 Silicon

Even though silicon is a ferrite stabilizer element, it also stabilizes retained austenite by retardation of cementite formation (23-26,40,41). Silicon contributes to solid-solution hardening and increases both yield and tensile strength. It also improves the resistance of steel to oxidation.

1.4 Strengthening Mechanisms in HSLA Steels

1.4.1 Grain Refinement

For a given composition of steel, it has been shown that the most important factor controlling strength and associated ductility is grain size (54,55). Pickering (3) calls this "the unique feature of grain-size strengthening" which is "the only strengthening mechanism that also increases the toughness." According to the Hall-Petch equation (56,57), the yield strength of a ferritic steel increases with decreasing grain size (58):

$$\sigma_y = \sigma_i + k_y d^{-1/2} \quad (1.1)$$

where σ_y = yield stress

σ_i = stress needed to move dislocations in the grains (friction stress), an experimental constant

k_y = an experimental constant

d = grain diameter

A typical plot is given for a 0.1%C mild steel in Figure 1.11, where the linear relationship of the impact transition temperature with $d^{-1/2}$ has been described and the effect of refining the grain size on both the yield strength and the impact transition temperature is shown. Armstrong (59) has an excellent analysis on the influence of grain size on several mechanical properties of

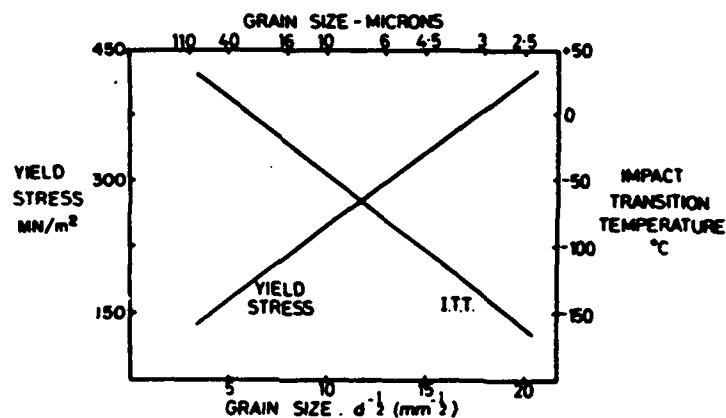


Figure 1.11 Effect of grain size of ferrite on yield stress and impact transition temperature of a steel containing 0.1%C, 0.5%Mn, 0.2%Si and 0.006%N (3).

polycrystalline materials, based on the Hall-Petch relations. He concludes that hardness, fatigue strength and yield strength, as well as ductile-brittle transition temperature, usually increase by decreasing the grain size. The effect of grain size on creep properties of materials are dependent on the temperature and the applied stress.

A simple explanation of how grain size can affect mechanical properties of materials is that grain boundaries act as barriers to dislocation motion. However, Irvine (60), in reference to upper and lower bainite, states that: "In the bainitic structure, the yield strength is related to the bainitic ferrite grain size, whereas the impact strength is more dependent on the prior austenite grain size." Irvine goes on to explain that the low-angle boundaries of the bainitic ferrite plates are not as effective barriers to propagation of cracks as the high angle boundaries and the carbide aggregates. In lower bainite, the situation is different, because of the presence of carbide within the ferrite grains, which impedes fracture. Here, the prior austenite grain size becomes less important and the microstructural features of fine ferritic grain size plus the carbides precipitated within the ferrite control the fracture characteristics (60). Different methods of grain refining have been used, such as controlled rolling, thermal

cycling and microalloying with aluminum, niobium, titanium, vanadium, etc.

By the late 1960s and early 1970s, it was widely recognized that attractive combinations of yield strength and toughness could be developed at modest cost using properly processed steels with unique compositions (61-63). Such methods as controlled rolling at a low finish-rolling temperature were found to produce a fine austenite grain size, and consequently a fine ferrite grain size, while retaining some potential for precipitation strengthening (2).

Tanaka et al. (64) defined controlled rolling in three stages. Their definition is based on the fact that there should be a definite difference in the physical properties of steel between deformation in the austenite region and that in the austenite-ferrite region.

Stage one includes simultaneous deformation and recrystallization. This stage takes place at high temperatures, above 1000°C, where dynamic recrystallization occurs concurrent with deformation. During this stage, there is a critical amount of deformation below which recrystallization will not be completed. A deformation level below 8% causes partial recrystallization as well as strain-induced grain boundary migration, which will result

in the formation of a mixed austenite grain structure. Completing this stage with a total reduction of about 60% gives a final grain size not smaller than $20\mu\text{m}$.

The second stage consists of low-temperature (950°C to A_{r3}) deformation of austenite, during which austenite grains become progressively more flattened with increasing amounts of deformation, resulting in an increasing surface-to-volume (S_v) ratio for the austenite grain structure. Increasing S_v results in increased ferrite nucleation rates, giving rise to a finer ferrite grain size in air-cooled specimens. An increase in the amount of deformation will increase the number of deformation bands, hence finer and more uniform ferrite grains will form. At this stage, variations in deformation temperature and the amount of deformation have a minor effect on yield stress and almost no effect on tensile strength.

Deformation in the austenite-ferrite region, the third stage, takes place at temperatures below A_{r3} . Deformation at this stage has a much larger influence on the mechanical properties than the deformation present in the second stage. IN addition with an increase in the amount of deformation, both yield and tensile strengths increase noticeably and impact transition temperature decreases rapidly. These considerable improvements in the mechanical properties after

deformation in stage three are mostly attributed to substructural strengthening in the ferrite.

The other method of grain refinement uses thermal cycling between room temperature and some temperature above the A_{c1} . This technique has been widely used to refine the grain size of steel, therefore improving its mechanical properties. Conventionally, this method involves multiple austenitization steps above the A_{c3} , followed by quenching or air cooling (65), whereas a non-conventional technique involves partial reaustenitization between A_{c1} and A_{c3} (66). In addition, rapid heat treatment has also been suggested by some investigators (66,67). Figure 1.12 shows a schematic representation of these thermal cycles.

On the average, each ferrite grain transforms to an austenite grain upon complete austenitization, and upon cooling each austenite grain will transform into about three polygonal ferrite grains (68). However, Karlsson (65) has reported a reduction of six times in ferrite grain size after one-step austenitizing. He suggests that this reduction is almost independent of carbon content, being only slightly smaller in low-carbon material. His data also indicate that the refinement diminishes on further cycling. Similar results will be reported later in this thesis.

The effects of microalloying with aluminum, niobium,

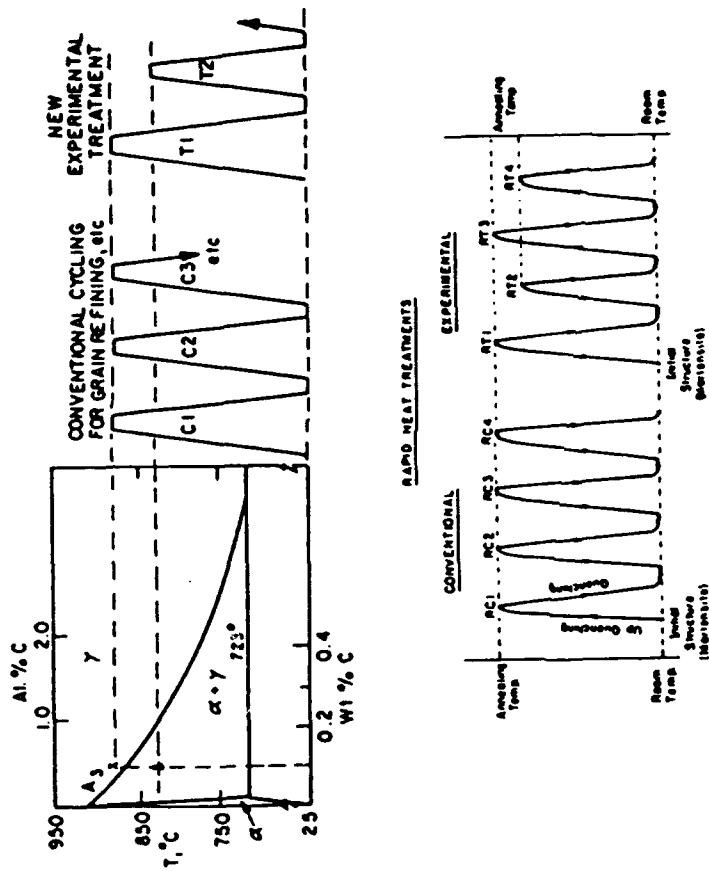


Figure 1.12 (a) Schematic representation of experimental thermal cycles in comparison with conventional heat treating cycles, referred to part of the Fe/C equilibrium phase diagram. (b) Schematic representation of conventional and experimental rapid heat treating cycles (66).

vanadium and titanium have been studied by many investigators (15,16,42,51-53,69-74). Aluminum was the first element to be used in combination with nitrogen as a grain refiner (2), with the effect that fine aluminum nitride particles control austenite grain growth (44).

But niobium is the alloying element which has attracted the most attention (2), and is considered to have the most beneficial effect on the strength-toughness combination of steels (63). In addition to grain refinement, niobium carbonitride particles also inhibit austenite recrystallization during rolling. Niobium in solution enhances hardenability, and it can promote precipitation strengthening (73). However, a substantial amount of niobium must remain in the austenite solid solution after hot rolling in order to maintain its effectiveness in the areas of hardenability and precipitation strengthening (60). As precipitation of Nb(C,N) in the austenite removes niobium from the solution, there will be effectively less niobium in solution to further precipitate in ferrite. In particular, Coldren et al. (15) observed a loss in acicular ferrite hardenability when the amount of niobium dissolved in austenite decreased below ~0.03%.

Furthermore, the effect of precipitates on austenite recrystallization and grain growth of austenite or ferrite

is strongly particle size-dependent. Small precipitates ($\leq 200\mu\text{m}$) are most effective in preventing recrystallization (61).

Vanadium is another strong carbide former that is used primarily as a microalloying element for precipitation hardening. Vanadium carbide has one of the highest solubility levels among other carbides and nitrides in steel (75) and tends to form interphase precipitates of vanadium carbonitride particles in polygonal ferrite (76).

Titanium reacts with interstitial atoms to form a variety of compounds. Some of these precipitates are desirable and some have detrimental effects on the mechanical properties of steel. Reasons for using titanium as a microalloying element include grain refinement and precipitation strengthening (74).

Figures 1.13 and 1.14 show the effects of niobium, vanadium and titanium on grain refinement and precipitation hardening as well as on continuous cooling transformations, respectively (62,77).

1.4.2 Solid-Solution Strengthening

Strengthening of pure iron by the introduction of alloying elements is well known (41,78). The effect of

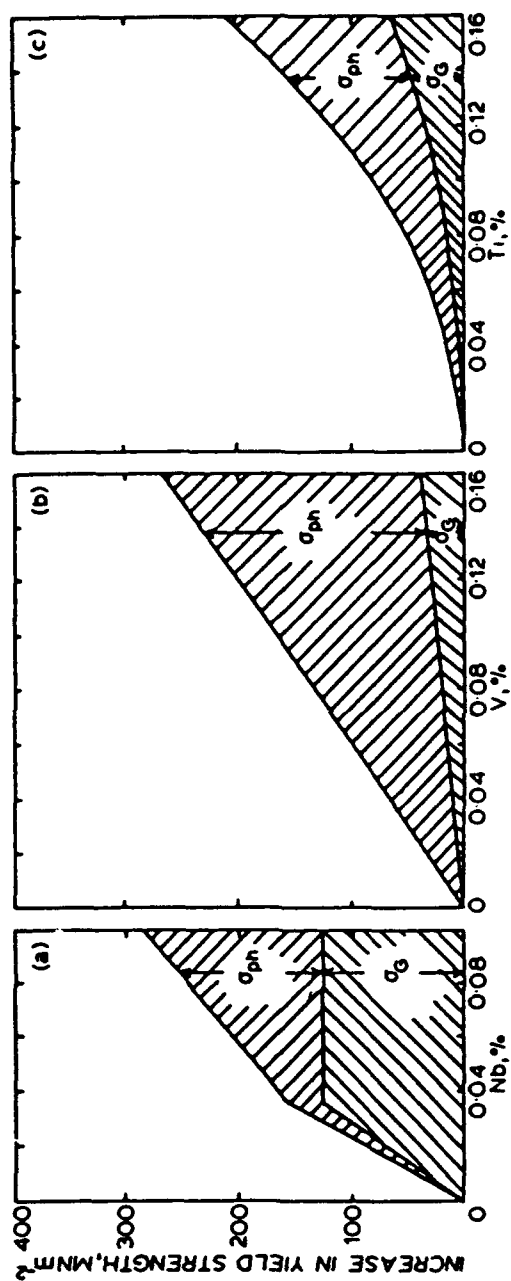


Figure 1.13 Correlation between microalloy content and increase in yield strength. Increase in strength is due to grain refinement (σ_g) and precipitation strengthening (σ_{ph}) (77).

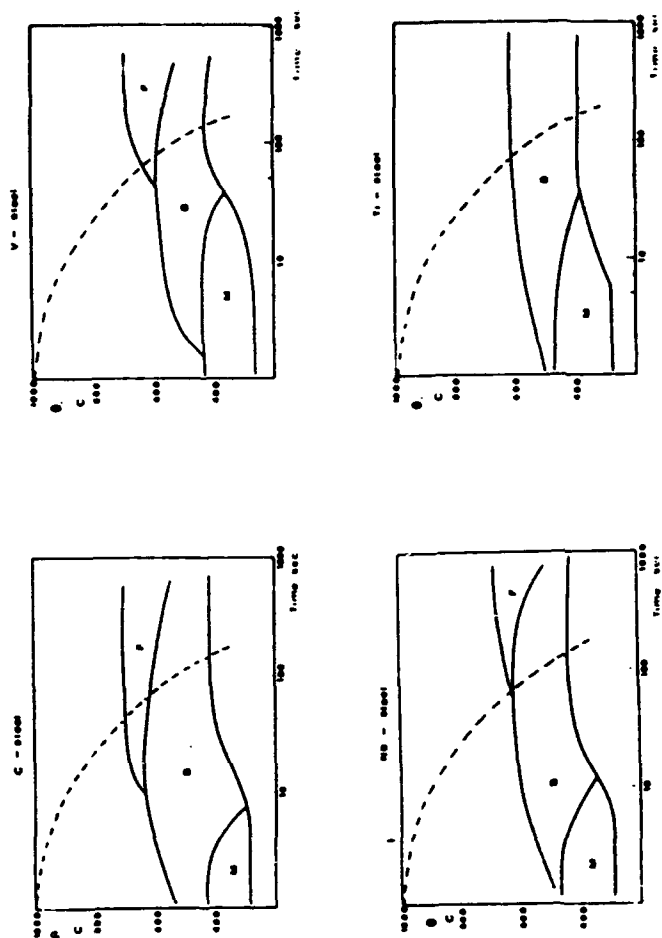


Figure 1.14 Effect of V, Nb, and Ti on CCT curves of a steel containing 2.5% Mn (62).

alloying elements on the mechanical properties of iron is the consequence of the influence exerted by the solution of those elements in the iron lattice. The result is a solid solution, which occurs in two types: substitutional and interstitial. In the case of a substitutional solid solution, the solute atoms take positions that are normally occupied by solvent atoms. By contrast, interstitial solid solution happens when solute atoms do not displace solvent atoms but enter the holes, or interstices, between the solvent atoms (79).

Generally, substitutional atoms cause symmetrical distortion of the solvent lattice, therefore the strengthening effect is relatively moderate. Asymmetrical distortion caused by interstitial solutes has a more pronounced effect on solid-solution hardening (3). However, there are a number of factors that contribute to solid-solution strengthening. According to Dieter (80), the most important ones are relative size factor, relative modulus factor, electrical interaction, chemical interaction (Suzuki interaction) and configurational interaction (Fisher effect). Relative size factor or the size misfit between solute and matrix atoms gives an approximate idea of the amount of strengthening that can be obtained, although some elements such as nickel and manganese provide more

strengthening than would be expected by their misfit (4). Detailed explanation of each of these factors is outside the scope of this brief review.

Although the overall effect of substitutional solid-solution hardening is considered to be small (81), Speich and Swann (82) showed that the solid-solution hardening effect of nickel (at 20%Ni) is responsible for 75% of the overall strength of a martensitic alloy with 0.0034 wt% of carbon. Figure 1.15 shows the effect of various substitutional solute atoms on the yield strength of high-purity polycrystalline ferrite at room temperature (78). Further discussion of solute atoms was presented in previous sections.

1.4.3 Precipitation Hardening

Precipitation hardening is defined (44) as "hardening caused by the precipitation of a constituent from a supersaturated solid solution." This strengthening phenomenon occurs when a solid solution at elevated temperature is cooled to a lower temperature, and precipitates form because of the decrease in solubility (80). More conventionally, a solid solution is quenched to "freeze in" the solute atoms, and a subsequent aging or

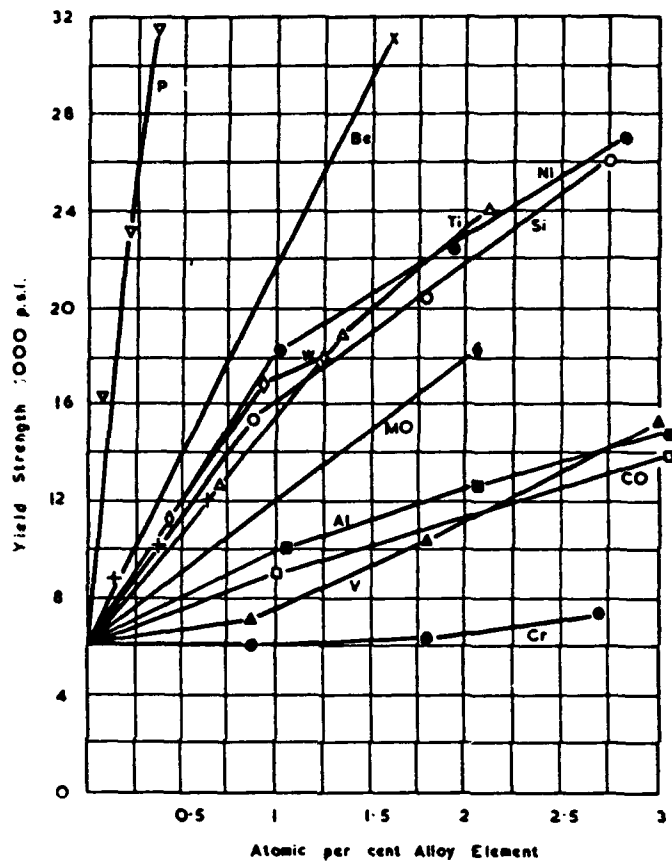


Figure 1.15 Yield strength versus atomic percent for various alloying elements (79).

tempering treatment results in precipitation. Any particles that form at high temperatures in austenite do not cause strengthening because they are typically too coarse and widely spaced (4). Additionally, these particles lose coherency with the matrix when the austenite-ferrite transformation interface sweeps through the structure (15). However, these precipitates are effective in controlling grain growth. The particles that form at low temperatures in austenite, during transformation at the austenite-ferrite interface, or in ferrite during cooling are effective strengtheners (4).

The ratio of carbide-forming element to carbon is also important for precipitation hardening. Elements such as vanadium, niobium, titanium, molybdenum, tungsten and chromium are used to strengthen HSLA steels. A number of investigators have studied the effect of these alloying elements on mechanical properties of HSLA steels (15,19,39,42,53,69). Some of these effects have already been discussed.

Copper has been used in HSLA steels to take advantage of its precipitation in absence of carbon (19,37,42,53,83). Copper taken into solid solution in the austenite can be precipitated as fine ϵ -copper particles in ferrite. To increase the degree of supersaturation of copper and to

avoid its premature precipitation in austenite, faster cooling rates (84) and lower transformation temperatures (53) have been recommended.

1.4.4 Transformation Strengthening

Substantial substructural strengthening produced by transformation is largely due to the temperature at which the transformation occurs (3,12). In general, lower transformation temperatures, result in a greater strengthening effect. Most strengthening mechanisms can be intensified by lower transformation temperatures (3). In particular, it is possible to produce finer grain sizes in the transformation product, increase the dislocation density and the precipitate density, and make the precipitate size finer if transformation temperature is lowered.

Because of a substantial decrease in the diffusion rate of alloying elements at lower transformation temperatures, there is a tendency to retain solute in solution, and subsequently solid solution strengthening will increase. Complex results (3) arise with the interaction of all of these effects. Although all of the aforementioned factors contribute to strength, each has its own unique impact on toughness and ductility (3).

An attractive combination of strength and impact properties can be obtained in low-carbon steels with acicular-ferrite structures (15,16,85). Acicular ferrite is defined by Smith et al. (16) as:

"a highly substructured, non-equiaxed ferrite that forms on continuous cooling by a mixed diffusion and shear mode of transformation that begins at a temperature slightly higher than the upper bainite temperature transformation range. It is distinguished from bainite for two reasons: (1) the limited amount of carbon available results in there being only a small amount of carbide present, and (2) the term bainite has been applied to such a wide variety of structures that it is somewhat indefinite."

One advantage of the acicular ferrite transformation is its function as a catalyst for aiding both the grain refinement and the precipitation hardening components of the strengthening mechanism, relative to those components obtainable from ferrite-pearlite steels. Because of its relatively low transformation temperature, acicular ferrite is characteristically fine-grained and contains a substructure with a high dislocation density (12,16). The low transformation temperature also leads to a slower rate of precipitation, which is easier to control. This controlled precipitation will result in more homogeneous properties throughout the plate thickness (16).

High yield strengths, in addition to excellent fracture toughness and weldability, are obtainable at unusually low carbon levels. This is due to the great potential of the

acicular ferrite matrix to extract large strengthening components from grain refinement, generating a highly dislocated substructure and aiding precipitation strengthening (15,16).

Proper alloying additions facilitate the formation of bainitic structures upon continuous cooling by suppressing the polygonal ferrite transformation. Alloying increases strength by lowering the transformation temperature, which in turn refines the bainitic structure and increases the dislocation density (39) and the solid solution hardening effect. In particular, the elements molybdenum, manganese (16) and niobium (53) help generate the acicular ferrite structure.

According to Morcinek et al. (85), an analysis of the strengthening mechanisms operating in low-alloy steels suggests that the two mechanisms which contribute the most are dislocation and precipitation strengthening. Although they suggest that the strengthening component related to grain refinement can be included with the dislocation contribution, others (2,86) believe that grain boundaries are the most effective barriers in resisting fracture. In the case of martensitic structures, internal twins also play an important role in strengthening (81). Bainitic lath size and dissolved carbon (which probably strengthens by

interacting with the dislocations) are other features which, in addition to the previously stated mechanisms, control the strength of bainitic steel (2,72). For example, Gemmill (87) states that hardening may occur by an interaction between solid-solution elements and dislocations, and he refers specifically to the influence of both manganese and nitrogen on the high-temperature properties of carbon-manganese steels.

Discussing transformation strengthening in steels without considering the martensitic transformation is incomplete. The diffusionless shear-type transformation of austenite to martensite is one of the most common strengthening mechanisms used in steels (80). The high strength and often brittle nature (88) of martensite in steels is an indication of strong barriers to dislocation motion in this structure (80). Although substructural strengthening in martensitic transformation is responsible for a great degree of the strength of martensite, this high strength can not solely be attributed to the fine structure of this austenite transformation product (88).

Hardness of martensite is very sensitive to the carbon content of steel for the values below 0.2%, and increases with an increase in carbon content (80). In the absence of carbon (0.0034%), Speich and Swann (82) showed that three

quarters of the strength of martensite in an Fe-Ni alloy with 20%Ni is due to the solid-solution strengthening effect of nickel.

1.5 Microstructure of Low-Carbon Steels

Decomposition of austenite at subcritical temperatures (11) produces a variety of microstructures that are somewhat well known to ferrous metallurgy students. As a result of this decomposition, pearlite, polygonal ferrite and Widmanstätten ferrite form at higher temperatures, while upper bainite, granular bainite (also called granular structure), granular ferrite, lower bainite and acicular ferrite form at intermediate temperatures or cooling rates. By decreasing the temperature even more or by increasing the rate of cooling, martensite forms.

There has been tremendous debate since the discovery of the intermediate structure over the mechanism of its formation and consequently the terminology, and there is no sign of compromise (27) between the three major theories developed by different groups of investigators. These three theories are summarized by Ironstone as follows:

1. The generalized microstructural definition considers bainite to be the product of a non-lamellar (as opposed to

pearlite), and more generally, a non-cooperative mode of transformation. It is believed that the two product phases form by diffusional nucleation and growth and differ continuously in composition from the matrix and each other.

2. The overall reaction kinetics definition views the bainite reaction as having its own C-curve on a TTT-diagram. No bainite can form above the highest temperature of this curve, which may be termed the kinetic- B_s temperature. At isothermal reaction temperatures near the kinetic- B_s , transformation of austenite to bainite is increasingly incomplete.

3. The surface relief definition usually considers bainite to be precipitate plates formed by shear at rates paced in some manner by diffusion. In some versions of this definition, however, the diffusion process may occur either before or after formation of a single product phase.

Upper bainite nucleates mainly at the austenite grain boundaries (11). The ferrite crystals are lath or plate-like in shape and second-phase carbide particles form between ferrite laths (28-30). According to the diffusion-controlled growth models, while ferrite laths grow, carbon is rejected into the austenite and makes the austenite rich in carbon. This enriched austenite usually transforms to carbide, although sometimes it remains between the adjacent

ferrite laths as pools of retained austenite or it transforms to high-carbon martensite (11). Bainite without carbide has been referred to as granular bainite (35).

In opposition to this model, Oblak and Hehemann (31) proposed that formation of both upper and lower bainite occurs by repeated nucleation of substructural units and rapid growth to a limited size. These subunits are thought to form by a cooperative transfer of lattice atoms, although not necessarily at the rate characteristic of martensitic reaction. The rate of formation is controlled by this multiple nucleation process. This suggestion has been supported by Bhadesia and Waugh (32).

Purdy and Hillert (33) suggested that subunits of bainite have similar orientations, whereas new units of plate martensite have new orientations, an indication that does not relate to the autocatalysis (martensitic) growth of bainite. On the other hand, they admitted that the elastic strains are important in both formation and growth of ferritic subunits.

Aaronson and Kinsman showed that the lengthening mechanism of proeutectoid ferrite and upper bainitic crystals is controlled by the formation and lateral movement of ledges. They also concluded that the subunits in the bainite sheaves lengthen by the formation and the diffusion-

controlled lateral movement of ledges. Both the "cessation of the thickening of sheaves" and the "rapid increase in the rate of sympathetic nucleation at lower temperatures" have been explained in terms of the "solute drag effect" (89). The special effect of alloying elements upon growth kinetics has been interpreted (90,91) in terms of a "drag effect" produced by the segregation of certain alloying elements to austenite-ferrite boundaries. Alloying elements that reduce the activity of carbon in austenite are particularly involved in such segregation. The tendency for segregation increases as the temperature decreases, especially because of the high concentration of carbon in the austenite at austenite-ferrite boundaries. When the drag effect is larger than the driving force for growth, an upper nose will develop in the TTT diagram (see Figure 1.3). On the other hand, as the drag effect approaches saturation and the driving force begins to overcome this effect, a bay will begin to develop in the TTT diagram. In this regard, the role of molybdenum in the formation of bay is very interesting. Based on thermodynamic arguments, it is expected that molybdenum displace the TTT diagram to shorter times compare with an iron-carbon alloy of the same carbon content, thereby decreasing rather than markedly increasing the hardenability of steel. It is the kinetic solute-drag

effect which provides the marked increase of hardenability in molybdenum-containing steels (90).

Regarding different forms of bainite, as the transformation temperature is lowered, bainite laths tend to become finer (10,11), but the carbide's orientation relationship with the ferrite does not change (30). The orientation relationship observed for the ferrite and cementite in upper bainite shows that they both are related through the parent austenite (30). As is true of upper bainite, prior austenite grain boundaries are the first sites for the nucleation of lower bainite, although nucleation tends to happen within the grains as well (11). The ferrite is acicular and the carbide particles form within ferrite laths at an angle of about $55-65^\circ$ to the long direction of the ferrite lath (30). The reason for the formation of carbide within the ferrite plates is that lower bainite is significantly supersaturated with carbon, therefore the formation of such a ferrite occurs in association with the precipitation of carbide particles (11). The orientation relationship between ferrite and carbide in lower bainite reinforces the opinion that carbide precipitates from ferrite (30). Therefore, even if the morphology of upper and lower bainite appears to be similar, these orientation relationships are very effective evidence

to distinguish between the two different microstructures (30).

Up-quenching experiments (34) have shown that there are major internal structural differences between upper and lower bainite, and that these differences appear as a result of alteration in the kinetics of formation and growth of these products. Even though there are some similarities between the bainitic (especially lower bainite) and martensitic structures, such as similar orientation relationships and surface relief effect, a true martensitic structure is diffusionless (92).

There is considerable confusion regarding the terminology for the microstructures formed at intermediate temperatures upon cooling (35,37), or by interrupted transformation of austenite to bainite and ferrite (93) in low-carbon alloyed steels. The equiaxed morphology of granular ferrite is similar to that of polygonal ferrite, however it is distinguished from polygonal ferrite by higher dislocation density content and finer structure (37). Granular ferrite formation is a diffusion-controlled transformation, and there has been evidence indicating that no shear is involved in the formation of this microstructure (37). Consequently, it should be differentiated from granular bainite or granular structure, since these structures are considered to be formed by a shear-type

transformation. Hehemann (89) believes that impingement is the reason for nonacicular nature of granular bainite. According to Bhadeshia and Christian (94), acicular ferrite is essentially identical to bainite, and the reason that its detailed morphology differs from conventional bainite is that the former nucleates intragranularly at inclusions within large austenite grains, while the latter nucleates initially at austenite grain boundaries (11,94). However, other investigators (95) believe that acicular ferrite is not identical to bainite. Although nucleation of acicular ferrite involves diffusion of carbon (and other interstitial atoms), there is no reconstructive diffusion during growth (96).

The observation of "steps" or "carbon-enriched residual austenite" can not be taken to imply that acicular ferrite is indeed Widmanstätten ferrite (as it is sometimes considered to be), because the mechanism for interface motion (step mechanism) does not relate to the mechanism of transformation. The enrichment of austenite, on the other hand, can happen during or after the transformation (94).

In an attempt to unify the definitions of the microstructures produced upon continuous cooling in very low-carbon steels, the bainite committee of the Iron and Steel Institute of Japan (ISIJ) has conducted some

investigations. The results appear in "Atlas for Bainitic Microstructures, Vol. 1", which lists some microstructures and the corresponding definitions (97).

Martensite in steels is formed by a diffusionless transformation of austenite (44) in an athermal or near-athermal fashion (33). This transformation involves a complete military (displacive) motion of atoms where (in principle) none of the atoms changes place with its neighbors (88). The martensitic transformation occurs by a shear mechanism, which is more likely to happen at lower temperatures. Surface tilting is an important feature of this transformation, which implies that there has been cooperative transfer of atoms across the advancing interface (44,98).

By lowering the transformation temperature or increasing the carbon content (or in general increasing the strength of austenite by alloying addition), the morphology of martensite changes from lath to plate morphology (44). The inhomogeneous shear in plate martensite is compensated by internal twinning, and therefore the dislocation density in plate martensite is not very high (44,81). In lath martensite, however, less internal twinning is observed and this is an implication of the compensation of the inhomogeneous shear by slip, which is in agreement with high

dislocation density of lath martensite (44,81).

Very low-carbon steels exhibit massive transformations at both high and low transformation temperatures. While the high-temperature product has been identified as massive ferrite, martensitic transformation is the product of lower transformation temperatures (99-106). Goodenow and Hehemann (107) reported similar observations of martensitic reactions in very low-carbon steels containing 9% Ni. They proposed "low-carbon martensite" as an alternative name for "massive martensite". Massive martensite refers to the blocks (108) of martensitic regions which consists of fine laths of martensite. Because of the similar shear directions (100) and low-angle boundaries of martensitic laths, only packet boundaries respond to etching. For this reason, the microstructure appears to consist of blocks of non-etched regions. Similar observations are presented in Chapter 3. Although it was once customary to use surface rumpling as an indication of martensitic reaction, Gordon Parr (101) suggested that this criterion was not a fair indication of the transformation mechanism.

To reduce the confusion about terminology, Krauss and Marder (109) proposed the term "lath martensite" for the martensitic structure produced in low carbon steels. This structure had been referred to as massive or low-carbon

martensite, or as high-temperature martensite. Acicular or low-temperature martensite was proposed to be called "plate martensite". Plate martensite forms at lower temperatures than lath martensite, and that is because of a high carbon content or other alloying elements in the steel. These two terms - lath martensite and plate martensite - have been accepted by many investigators.

2.0 EXPERIMENTAL PROCEDURES

2.1 As-Received Plates

2.1.1 Geometry and Chemistry

David Taylor Research Center (DTRC) supplied five plates with three different chemistries for this project. Plates were melted as experimental heats at the USS Technical Center, hot-rolled and air cooled. The general designation for these plates is 8902-8025. The specific designation for each chemistry and the dimensions of each plate are shown in Table 2.1:

Table 2.1 The specific designations and the dimensions of the as-received plates.

Designation	Dimensions (mm)	Dimensions (in)
A3	254x165x19	10x6½x¾
A4	267x203x19	10½x8x¾
B3	267x197x19	10½x7¾x¾
B4	260x197x19	10¼x7¾x¾
C4	254x197x19	10x7¾x¾

Ladle and product chemistries were provided by the USS Technical Center. Additional chemical analyses of the steel

products were performed at the Colorado School of Mines (CSM) and at the USS Technical Center, using emission spectroscopy and Leco interstitial analyses. The emission spectroscopy and interstitial analyses were conducted at various points on each plate, ranging from the edge to the center of the plate. The resulting compositions are reported in the Chapter Three. Because the compositions of the plates A3 and A4 and those of the plates B3 and B4 are identical, they were recorded by using the names A for A3 and A4, B for B3 and B4, and C for C4. It should be mentioned that plate C in the as-received condition was bent and contained a number of cracks.

2.1.2 General Structure

A long sample (half the width of each plate) was examined to reveal the possibility of any variation in general microstructure and grain size as a function of position (Figure 2.1). Standard metallographic techniques were employed in preparing the specimens. After cutting and grinding the samples to a 600 grit finish, they were polished with 6 μm diamond paste. After the remainder of the diamond paste was cleaned from the surface of the specimens, they were polished on microcloth with 1 μm and

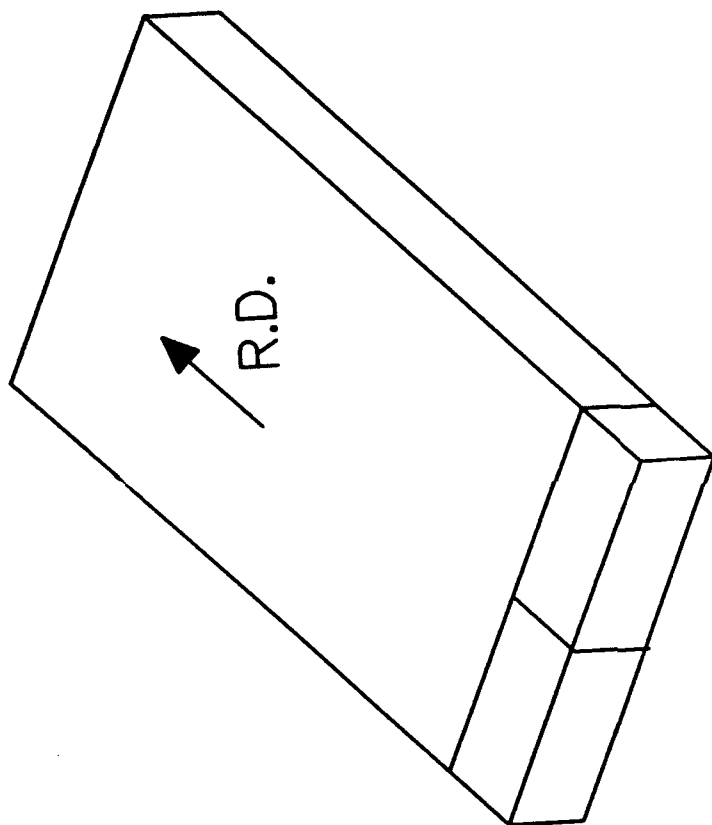


Figure 2.1 Blanks cut to examine the variation in the microstructure and grain size.

0.05 μm alumina powder. A solution of 2% nital was used to etch the above samples.

2.1.3 Inclusions

To determine the morphology, comparative number density, distribution and chemical composition of inclusions in the as-received plates, samples from all three steels were cut as shown in Figure 2.2. The above samples were then mounted in bakelite, polished using standard procedures, and examined with a Neophot 21 metallograph. In addition, energy dispersive spectroscopy (EDS) in the scanning electron microscope (SEM) was used to determine the chemistry of the inclusions observed in these steels.

2.1.4 Hardness Testing

Hardness testing was conducted on specimens from the as-received steels. One piece, the same as that shown in Figure 2.2, was cut from each plate and surface ground to a 600 grit finish. A Leco Hardness Tester (model R-600) was used to obtain the hardness measurements (110). The average of measurements on eight areas of each sample were taken, and reported as the Rockwell C Hardness Number (HRC).

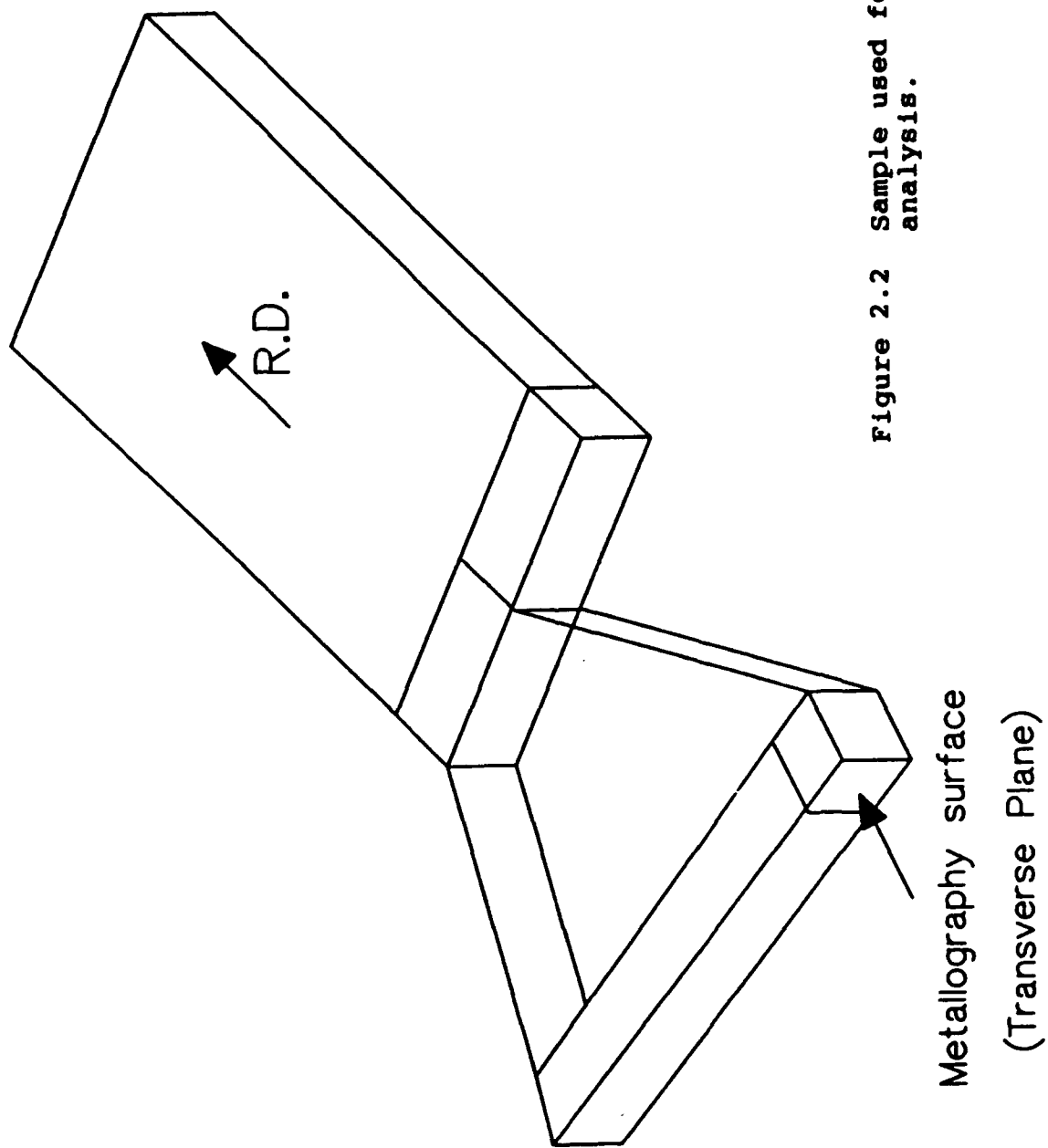


Figure 2.2 Sample used for inclusion analysis.

2.2 Ac_1 and Ac_3 Determination

In an attempt to study microstructural evolution in these steels, pieces from the as-received plates (as documented in the previous section) were used as the basis for further study. The temperatures at which austenite starts to form and finishes forming during heating, Ac_1 and Ac_3 , respectively, were determined. A thermomechanical test system, Gleeble 1500, was used for this study. This system will be described in detail in section 2.4.1.

As shown in Figure 2.3, cylindrical samples were machined from each steel to a size of 105 mm (4 inches) in length and 6 mm (0.25 inch) in diameter. These samples were then heated up to 1200°C at the rate of 3°C per second (111) and cooled to room temperature at a rate of 50°C/s. The dilatometric data and the corresponding critical temperatures for all the three steels, as well as the calculated data, are given in the Chapter 3. Microstructures, prior austenite grain sizes and hardness values of specimens cooled at various rates were examined in detail.

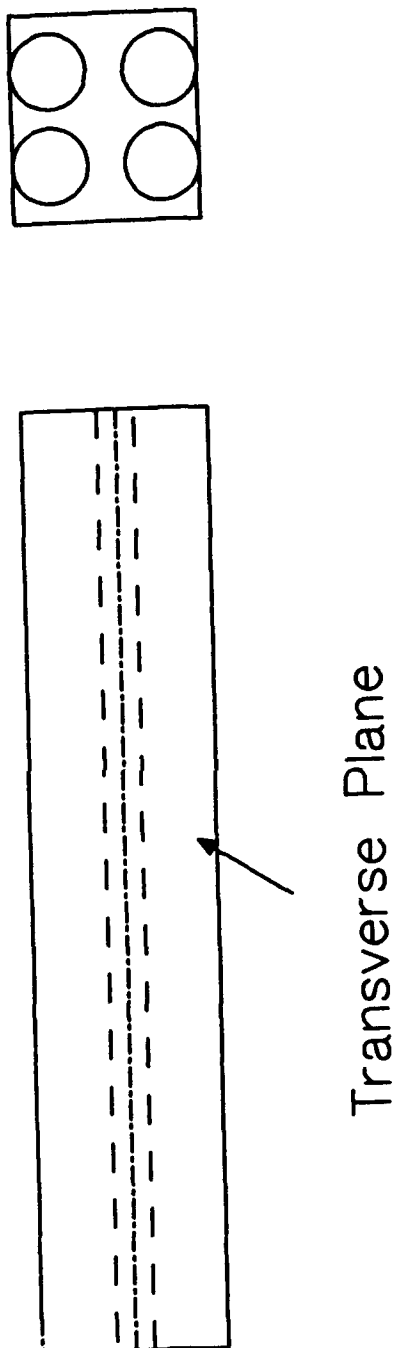


Figure 2.3 Samples used for Ac_1 and Ac_3 determinations. Note that the center-line segregation was avoided.

2.3 Prior Austenite Grain Size

2.3.1 Introduction

Grain size is one of the most important factors in determining the strength and toughness of steel products. One of the goals of this research program is to compare as closely as possible the transformation behavior of a low-carbon modified A710 steel (20) with the present alloys. Colvin (20) used 905°C as the austenitizing temperature for his study, and determined that an austenite grain size of about 10 to 15 μm was generated after 5 minutes holding at this temperature. He also found that longer holding times do not have a noticeable effect on the grain size.

2.3.2 Heat treatment

Small coupons (19x19x3 mm) were cut from each steel using a band saw (Figure 2.4). Various heat treatments were conducted in salt baths or muffle furnaces, and because the results of the muffle furnace method were reproducible, this method was chosen for the rest of study.

At the first stage of these heat treatments, six samples from each steel were used. Samples were soaked in

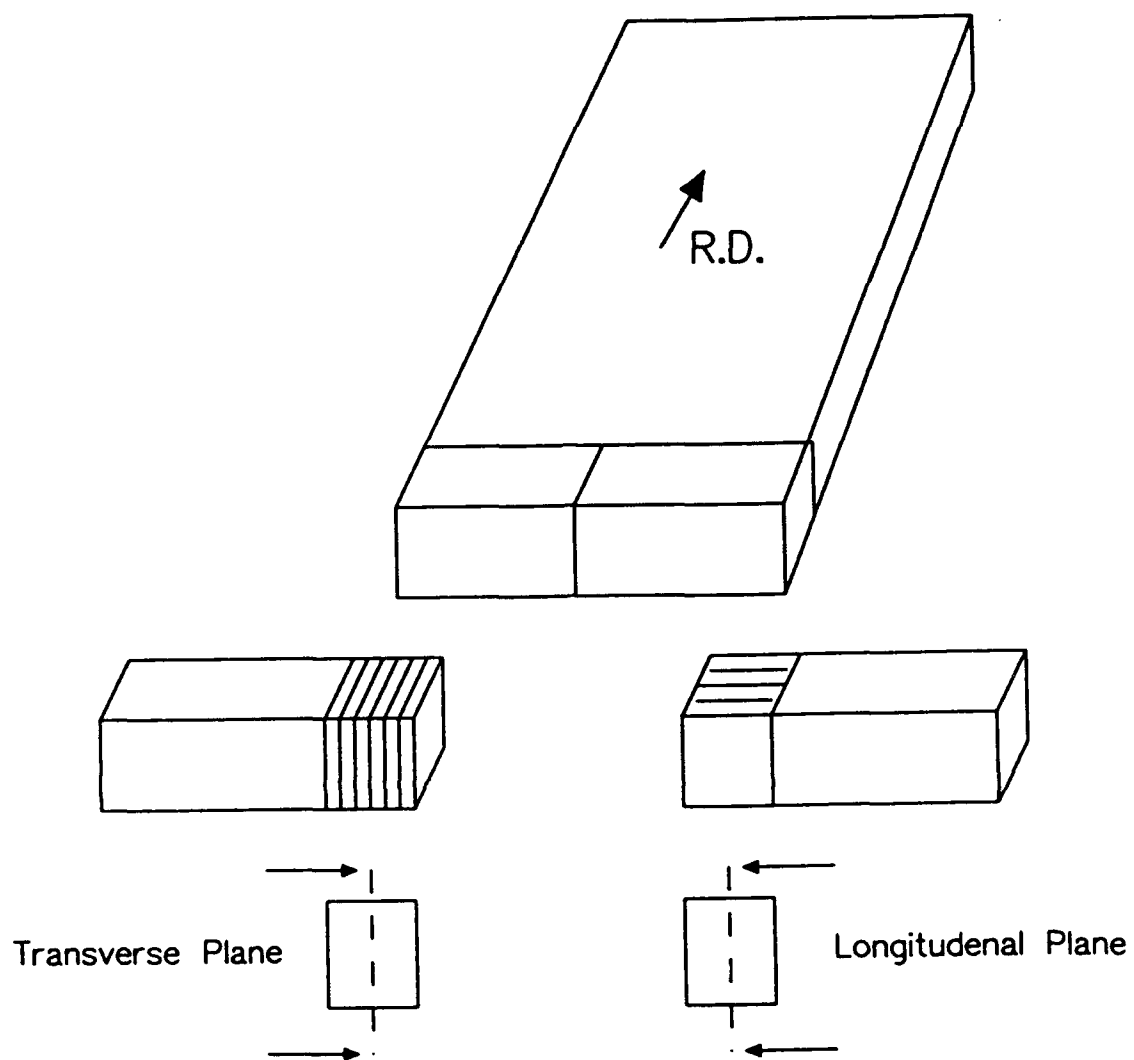


Figure 2.4 Wafers used for grain size studies.

the salt pot at 905-910°C for ¼, 2, 5, 15, 30 and 60 minutes and then water-quenched. These samples were designated A1-A6, B1-B6 and C1-C6.

The results of these treatments showed that a holding time of 905°C does not have a noticeable effect on the grain size. A five minute holding time was chosen. Since the goal of a 10-15 µm grain size was not achieved, new heat treatment schedules were designed. These treatments included reaustenitizing at 905°C, followed by water quenching and/or air cooling. Another important factor was the composition of the steels. For the sake of comparison with other research programs at CSM, steel B (which contains 1.5% Mn) was chosen for the rest of study. These heat treatments are shown in Figure 2.5 and the results are presented in Chapter 3.

Specimen preparation for optical microscopy was done as described before. Chemical etching was very difficult. Eventually several etching and staining solutions were used. The best way found to reveal the prior austenite grain boundaries for these steels was a combination of the following etchants and back polishing technique.

1. 50 ml of Nital 2%, two drops of each H_2SO_4 and H_2O_2 .
2. Saturated picric acid and wetting agent.

The first etchant was developed experimentally and works the

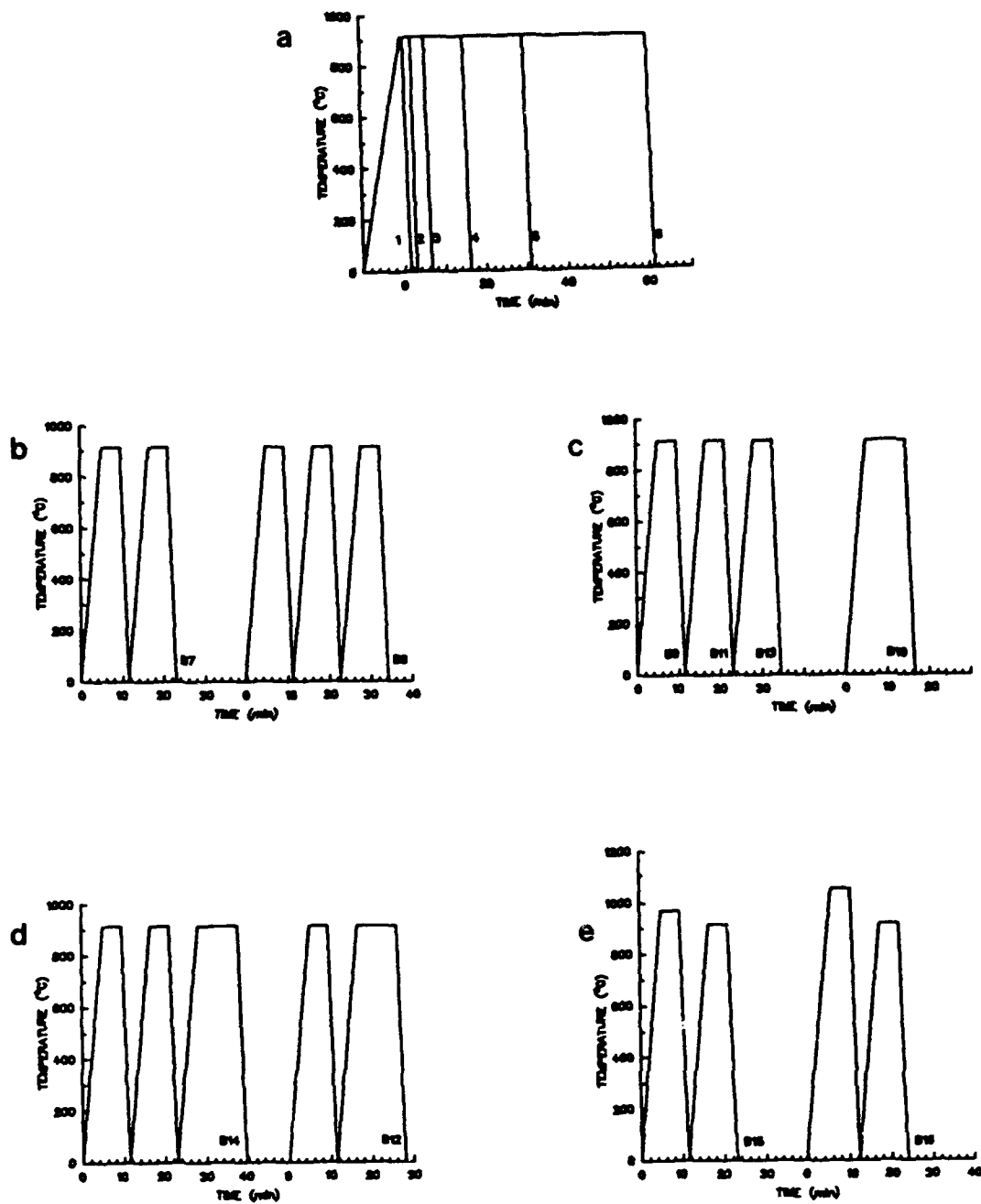


Figure 2.5 Different heat treatment schedules for grain size study.

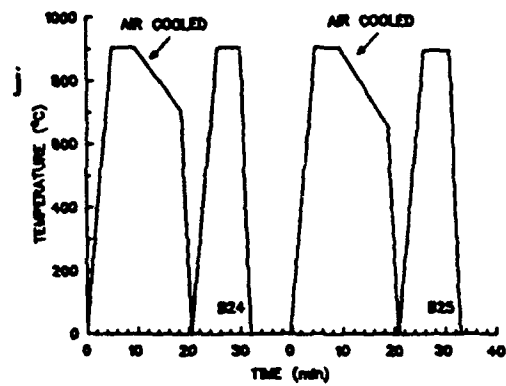
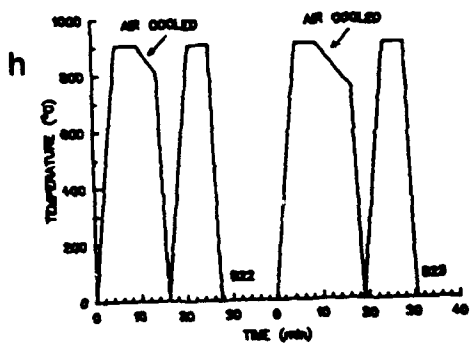
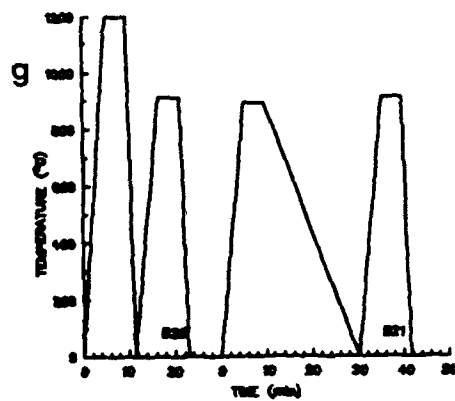
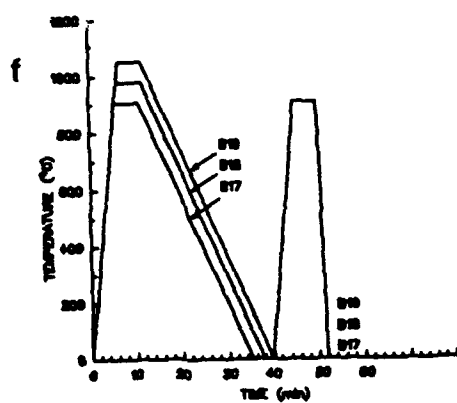


Figure 2.5 (continued)

best in conjunction with back polishing. Tempering at 450°C for 24 hours is sometimes helpful to reveal the prior austenite grain size. The second etchant has the best effect on small-grain-size samples. The average grain size was determined by standard linear intercept method techniques (112-115).

2.4 Transformation Kinetics

2.4.1 Gleeble 1500

The Gleeble 1500 is a computer controlled thermomechanical test system which has been designed for both thermal and mechanical testing simultaneously (116). The thermal capability of the Gleeble 1500 was used exclusively for the current study.

Since electric resistance heating produces a more uniform temperature zone than induction heating (Figure 2.6), the Gleeble 1500 has been designed to take advantage of electric resistance heating, although it is possible to use induction heating as well. In the case of induction heating, the magnetic field generated by a high frequency current passing through a coil produces heat inside the material at the core of the coil. By contrast, electric resistance heating consists of passing an electric current

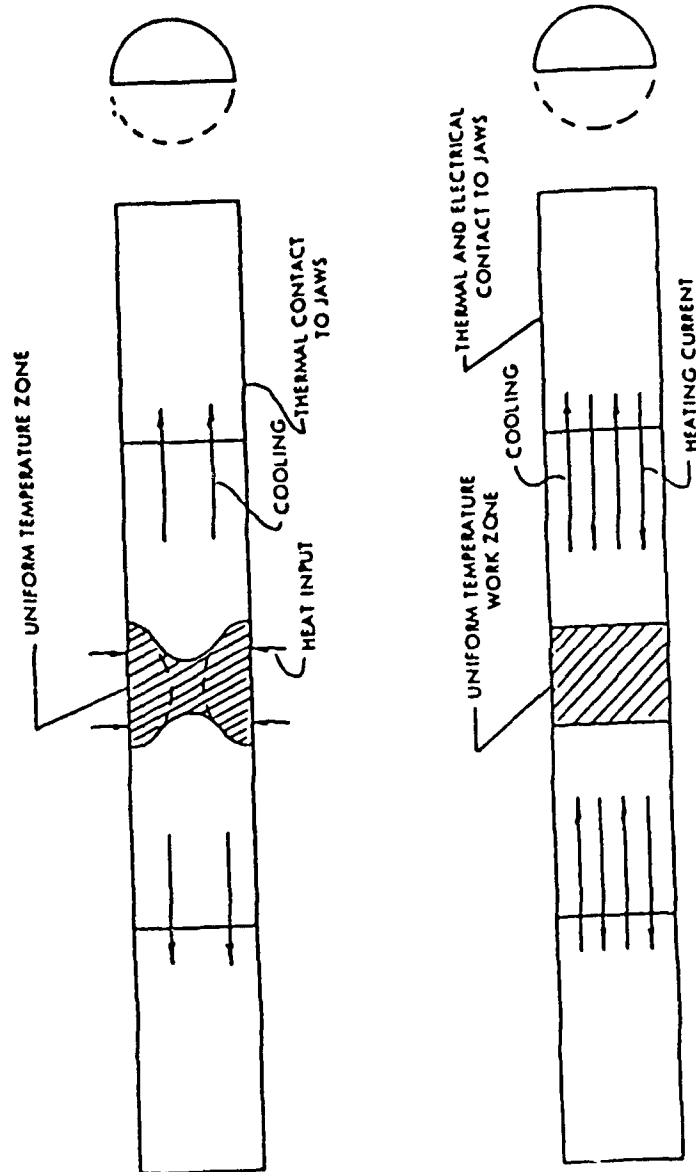


Figure 2.6 Schematic diagrams showing the uniformity of temperature zone in: induction heated/conduction cooled (top), and electric resistance heated/conduction cooled (bottom) specimens (116).

through a metallic specimen, where the resistance of the material to the flow of current generates heat.

Distribution of this current in the test sample is controlled by the cross section of the specimen and the water-cooled copper jaws which are used to mount the specimen. Although water-cooled jaws cause the heat to flow through the length of the specimen, Figure 2.7 (116) shows that radial conduction cooling is practically zero and causes no temperature gradient through the cross section of the sample. This means a uniform temperature along the cross section of the work zone, which is illustrated in Figure 2.8, (116). Considering electric resistance heating and conduction cooling, the flat thermal profile can be maintained even if the heating rate exceeds 20,000°C/sec (116).

Because the measurements of temperature and dilation are taken across an isothermal plane, alumel and chromel wires must be welded to the center of the specimen's midspan very carefully. When electric current flows through these thermocouple wires, the instant temperature of the midspan will be transferred to the computer. For working temperatures up to 1200°C, alumel and chromel make an ideal thermocouple for most of materials.

A C-strain (circumferential strain) fixture

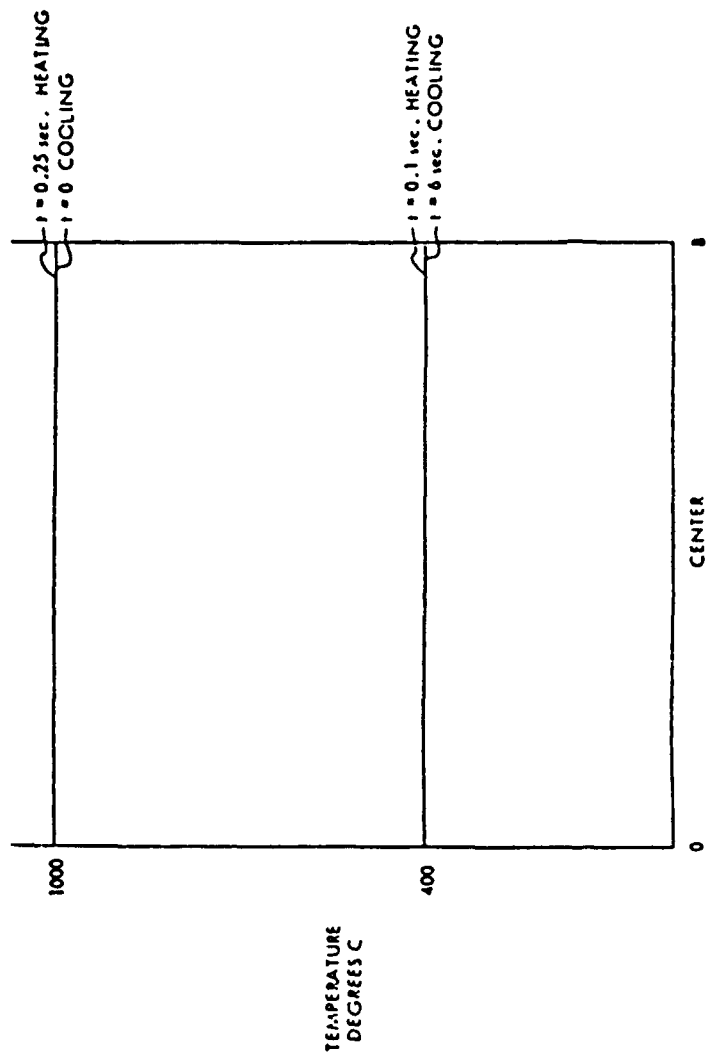


Figure 2.7 Radial temperature gradient for an austenitic stainless steel subjected to electric resistance heating and axial conduction cooling. Heating rate: 4000°C/s , cooling rate: 100°C/s (116).

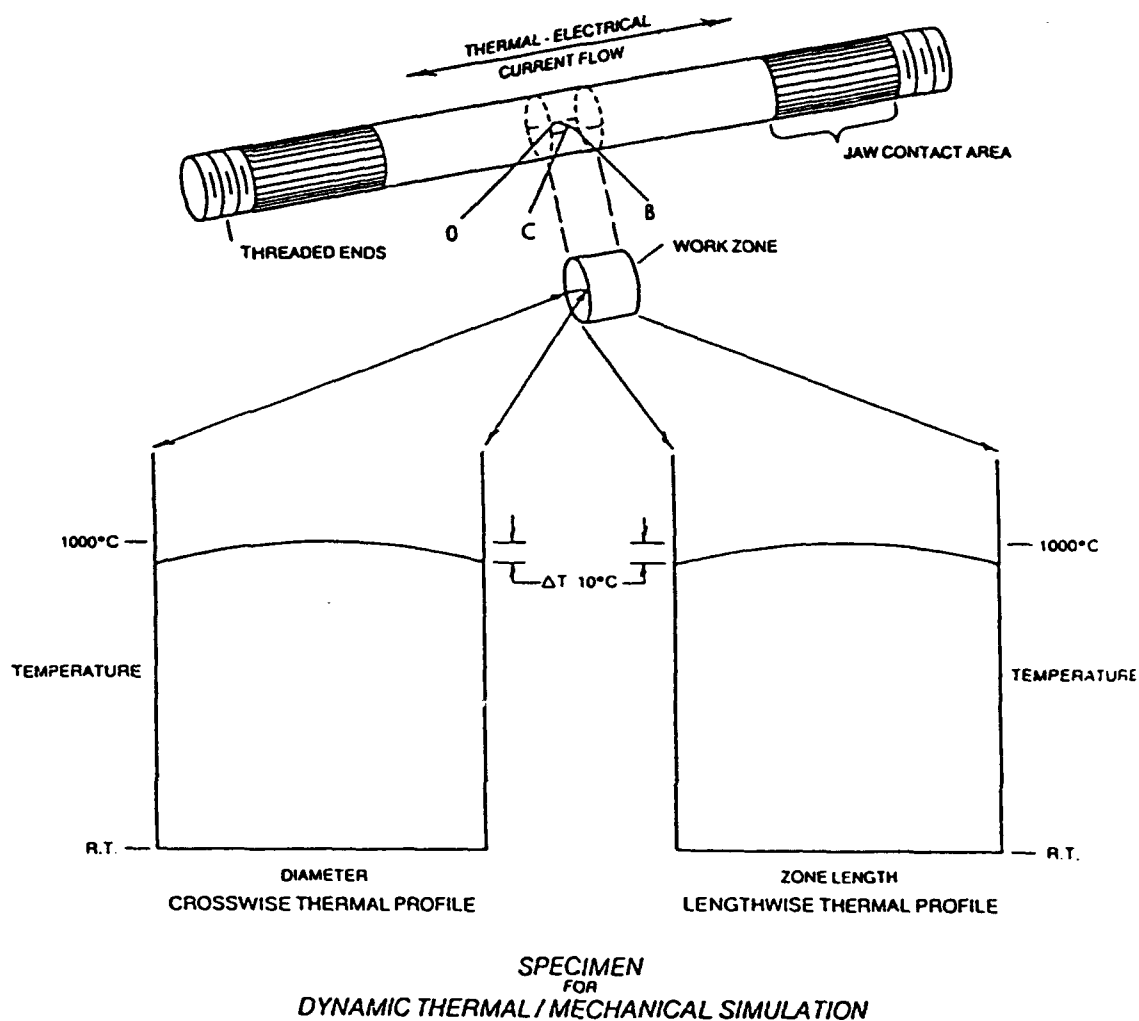


Figure 2.8 Schematic diagram showing the electric resistance heating of a sample used in the Gleeble. Uniform heating is produced across the work zone length and diameter (116).

(dilatometer) measures the strain in the isothermal plane. The resolution of the transducer used for these measurements is essentially infinite (116). The method of mounting of the C-strain fixture is very important. To get the best results, the dilatometer should be mounted as tightly as possible, to guarantee measurement of very small displacements.

The atmosphere of the test can be changed, depending on the desire of the researcher. Argon and helium gas are the most common types of atmosphere used for CCT tests.

A data acquisition system and ASYST data manipulation software have been installed to the system available at the Colorado School of Mines.

2.4.2 Sample Geometry

The geometries of the samples used in this study were determined based on previous and current work done at the Colorado School of Mines (20,38). The most important factors to consider when choosing a certain geometry are axial cooling capacity of the samples, without using any further aid such as gas cooling and so on, as well as maintaining an isothermal plane during heating and cooling.

Figure 2.9 shows the four different geometries that

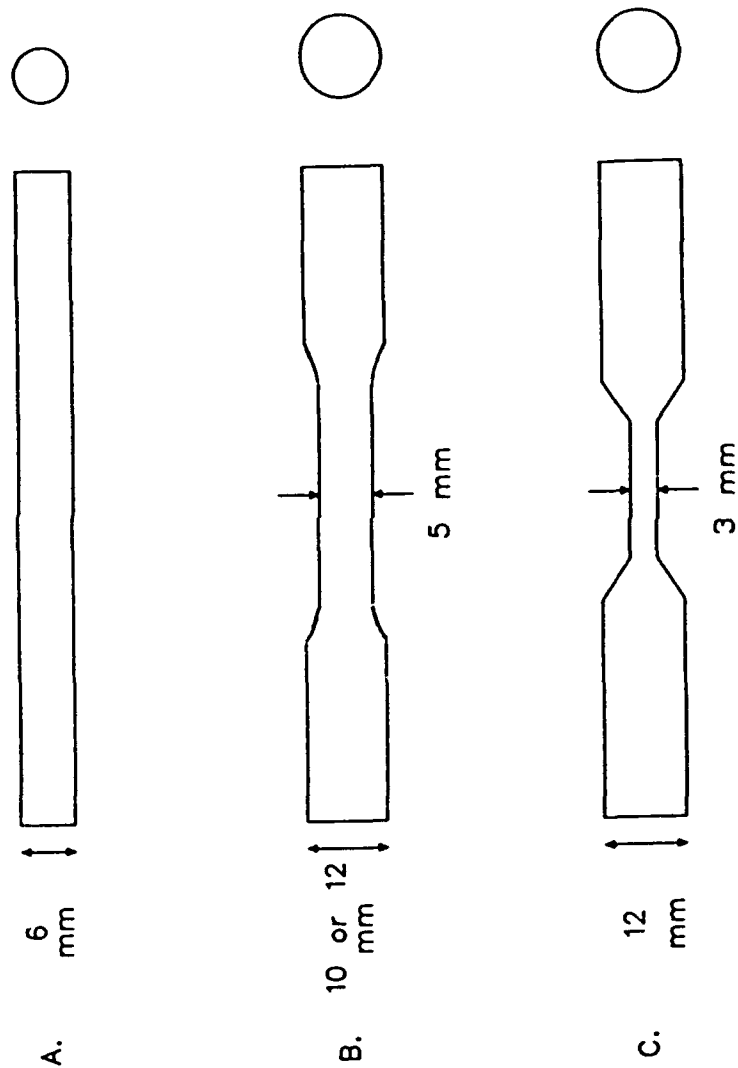


Figure 2.9 Geometry of Gleeble samples. (a) Cooling rates $\leq 10^{\circ}\text{C/s}$.
(b) (10 mm) $10^{\circ}\text{C/s} \leq \text{C.R.} \leq 25^{\circ}\text{C/s}$, (12 mm) $25^{\circ}\text{C/s} \leq \text{C.R.} \leq 75^{\circ}\text{C/s}$.
(c) $\text{C.R.} \geq 75^{\circ}\text{C/s}$.

were used in this study. Type A was used for slow cooling rates (below $10^{\circ}\text{C}/\text{sec}$), while type B (10 mm) was observed to be more suitable for intermediate cooling rates between $10^{\circ}\text{C}/\text{sec}$ to $25^{\circ}\text{C}/\text{sec}$. For cooling rates between $25^{\circ}\text{C}/\text{sec}$ up to $75^{\circ}\text{C}/\text{sec}$ type B (12 mm) was used, and type C was used for fast cooling rates above $75^{\circ}\text{C}/\text{sec}$. In this way, good agreement between the actual temperature and the program temperature was obtained (38).

To avoid the center-line segregation bands, machining of the above samples was designed to be off-centered. Figure 2.9 shows more details about the dimensions and design of these specimens.

2.4.3 Dilatometry as a Function of Heating and Cooling Rate

It is well known that the transformation temperature in both heating and cooling is affected by the rate of input or output of heat (40,116,117). It has been suggested (117) that heating rates of 2 to $3^{\circ}\text{C}/\text{sec}$ do not change the transformation temperatures. Considering the above facts, $3^{\circ}\text{C}/\text{sec}$ was chosen as the heating rate throughout the experiments. Since the exponential cooling is more close to reality than linear cooling, an exponential cooling program

(20), similar to those calculated with the Newtonian formula (118), was used for this research. The program considers the austenitizing temperature, upper and lower boundary temperatures and time between these temperatures to simulate the desired exponential cooling rates. Because most of the austenite transformations in hypoeutectoid steels occur between 800°C and 500°C (111,118), the time interval between these temperatures (Δt_{8-5}) is a base for calculation of the cooling rate. These temperatures were chosen as upper and lower boundaries for cooling programs.

2.4.4 Dilatometry

Determining transformation temperature has traditionally been done through measurements of the magnetic properties of a material, or by measuring volume change taking place during the transformation (111). The latter method is preferred by many investigators (111), and was used in the present work.

In the process, radial contraction of the midspan for each specimen as a function of the actual temperature was measured. Dilatometric measurements were conducted using a dilatometer that has been designed by Duffers Inc. for the Gleeble 1500. It is also important to note that proper

insulation or water cooling should be used to protect the transducer at working temperatures higher than 900°C.

2.4.5 Testing Procedure

Thermocouple wires and a C-strain fixture are the two most important accessories in these tests. To guarantee proper attachment of the above accessories, all the test specimens were polished with a 600 grit silicon carbide paper and then cleaned with ethanol. To maintain good electric contact, copper grips were checked and cleaned. After putting a ceramic insulator on thermocouple wires prior to each test, they were attached to the proper switch. The dilatometer then was placed on the center of the midspan opposite the thermocouple. The test chamber was evacuated after a chamber cover was put in place. This process was continued by purging argon gas into the chamber. After filling the chamber with the argon gas and maintaining a positive pressure, evacuation took place one more time, followed by a second argon gas purging. When the chamber pressure reached a level slightly above the atmosphere pressure, a safety valve was opened to maintain constant pressure inside the chamber. The above procedures prevent samples from being oxidized or decarburized.

Samples were heated at a rate of 3°C/sec to an austenitizing temperature of 905°C for steels B and C, and 920°C for steel A. Each sample was held at the proper temperature for 5 minutes prior to cooling. Basically, two CCT diagrams were generated for steel B based on prior austenite grain size.

A single cycle (1X) of austenitizing corresponded with a prior austenite grain size of about 30 μm ; 2X meant a double-cycle austenitizing treatment and corresponded with about a 7 μm prior austenite grain size. Cooling rates varied between 150°C/sec and 0.05°C/sec.

The dilatometric data for each test of steel B were plotted versus actual cooling temperature. Therefore, for each cooling rate, there are at least two sets of data based on two starting grain sizes. Five cooling rates were chosen to examine the relative position of CCT diagrams for steels A and C with respect to steel B.

Transformation start and finish temperatures were determined by drawing tangents to C-strain versus temperature curves. An excellent guide to this practice is given by Eldis (111). The times and temperatures which correspond to 25%, 50% and 75% transformation of austenite were also measured graphically using the dilatometric graphs (Figure 2.10).

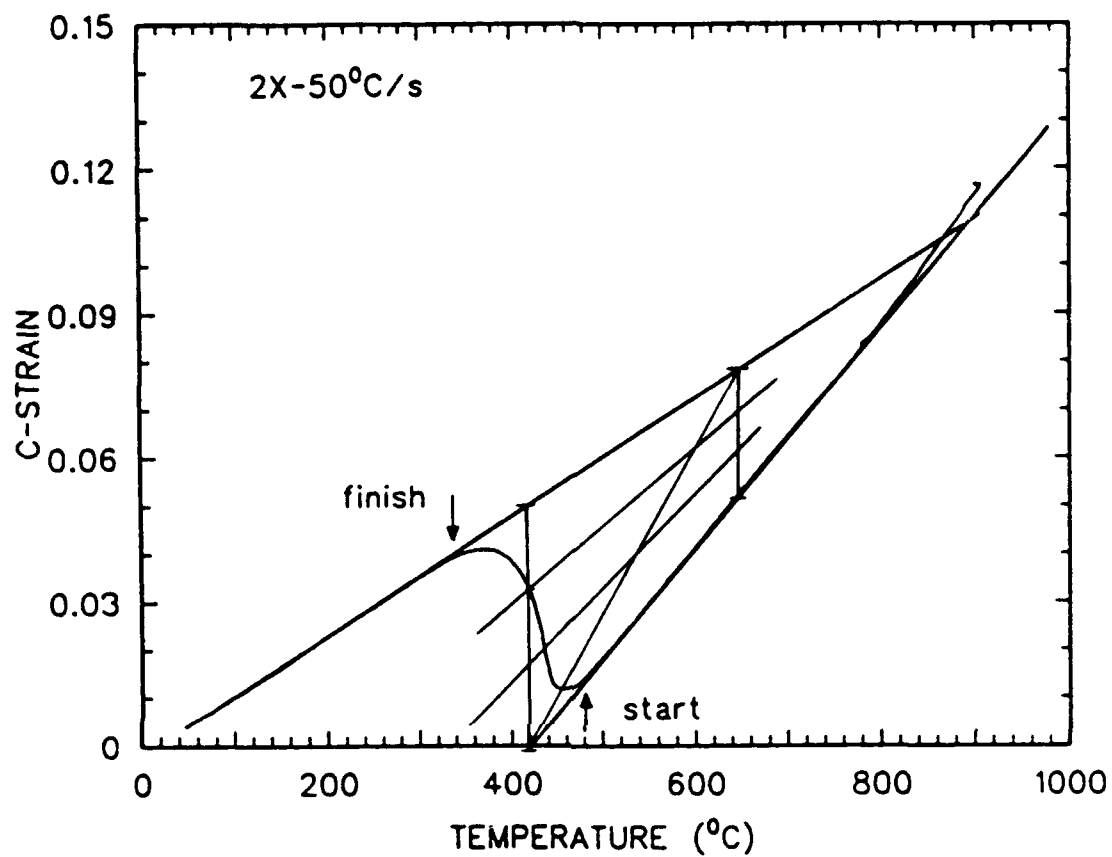


Figure 2.10 Graphical measurements of the 25, 50 and 75 percent transformation temperatures.

According to Eldis's studies (111), the uncertainties in the transformation start and finish temperatures recorded for any cooling curve are as much as $\pm 5^{\circ}\text{C}$ and $\pm 35^{\circ}\text{C}$, respectively. The high uncertainty number for the finish temperature has been observed when there is a marked curvature in the C-strain versus temperature data.

Following each test, a diamond saw was used to cut each sample from the middle of the dilatometry sample (as close to the isothermal plane as possible.) After mounting and polishing, the small samples were examined microstructurally, and microhardness was determined.

2.5 Microhardness Measurements

Diamond pyramid hardness (DPH) measurements were conducted according to ASTM specification E92-82 (119) on a Leco microhardness tester (model M-400A). A 500-gram load was used throughout the tests. The average of at least ten measurements is given as the overall hardness of each specimen. These values (Vickers hardness numbers (VHN)) were plotted versus $\log (\Delta t_{8-5})$ to produce a hardness profile.

2.6 Microstructural Characterization using SEM and TEM

In addition to optical microscopy, scanning electron microscopy (SEM) was performed with a JEOL JXA 840 electron probe micro-analyzer. Samples were coated with gold to 300 μm thickness to reduce the charging effect.

Transmission electron microscopy (TEM) was performed using a PHILIPS EM400. Three specimens for examination by TEM were sliced from the middle of each Gleeble sample to a thickness of 0.20 to 0.40 mm using a Buehler Isomet saw, as shown in Figure 2.11.

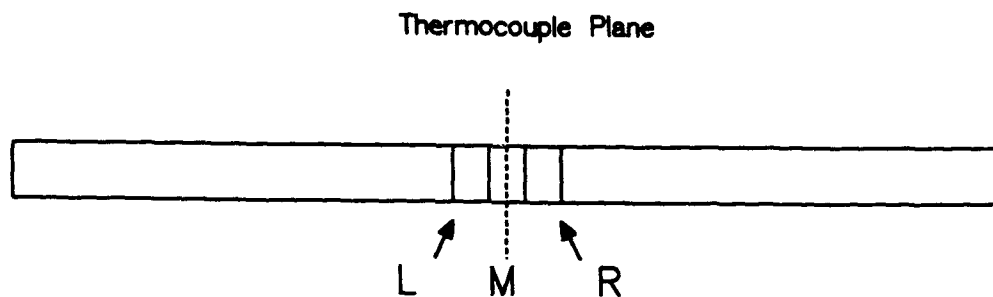


Figure 2.11 Positions of the wafers cut for TEM studies relative to the position of the thermocouple (L: left, M: middle, R: right).

Some wafers from the as-received steel B were also cut to be examined by TEM. The above specimens were then mechanically thinned on 600 grit silicon carbide paper to about 0.2 mm. Three-mm discs were punched except for the samples with 3-mm reduced section. These discs were thinned

mechanically by the same technique as before to approximately 0.06 mm. Care was taken during the mechanical cutting and thinning processes in order not to induce heat or deformation into the specimens.

A Fischione twin-jet electropolisher was used for further thinning of the foils. Two different solutions were used in order to get the best results. The first solution consisted of 95 volume percent acetic acid and 5 volume percent perchloric acid. The discs were thinned to perforation at room temperature with 80 volts and 80 milliamperes. Microscopy was performed at an accelerating voltage of 120 kV. Bright field (BF) and centered dark-field (CDF) imaging techniques were employed in addition to the generation and analysis of selected-area diffraction patterns (SADPs). The quality of foils was fair.

A second solution was used which consisted of 70 volume percent ethanol, 25 volume percent glycerol and 5 volume percent perchloric acid. This solution improved the quality of the foils slightly.

2.7 Tensile Testing

Six blanks from each steel were cut with a band saw and heat treated prior to machining (Figure 2.12). Heat

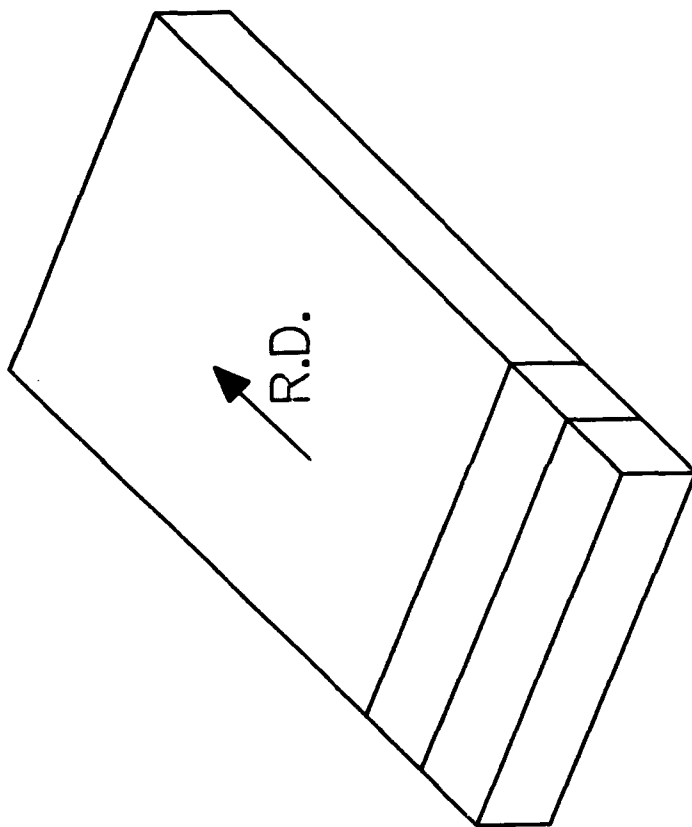


Figure 2.12 Orientation of tensile bars and Charpy bars relative to the rolling direction.

treating was conducted in a muffle furnace at 905°C for steels B and C, and at 920°C for steel A. The austenitizing time was five minutes for each sample. Three samples from each steel were air cooled after five minutes of heat treating, and they are designated as 1X. The remaining three samples from each steel underwent two rounds of austenitizing, of which the first round was followed by water quenching and the second round by air cooling. These samples were designated as 2X. To maintain a 1°C/sec cooling rate, each sample was placed between two plates to simulate a total thickness of one inch.

In addition to the above samples, six extra blanks from steel A were cut and underwent the same heat treatments as well as 35 minutes tempering at 620°C for each sample. These samples were designated AT-1X and AT-2X. Off-centered machining was conducted on all of the specimens to avoid the segregation line. Standard, round-bar tensile specimens were machined to ASTM specification E8-82 (120). Tests were performed on an Instron screw machine located at the Colorado School of Mines. Cross-head velocity was set at 0.02 mm/sec (0.05 in/min). A 25.4 mm (1 inch) extensometer was used to measure displacement. Load-displacement data were plotted directly on a strip-chart recorder. This chart was used to determine the 0.2% offset

yield point and ultimate tensile strength (120,121). In addition to the strip chart recorder, load-displacement-time data were collected on a Compaq 386-SX PC which is equipped with a Labtech Notebook software.

2.8 Charpy Impact Testing

Charpy V-notch specimens were machined according to ASTM specification E23-82 (122) after being heat treated. Blanks were cut from each steel as shown in Figure 2.12. These samples were then heat treated as explained for tensile specimens and were designated 1X and 2X, depending on the number of reaustenitizing treatments. Extra specimens from steel A were tempered for 35 minutes at 620°C after the aforementioned heat treatments and designated as AT-1X and AT-2X.

The heat-treated specimens were then machined in transverse orientation, with the notch perpendicular to the rolling plane (the T-L orientation). In this way, the crack will propagate in the rolling direction during testing.

For Charpy V-notch testing, temperatures higher than room temperature were achieved in a hot oil bath, while temperatures below 0°C were obtained with a mixture of ethanol and liquid nitrogen. Ice water was used for tests

at 0°C. Specimens were held at the test temperatures for a minimum of seven minutes and were broken within three seconds after being taken from the bath. The absorbed energy on impact was directly read from the Charpy impact machine dial. Samples were then preserved for further study by spraying lubricant on the broken surfaces. To determine the ductile-to-brittle transition, CVN energy data were plotted versus test temperatures.

2.9 Fractography

Fractography was performed on six samples from steel B. Three samples from 1X condition and three samples from 2X condition were each chosen from the upper shelf (100°C test temperature), transition region (room temperature) and lower shelf (liquid nitrogen temperature). A JEOL JXA-840 scanning electron microscope (SEM) was used. Pictures were taken at three different magnifications of 500X, 1000X and 2000X from each sample.

3.0 RESULTS AND DISCUSSION

3.1 As-Received Plates

The chemical compositions, general microstructures, inclusion contents and hardness values of all the as-received steels are presented in this section. Chemistries of the three as-received steels are reported in Table 3.1. For each steel, no variation in chemical compositions was observed (considering the size of the probe) through the width of the plate.

All of the three steels exhibited very large, unrecrystallized prior austenite grains at the plate edges. These grains were limited to regions of about two centimeters from the edges, and these regions were avoided in further studies. Segregation in the form of wide, white (difficult-to-etch regions) bands was present, to some extent, at the centerline of all steels. The effect was most pronounced in steel C, which has the highest manganese content. This phenomenon was probably due to manganese and/or molybdenum and nickel segregation. Off-centered cuttings were designed, when it was necessary, to avoid the segregation bands.

Light microscopy and SEM were used to determine the

Table 3.1 Chemical Compositions of Steels.

STEEL/ANALYSIS	C	Mn	Mo	Ni	Nb	Cr	Si	S	P	Al	Ti	N	O
A/USS-L	.022	1.00	3.51	3.40	.061	.230	.200	.003	.001	.015	.010	.010	.006
*A/USS-P	.025	1.00	3.67	3.57	.054	.230	.200	.003	.005	.013	.010	.010	NA
A/CSM-P	.012	1.08	3.47	3.34	.062	.273	.277	.003	.007	.003	.009	NA	NA
B/USS-L	.023	1.47	3.46	3.35	.061	.230	.220	.003	.002	.011	.006	.010	.006
*B/USS-P	.024	1.45	3.62	3.55	.055	.230	.180	.003	.005	.009	.009	.010	NA
B/CSM-L	.010	1.55	3.34	3.18	.063	.270	.329	.003	.008	.003	.007	NA	NA
C/USS-L	.023	2.00	3.44	3.34	.060	.230	.200	.004	.001	.010	.006	.009	.006
*C/USS-P	.020	1.76	3.58	3.47	.054	.230	.210	.003	.005	.009	.008	.012	NA
C/CSM-P	.012	1.97	3.16	2.98	.065	.277	.286	.004	.009	.004	.006	NA	NA

USS: U.S. Steel Technical Center

CSM: Colorado School of Mines

L: Ladle Analysis

P: Product Analysis

*: This analysis is considered to be the most accurate analysis.

morphology, distribution, relative number density and chemistry of numerous inclusions in all three steels. In general, all three steels contained three types of inclusions: rectangular or square inclusions (which will be referred to as type I), round or oval inclusions (type II) and irregularly shaped, angular inclusions (type III). All three steels exhibited large numbers of types I and II inclusions. The relative numbers of type III inclusions increased dramatically in going from steel A to steel B to steel C.

Energy-dispersive-spectroscopy (EDS) analysis in the SEM showed that the type I inclusions contained high titanium levels. These could be titanium nitrides formed during solidification. Type II inclusions showed high manganese and sulfur levels, with some aluminum or silicon. These inclusions are manganese sulfides, with the aluminum probably present in the form of a small region of Al_2O_3 , and silicon as SiO_2 . As shown in Figure 2.2, these observations were made from transverse sections, and that explains why these inclusions are not described as stringers (the elongation of MnS is obscured in this section). The type III inclusions were apparently clusters of type I and type II inclusions. Figure 3.1 shows a light micrograph and two SEM micrographs, which are representative of the observed



Figure 3.1 (a) Light micrograph showing MnS inclusions.
(b) and (c) SEM micrographs showing Ti- and Mn-rich inclusion clusters.

inclusions. The EDS analysis of manganese- and titanium-rich inclusions are shown in Figure 3.2.

Table 3.2 shows the as-received hardness values of the three steels. The hardness increases by two HRC for each 0.5 wt. pct. increase in manganese content. This trend could be

Table 3.2 Hardness of the as-received steels.

STEEL	HRC
A	30
B	32
C	34

explained by the role of manganese as an austenite stabilizer. By increasing the amount of manganese in steel, the austenite to ferrite transformation temperature as well as the M_s temperature will decrease, which, in turn, promotes smaller ferrite crystals, finer packet sizes, more martensite and higher dislocation density. One or more of these features are responsible for increasing strength and hardness.

Figure 3.3 shows the microstructures of the as-received steels A, B and C. At this magnification, the bainite-like



Figure 3.2 Representative of EDS spectra for inclusions in steels A, B and C.

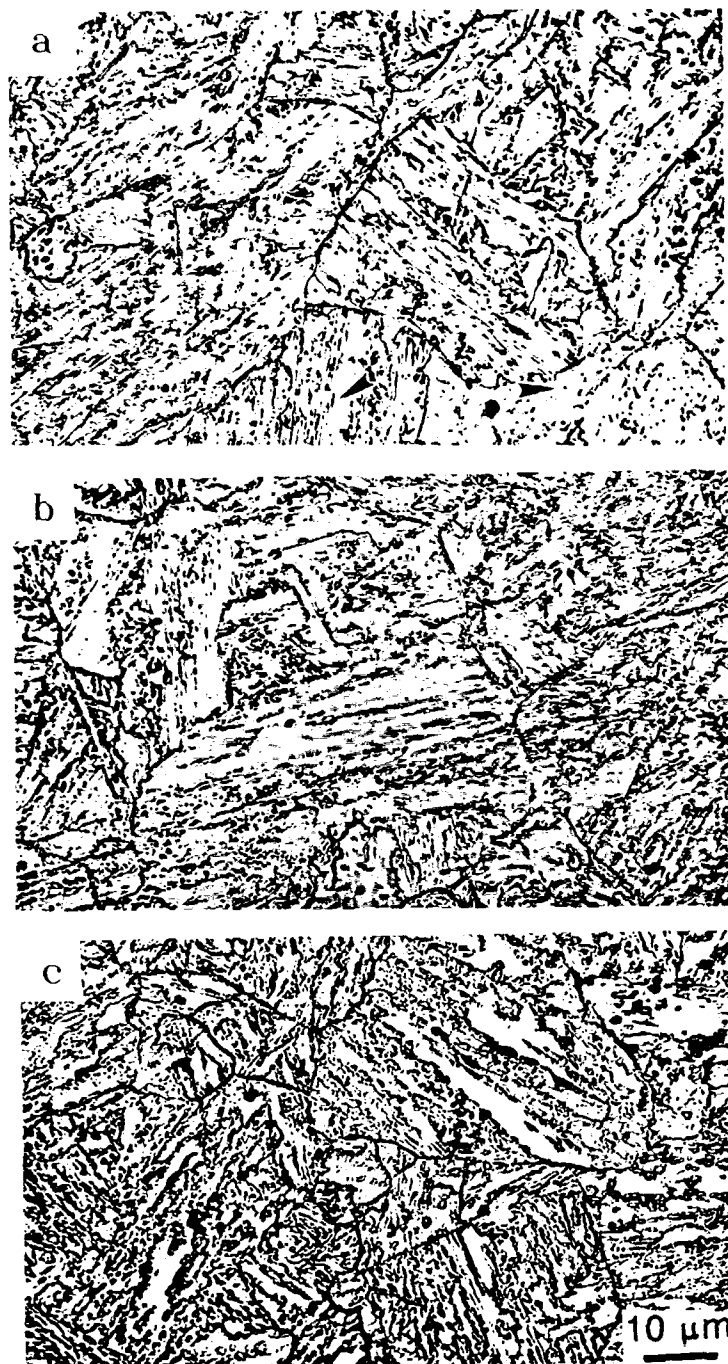


Figure 3.3 The as-received microstructures of (a) steel A, (b) steel B and (c) steel C. Light micrographs.

microstructures of all of these steels are clear.

Bainite consists of bainitic ferrite laths with interlath second-phase particles. The dark small particles are probably retained austenite or martensite. Because of the very low carbon content of these steels, these particles could not be considered as interlath carbides, since the volume fraction of these particles is too high to be carbides. This will be shown later in TEM micrographs. The high alloy content of these steels could be responsible for the presence of retained austenite.

3.2 Ac_1 and Ac_3 determinations

Empirical formulas were used to calculate the transformations from ferrite to austenite, upon heating. According to Andrews (123) the Ac_1 is given in ($^{\circ}C$) as:

$$Ac_1 = 723 - 10.7 Mn - 16.9 Ni + 29.1 Si + 16.9 Cr + 290 \\ As + 6.38 W \quad (3.1)$$

and the Ac_3 is given by:

$$Ac_3 = 910 - 203\sqrt{C} - 15.2Ni + 44.7Si - 104V + 31.5 Mo \\ + 13.1W - 30Mn - 11Cr - 20Cu + 700P + 400Al \\ + 120 As + 400 Ti \quad (3.2)$$

In both of the above formulas, all elements are in wt. pct. These equations have been determined for certain ranges of weight percent for each alloying element (123). Although most elements were within the given ranges by Andrews, there were two exceptions. The lowest carbon content used for the above formulas was 0.08% and the highest molybdenum content was 2.6%. Calculated values based on these equations as well as experimental (measured) values are presented in Table 3.3. The experimental values are based on dilatometric inflections in the C-strain versus temperature curves obtained for specimens heated up to 1200°C at the rate of 3°C per second.

Table 3.3 Calculated and measured austenite formation temperatures for steels A, B and C.

Steel	Calculated		Measured	
	Ac ₃ (°C)	Ac ₁ (°C)	Ac ₃ (°C)	Ac ₁ (°C)
A	926	664	901	794
B	909	660	886	766
C	890	655	882	755

Figure 3.4 shows a plot of C-strain (circumferential strain) versus temperature as obtained in the Gleeble for steel B.

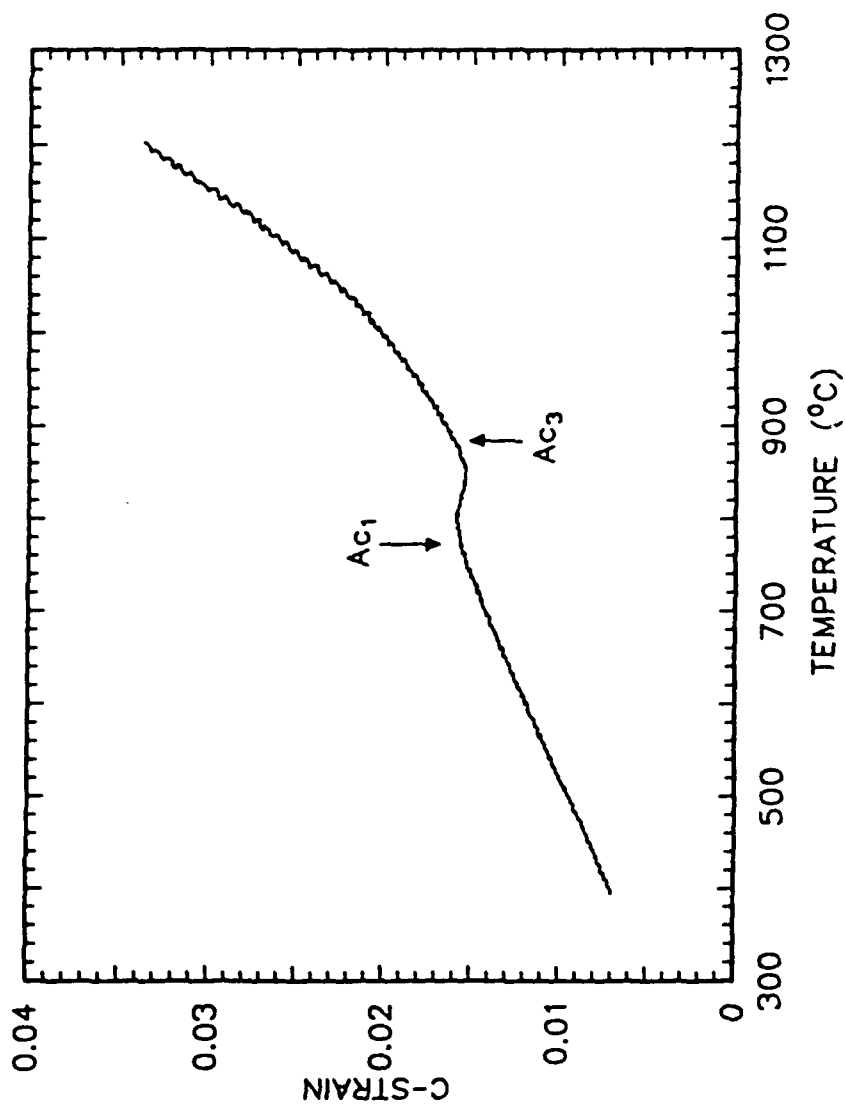


Figure 3.4 Dilatation curve for steel B. The specimen was heated up to 1200°C at a rate of 3°C/s.

The arrowed inflection points clearly define the start of austenite formation at low temperatures and the completion of austenite formation at high temperatures.

A comparison between calculated and measured critical transformation temperatures for all of the three steels shows inconsistency, especially for the lower critical temperatures. These are due to the lack of molybdenum term in the formula used to calculate the Ac_1 temperature, and the slight difference in the range of compositions for carbon and molybdenum. For comparison, Colvin (20) reported Ac_1 and Ac_3 temperatures of 690°C and 860°C, respectively, for a modified A710 steel with 1.45% manganese.

3.3 Prior Austenite Grain Size

A grain size study was initiated to determine what range of austenite grain sizes could be developed in the present set of alloys. The aim was to produce about 10 to 15 μm grain size, similar to Colvin's work. An austenitizing temperature of 905°C was chosen (except for steel A which was 920°C). The first component of the grain size study showed that grain size did not change much with time, up to a one-hour austenitizing time. Hence five minutes holding time was chosen for the rest of study. The

results of these studies are presented in Table 3.4. The first stage of the study showed that grain sizes are mostly about 30 μm , except for steel A after a 60 minutes holding period. This specimen was accompanied by an abnormal grain coarsening phenomenon. This stage also showed that these grain sizes are about two to three times larger than desired.

A great number of heat treatments were examined in an attempt to create a 10 to 15 μm grain size. Despite many processing variations, two extremes were typically observed. Specimens which were reaustenitized once and quenched, usually had a grain size close to about 30 μm . Specimens which were reaustenitized twice and quenched after each treatment had grain sizes of about 7 μm and less. Figure 3.5 demonstrates some representative micrographs of these studies. Because of the above results, it was decided to examine both fine-grained and coarse-grained material throughout the rest of study.

3.4 Transformation Kinetics

Although most of this study was based on steel B with 1.5% manganese, it was decided later to choose four cooling rates to examine the relative positions of the CCT diagrams

Table 3.4 Austenite grain sizes for various austenitizing times.

Steel	Heat Treating Time (min)	Grain Diameter (μm)
A	0.5	27.3
A	2	29.7
A	5	29.2
A	15	33.7
A	30	37.8
A	60	67.5
B	0.5	33.7
B	2	33.7
B	5	34.8
B	15	33.7
B	30	33.7
B	60	38.2
C	0.5	30.3
C	2	31.4
C	5	33.7
C	15	34.8
C	30	33.7
C	60	38.2

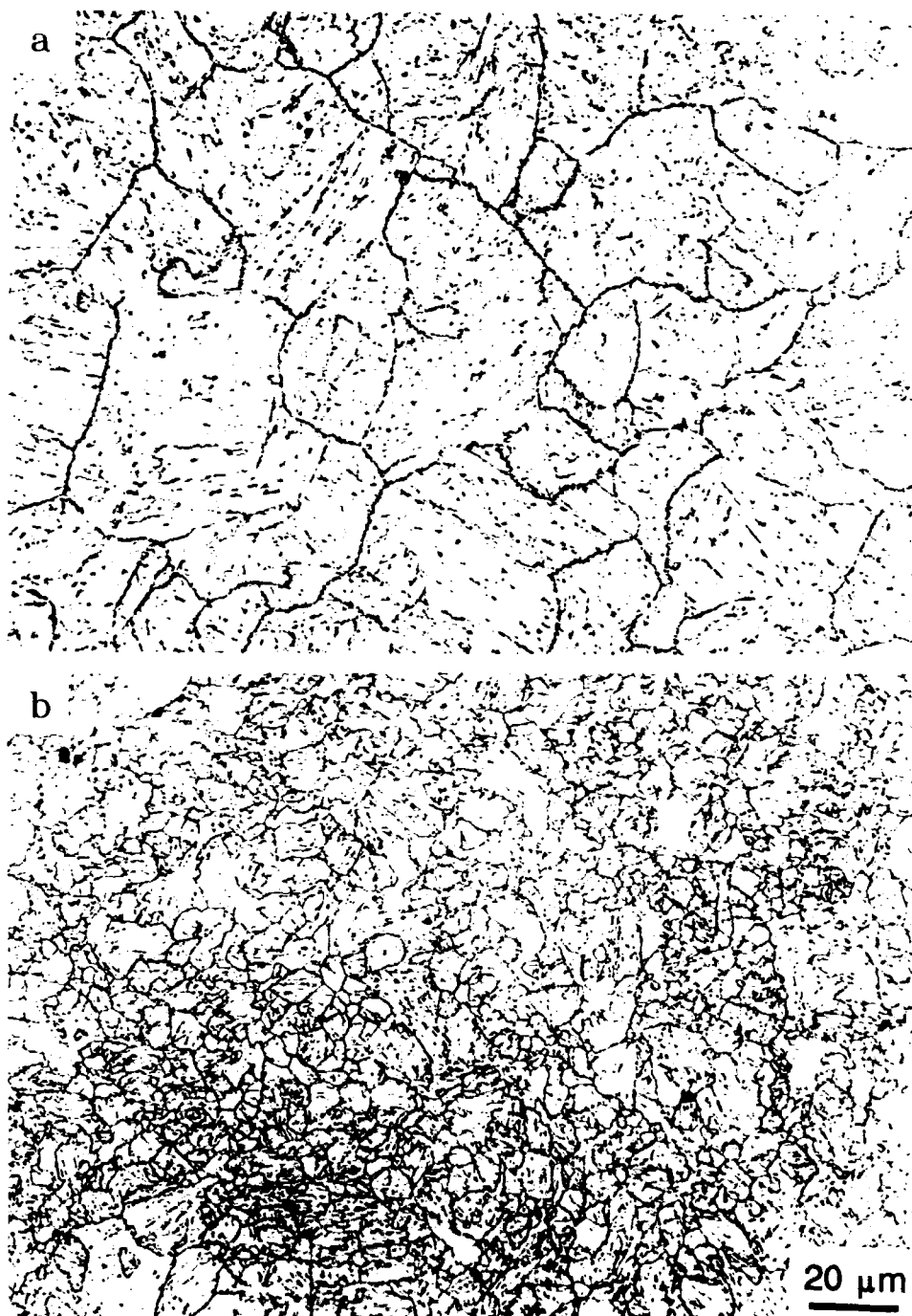


Figure 3.5 (a) Prior austenite grain size (about 30 μm) after one reaustenitizing treatment followed by quenching. (b) Prior austenite grain size (about 7 μm) after two reaustenitizing treatments followed by quenching.

for steels A and C. This section presents the results of these studies.

3.4.1 Dilatation Curves

A typical dilatation curve for steel B is shown in Figure 3.6. C-strain is plotted versus temperature for cooling of typical samples. For the 2X condition, these data are for cooling after the second austenizing treatment. Austenite transformation start and finish temperatures were determined by inflections in the C-strain versus temperature curve. The well-defined austenite transformation start and finish temperatures observed for all of the dilatation curves is an indication of significant transformation of the parent austenite phase. The change in C-strain at the start point is abrupt, but slows down by decreasing the temperature. This indicates that most of the transformation occurs closer to the start points (also see the CCT diagrams).

There is no significant difference between the dilatation curves for the specimens examined at slow and fast cooling rates. In addition, the dilatation curves for all of the three steels are similar.

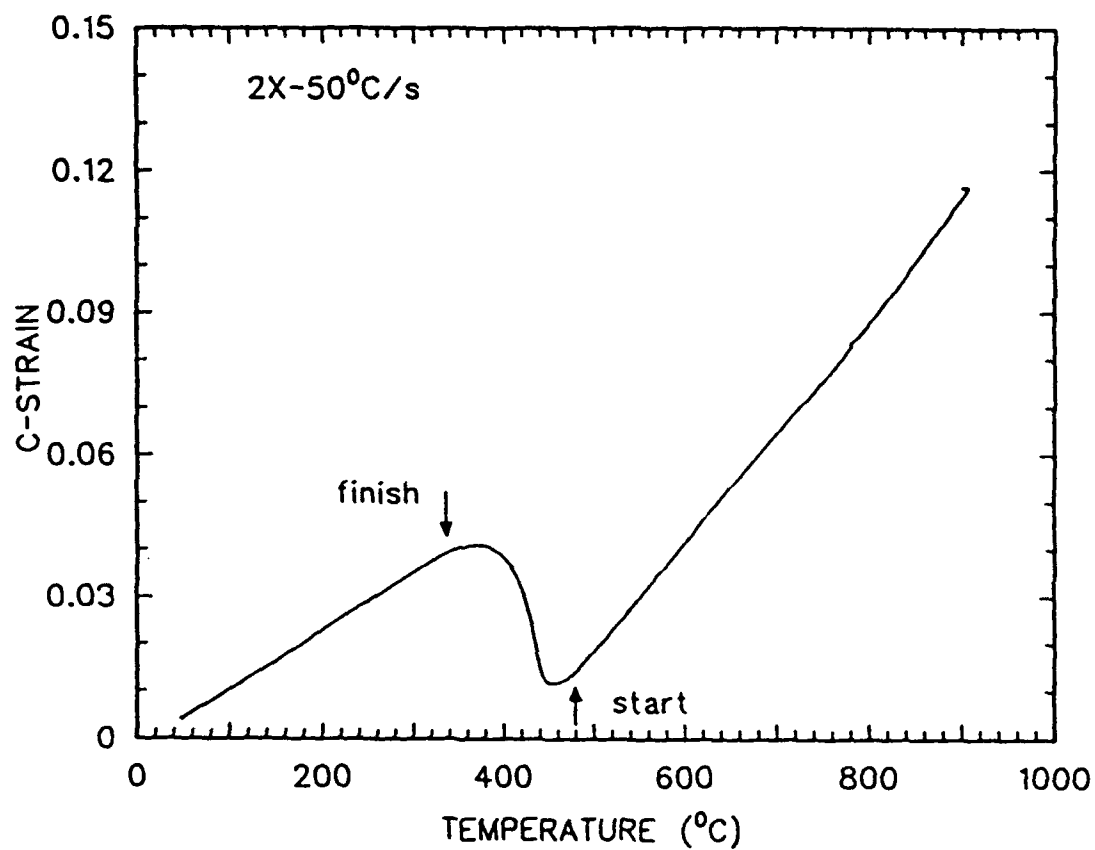


Figure 3.6 Plot of C-Strain versus temperature for a sample reaustenitized twice followed by cooling at 50°C/s after the second austenitizing treatment.

3.4.2 Continuous Cooling Transformation Diagrams

The results of 15 tests for each condition (1X and 2X) are plotted versus log time in order to outline the general form of the CCT diagrams. Figure 3.7 shows the CCT diagram for steel B. As was expected, the 2X condition (finer grain size) exhibits higher transformation start and finish temperatures. The CCT curves are consistent with several other studies (20,35,36,62) in that there is a broad, flat-topped transformation region typical of bainitic steels. In contrast to some of the aforementioned studies (20), there is no indication of high-temperature transformation products (polygonal ferrite). This could be due to the high molybdenum content of the present steels. The presence of molybdenum and other strong carbide formers lead to a solute drag like effect (SDLE) (90-91) which has been discussed before. All the transformation start temperatures are about the same for all cooling rates, with slight variations which could be attributed to experimental errors.

The 25, 50 and 75 percent transformation temperatures for the 1X and 2X conditions are shown in Figure 3.8a and 3.8b, respectively.

As these figures show, the lines representing the 25%, 50% and 75% transformation are much closer to the transformation

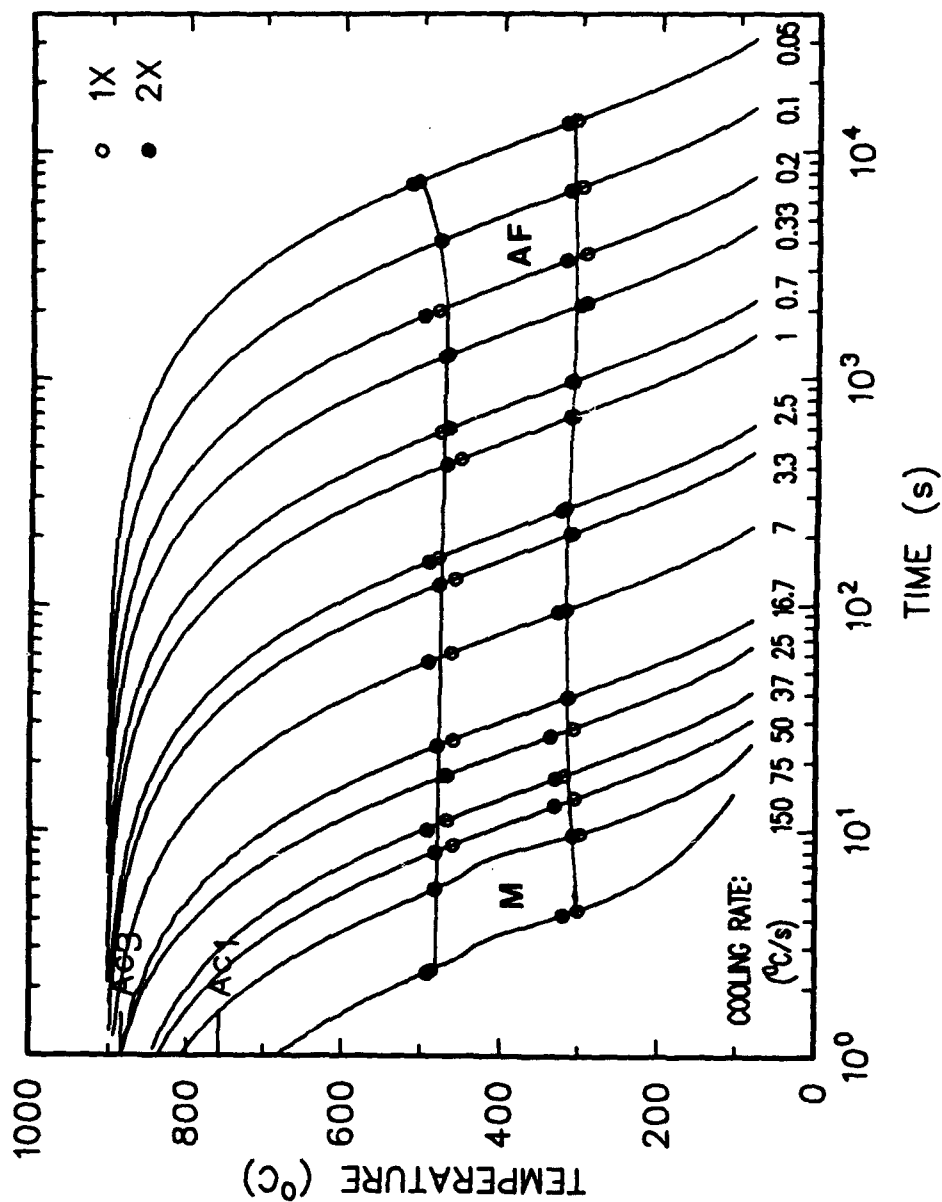


Figure 3.7 CCT diagram for steel B.

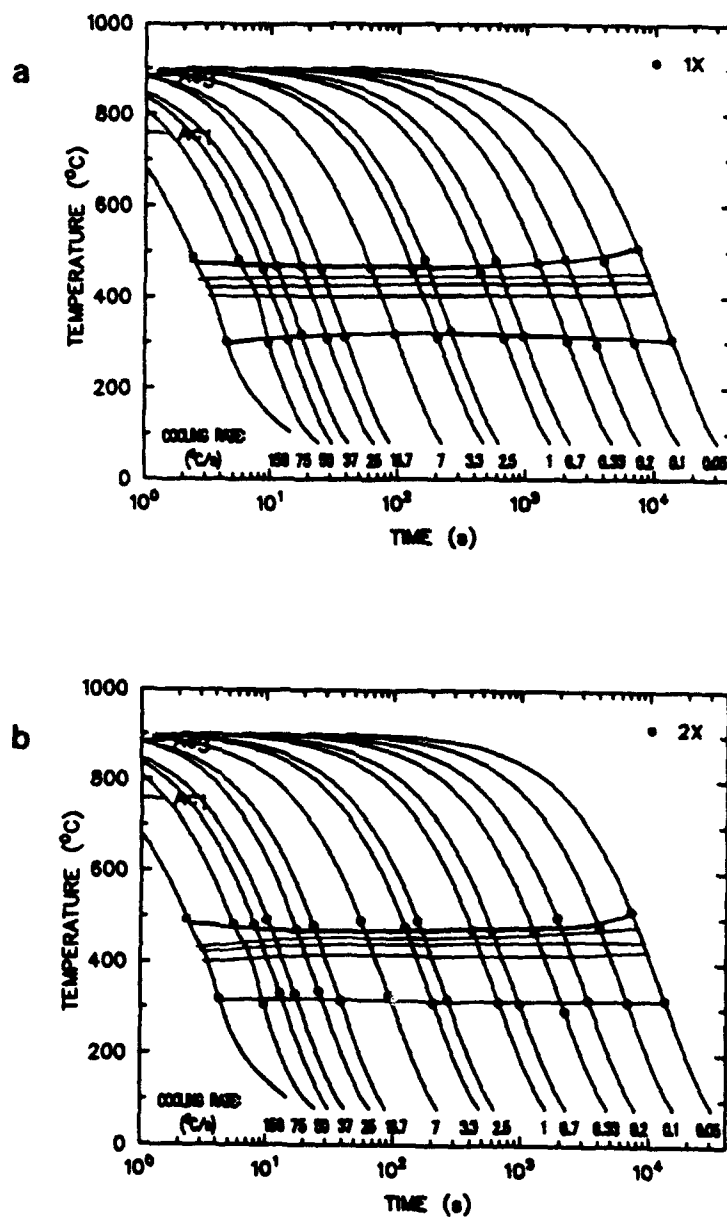


Figure 3.8 25, 50 and 75 percent transformation lines for (a) 1X and (b) 2X conditions.

start temperatures than the transformation finish temperatures. This observation indicates that the reaction occurs to a greater extent at higher temperatures. Table 3.5 presents the calculated and measured B_s and M_s temperatures for steels A, B and C.

Table 3.5 Calculated and measured B_s and M_s temperatures for steels A, B and C.

STEEL	measured B_s	Calculated B_s	Measured M_s	Calculated M_s	
				H^*	A^{**}
A	495	280	495	375	405
B	482	245	482	361	392
C	458	225	458	355	386

* Ref 124

** Ref 123

To identify the microstructures which form as the result of different cooling rates, microhardness tests and light microscopy were conducted for each cooling rate. Hardness values for each cooling rate were plotted versus $\log \Delta t_{8-5}$ to show a "hardness profile." A change in the slope of the hardness profile was taken to indicate a change in the transformation products, or a change in the amounts of two different austenite decomposition products. Figure 3.9 shows the hardness values for both the 1X and the 2X

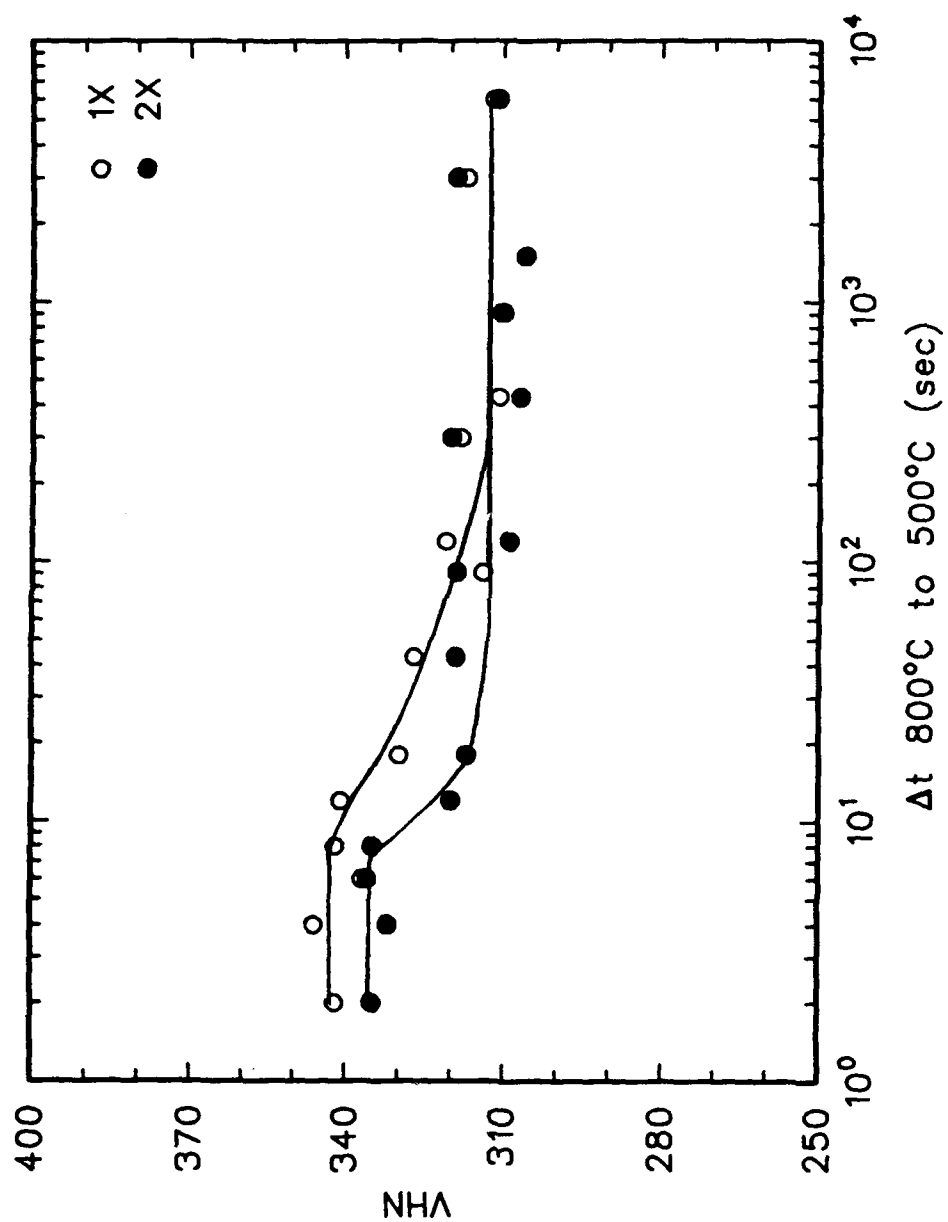


Figure 3.9 Microhardness versus cooling time between 800°C and 500°C.

conditions for steel B. Although at slower cooling rates the hardness values of both the 1X and 2X conditions seem to be similar, with increasing cooling rates these values are consistently higher for the 1X conditions than those of the 2Xs. The lower hardness of the 2X condition in the martensite region may be related to precipitate formation and coarsening.

Four cooling rates out of each condition were chosen for more detailed studies. These cooling rates are 0.1, 3.3, 25 and 150°C/s. What follows is a discussion of the microstructures produced as a result of these cooling rates, accompanied by the as-received microstructures of the steels under investigation.

3.5 Microstructural Characterization

Figure 3.10 shows a light micrograph of the structure generated as part of the Ac_1 - Ac_3 study, where the specimen was heated up to 1200°C and then cooled to room temperature. The measured cooling rate was 50°C/s. This microstructure strongly resembles the martensitic microstructures observed by other investigators (98,106,107,109) in low-carbon steels. The coarse prior austenite grain size (about 50-100 μm), which corresponds to



Figure 3.10 Martensitic microstructure of steel B after austenitization at 1200°C followed by quenching.

the high austenitizing temperature, provides excellent martensite hardenability compared with the finer grain sizes examined in the rest of study. Therefore this microstructure provides an excellent basis for comparison.

It was mentioned before that of etching these steels is very difficult. The similar orientation of martensitic laths within a prior austenite grain may be responsible for this difficulty. It should be mentioned here that prior austenite grain size, as well as cooling rate, had a noticeable effect on the etching behavior of these steels.

In contrast to the microstructure shown in Figure 3.10, recall the as-received microstructure (Figure 3.3). These microstructures seem to be very different from the martensitic structure in Figure 3.10. Etching of these as-received steels was relatively easy. In Figure 3.3, the thick, dark linear features are prior austenite grain boundaries and packet boundaries, while the small dark regions (arrowed) are islands of austenite and/or martensite. The size of these small islands ranges from about 1 to 2 μm , for larger blocky islands, to about 0.2-0.4 μm , for small islands. The bright areas are crystals and/or packets of acicular ferrite. These features are consistent with those observed in thin foils. Figure 3.11 shows a representative feature of the as-received microstructure in



Figure 3.11 Acicular ferrite with islands of retained austenite. (a) BF and (b) CDF images of the as-received steel B.

steel B. The small dark regions in the bright-field image (BF) are martensite-austenite islands (M-A constituent). These islands of retained austenite/martensite are aligned at both prior austenite grain boundaries and at interlath boundaries. The direction and alignment of these constituents indicates the acicular nature of this structure. Acicularity is more evident in the centered-dark-field (CDF) image of the same area in Figure 3.11. The diffraction pattern (DP) corresponding with this pair is shown in Figure 3.12, and the CDF image was produced with an austenite reflection. Notice the size of these islands, which are the same as those shown in light micrographs.

The acicular nature of the as-received steel B is more evident in Figure 3.13. The average lath width measured is about $0.8\text{ }\mu\text{m}$. An interesting features in these images is the jagged edges (arrowed) of a lath in this structure. There are two possible explanations for this feature. The first explanation that comes to the mind is that these jagged edges could be "bainitic" subunits. The repeated nucleation and growth model proposed by Oblak and Hehemann (31) suggests that the formation of bainite involves repeated nucleation of substructural units. A schematic presentation of this mechanism is shown in Figure 3.13c. The dark regions between light areas appear to be regions of

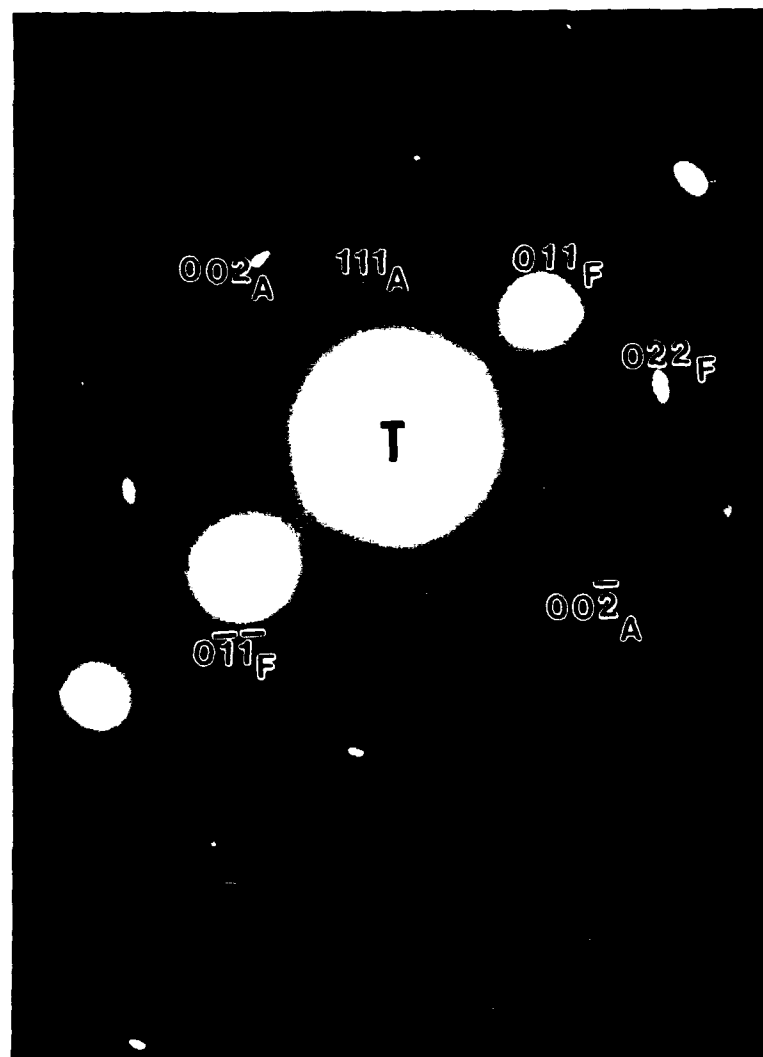


Figure 3.12 SADP corresponding to the region shown in Figure 3.11.

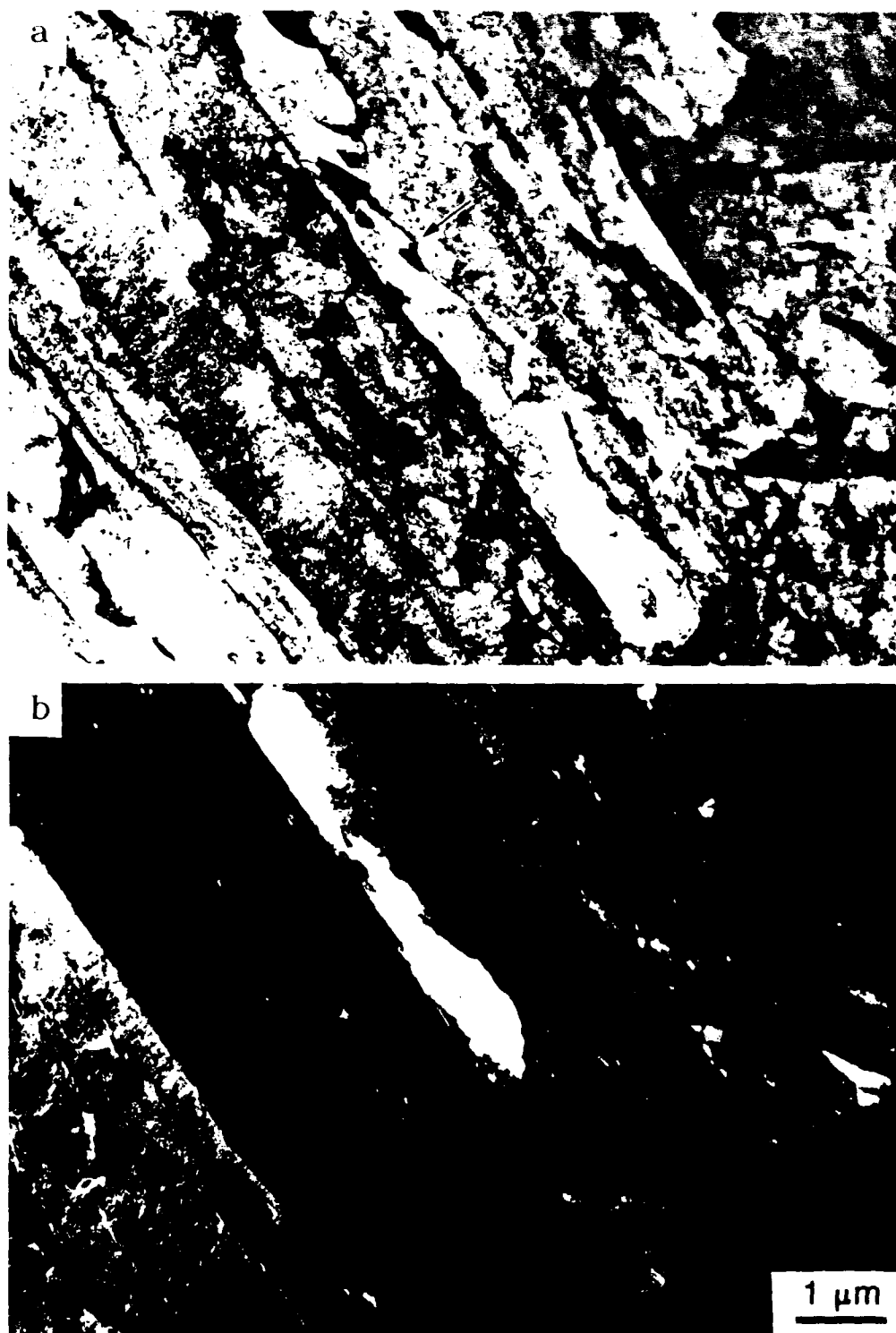


Figure 3.13 The as-received steel B showing acicular ferrite structure and the jagged edges of a lath. (a) BF and (b) CDF images.

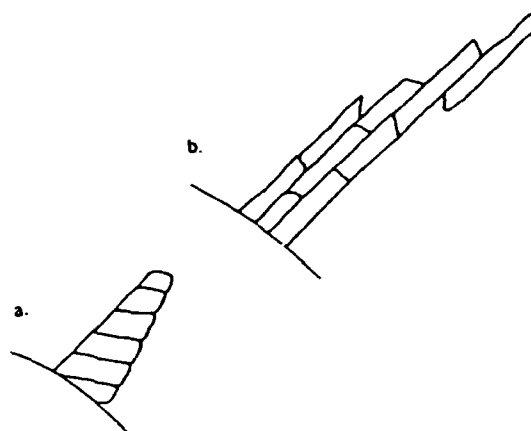


Figure 3.13c Schematic illustration of structures produced by repeated nucleation of subunits, which rapidly attain a limiting size (31).

high-carbon martensite and/or austenite entrapped between these subunits. The reason for retention of some untransformed austenite adjacent to ferrite laths is the enrichment of these regions with carbon as a consequence of transformation. Therefore, the formation of ferrite becomes progressively more difficult. It has been shown that the existence of M-A constituents depends on the cooling rate and the composition of steel (125). Coldren et al. suggested that rapid cooling tends to prevent the formation of M-A islands. At slower cooling rates, there is more time for diffusion of carbon and consequently more M-A islands are present (126).

A second possibility for the unusual image of Figure

3.13 is related to sectioning effects. These jagged edges could be the intersections of different laths. It is very likely that a combination of the two explanations is responsible for this phenomenon.

More evidence of retained austenite regions in the as-received steel is shown in Figures 3.14 and 3.15. These observations strongly dispute the conclusion made by DeArdo and co-workers (125) that these very low carbon steels lack the M-A constituent. DeArdo et al. showed no direct evidence to prove this conclusion. In fact, the presence of retained austenite at room temperature, considering the high transformation temperature of these steels (about 500°C), could be because the high strength of austenite does not allow the shear transformation to martensite to happen.

Figure 3.3 and Figures 3.10 through 3.15 show that distinct microstructural types are produced in steel B. These figures provide a basis for comparison for the remainder of the Gleeble work. Figure 3.16 shows representative light micrographs of steel B in the 1X and 2X conditions followed by cooling at 150°C/s. There are almost no similarities between these microstructures and what was shown in Figure 3.3 for the as-received steels. Unlike the as-received samples, these microstructures were very resistant to etching. Although the prior austenite grain



Figure 3.14 Microstructure of the as-received steel B.
Notice the black M-A islands.



Figure 3.15 Austenite centered-dark-field Image of the same region shown in Figure 3.14.



Figure 3.16 Martensitic microstructures in specimens cooled at 150°C/s. (a) 1X condition and (b) 2X condition. Light micrographs.

boundaries are not evident, the martensitic structure in Figure 3.16 is similar to what is shown in Figure 3.10. There is no strong evidence of retained austenite, especially in the 1X condition (Figure 3.16a). In other words, small, dark islands are mostly absent. The elongated ferrite packets resemble the blocky martensitic nature of the micrograph shown in Figure 3.10.

Figures 3.17 and 3.18 represent different regions of the thin foils examined for this cooling rate. These structures exhibit the fine lath structure typical of low-carbon martensite. The lath structure of martensite is much more evident in these micrographs. It is noteworthy that there was no evidence of retained austenite regions, as in the as-received structure, despite an extensive search. While this result is consistent with the light micrographs in Figure 3.16, other researchers (127) have reported thin films, and not equiaxed islands, in martensite. Still, no thin films of retained austenite were observed either. The average lath width for this cooling rate is about $0.2\text{ }\mu\text{m}$ for the 1X and the 2X conditions. A comparison between the lath width of the as-received steels and the specimens cooled at 150°C/s shows at least four times bigger lath size for the as-received condition. This is consistent with the previous observations indicating that the lath size decreases with an



Figure 3.17 Fine martensitic lath structure produced after cooling at 150°C/s for the 1X condition.



Figure 3.18 Fine martensitic lath structure produced after cooling at 150°C/s for the 1X condition.

increased cooling rate.

A comparison between the etching and electropolishing behaviors of the as-received structure and the specimens cooled at 150°C/s indicates major differences between these two structures. The as-received steels are much easier to etch than specimens cooled at 150°C/s and show better contrast in the light microscope. The strong etching effect was evident in electrochemically polished foils examined in the TEM as shown by Figure 3.14. In contrast, the martensitic specimens are not easy to etch, yet they electropolish much more uniformly. These observations, coupled with the existence of the upper plateau in the hardness profile for fast cooling rates, are indicative of differences in the mechanisms of austenite transformation for the as-received steels (cooling rate of about 1°C/s), versus the Gleeble samples cooled at 150°C/s. Figure 3.19 strongly supports the martensitic reaction at these high cooling rates. The presence of twinned martensite leaves no doubt about the martensitic nature of these microstructures.

The presence of the spherical particles arrowed in Figure 3.19 is probably a redeposition effect associated with thin foil preparation during electropolishing. These particles were not observed on surfaces of specimens from the as-received condition. This indicates that the

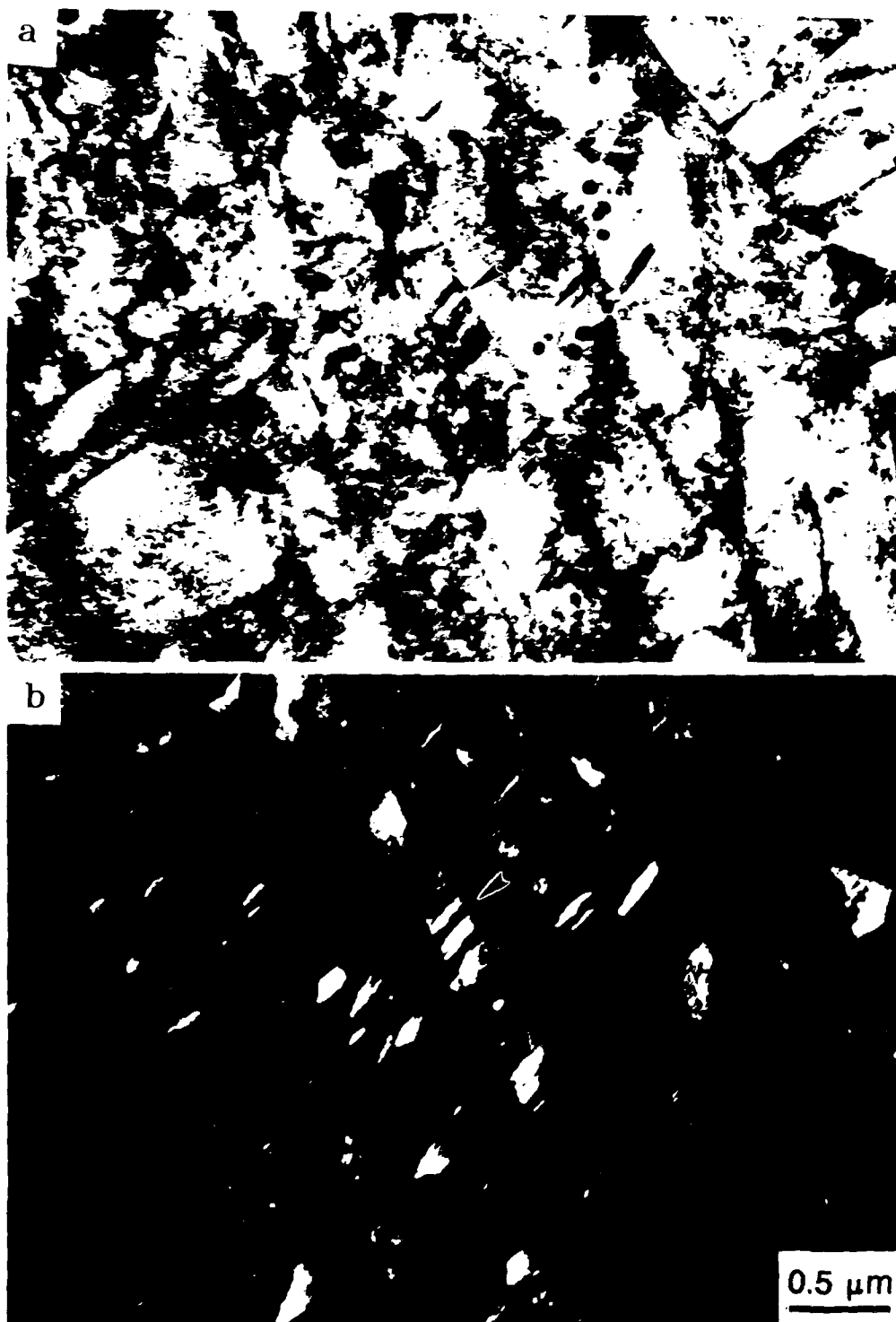


Figure 3.19 Twinned martensite in the sample cooled at 150°C/s, for the 1X condition.

reaustenitizing treatments are somehow responsible for the formation of these particles. EDS analysis was employed in order to establish the nature of these particles. The result of one analysis is shown in Figure 3.20. Because the EDS analysis did not show any evidence of an unusual concentration of any alloying elements in these particles, the nature, reason and mechanism of formation of these particles remain unclear.

Figure 3.21 represents the light micrographs for the cooling rate of 25°C/s, for both the 1X and 2X conditions. These microstructures are still very different from those of the as-received steels. Their etching behavior is the first noticeable factor. Here again, the etching was not as easy as with the as-received specimens, but the samples cooled at 25°C/s show much better contrast than those of the 150°C/s samples. There also appears to be more M-A constituent in these microstructures than those of the 150°C/s microstructures, particularly for the 2X condition. Recall the 2X micrograph for 150°C/s cooling rate. There was some evidence of dark, etched small particles in that microstructure which did not exist in the 1X condition. A comparison of these two cooling rates shows indications of some variations in the 1X and 2X transformation products. The first major difference between the 1X and the 2X

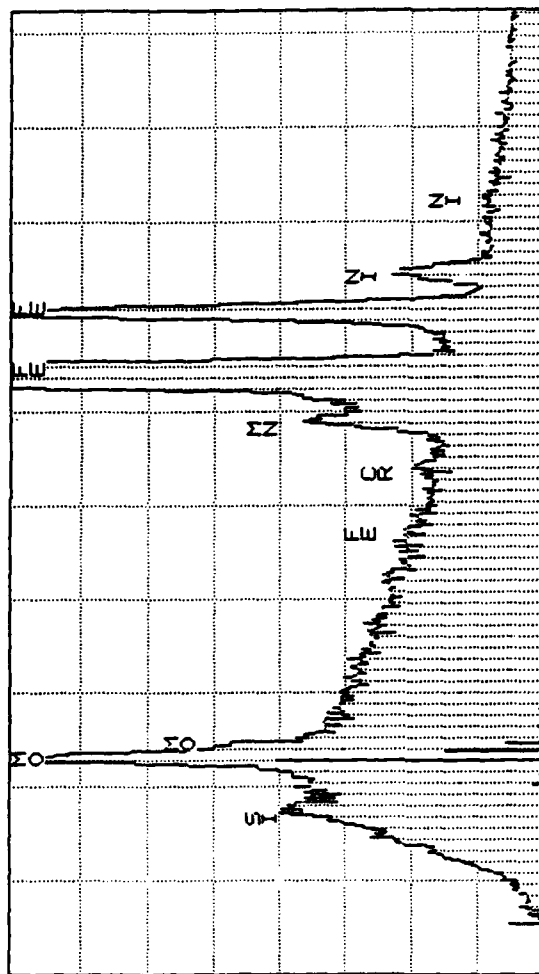


Figure 3.20 EDS analysis of the dark particles found on all of the TEM foils, except for the as-received condition.

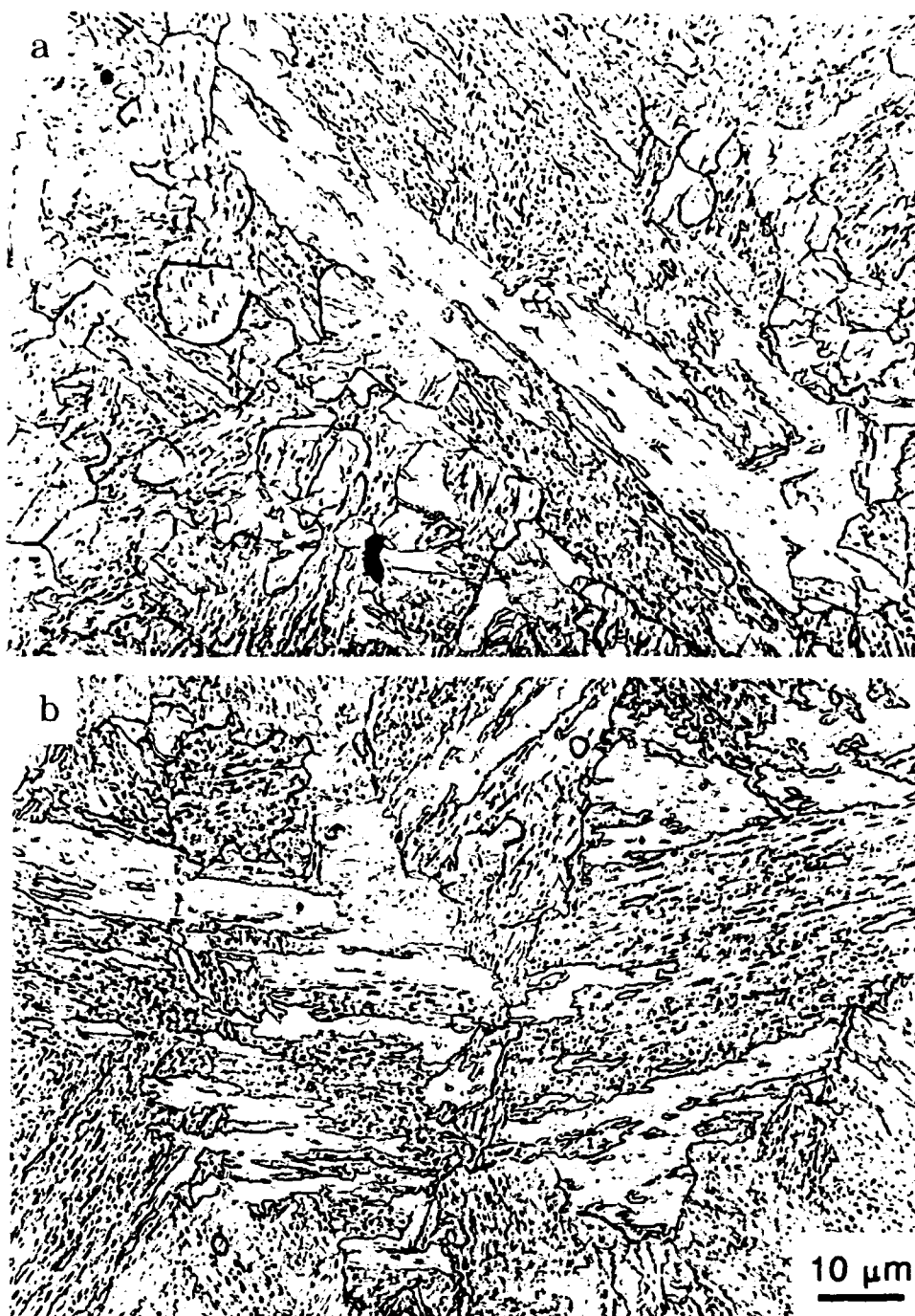


Figure 3.21 Light micrographs of the specimens cooled at 25°C/s. (a) 1X condition and (b) 2X condition.

conditions is in prior austenite grain sizes. Coupled with this difference, the 1X condition has a lower transformation temperature and a higher hardness value. This is consistent for all of the cooling rates. The double reaustenitizing treatment (2X) could give rise to a more complex transformation behavior which leads to these differences. It is very likely that the 2X samples contain more acicular ferrite and retained austenite than the 1X samples.

To get a better understanding of the transformation behavior of these steels at cooling rate of 25°C/s, examination of TEM micrographs is essential. Figures 3.22 and 3.23 represent the lath structure of the 1X and 2X conditions, respectively. A close look at these micrographs shows that the M-A islands are present, but the relative number density of these islands seems to be far less than that of the as-received steels. This structure is concluded to be acicular ferrite with M-A islands. In addition to this feature, the average lath width is about 0.35 μm , which shows about 0.15 μm increase compared to that of the 150°C/s cooling rate.

Figure 3.24 shows a pair of BF and CDF images for the 1X condition. These micrographs compare favorably with Figures 3.17 and 3.18, suggesting similar austenite transformation behavior. The only apparent difference is in



Figure 3.22 Microstructure of specimen cooled at 25°C/s for the 1X condition.



Figure 3.23a Acicular ferrite microstructure in specimen cooled at 25°C/s for the 2X condition, BF image.



Figure 3.23b Austenite centered-dark-field image of the same region shown in Figure 3.23a.



Figure 3.24a Martensitic microstructure in specimen cooled at 25°C/s for the 1X condition, BF image.

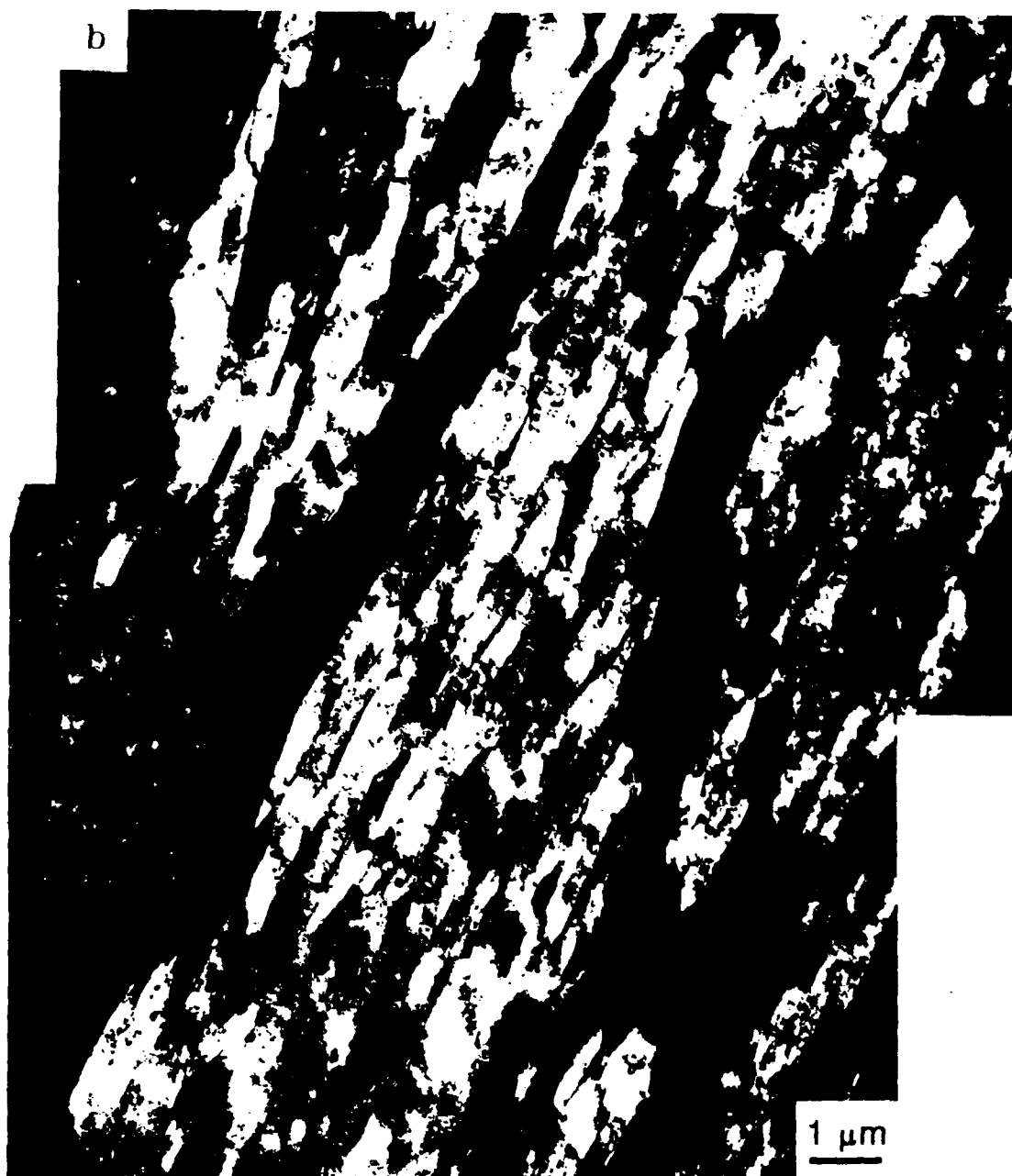


Figure 3.24b Martensite centered-dark-field image of the same region shown in Figure 3.24a.

the lath width, which is related to the cooling rate. The diffraction pattern related to these packets of martensite shows two dominant orientations, Figure 3.25. The angle between these two packets of martensite is 31.5° , which is a high angle boundary.

Figure 3.26 shows a dark-field image of the precipitates present within foils of the 2X condition. The origin and identity of these particles are uncertain.

A closer examination of TEM micrographs for the 2X condition shows that, opposed to the 1X condition, there are further differences between this treatment and the ones at 150°C/s . Figure 3.27 shows a pair of BF-CDF images for the 2X condition at 25°C/s cooling rate. The presence of M-A islands at a density similar to what was observed in the as-received steels is another indication that there is a transition in transformation mode in the vicinity of this cooling rate, at least for the 2X condition. The presence of the fine twinned martensite and interlath M-A islands in Figure 3.28 supports this hypothesis. A comparison between the hardness values shows that the hardness value for the 1X condition is on the upper plateau, but that of the 2X condition is in the lower plateau region. This evidence suggests that at 25°C/s there is a mixture of acicular ferrite and primarily martensite for the 1X condition, while

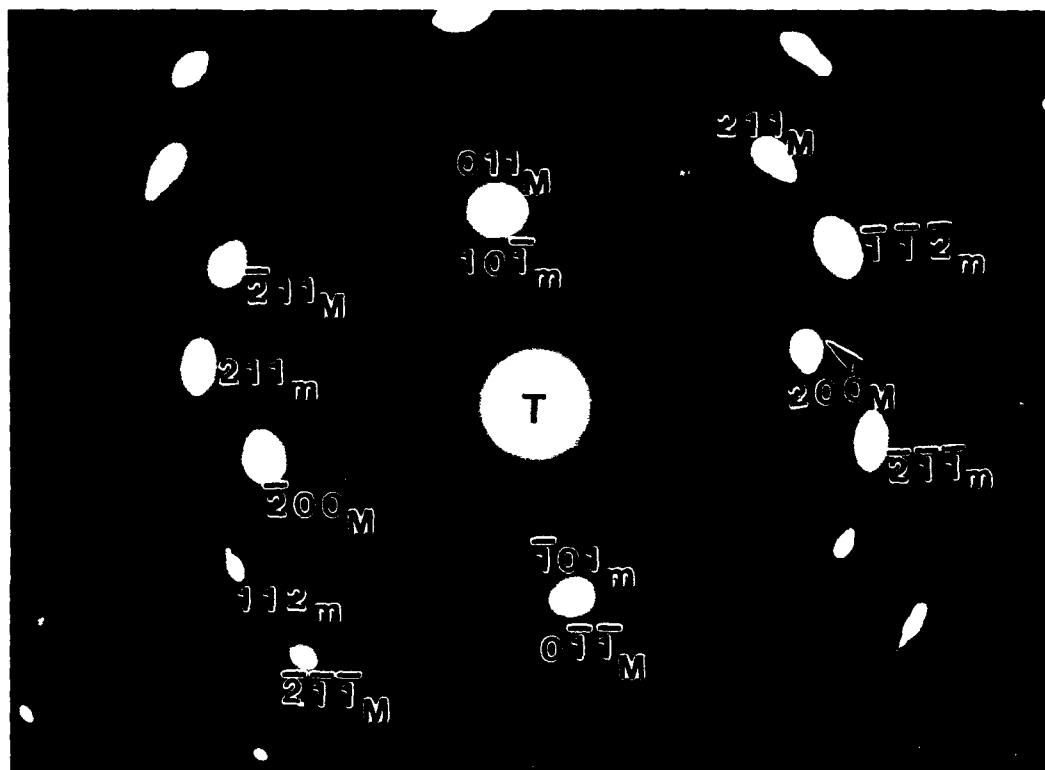


Figure 3.25 SADP from the region shown in figure 3.24. "M" and "m" are from two different martensite packets. "T" is the transmitted beam.

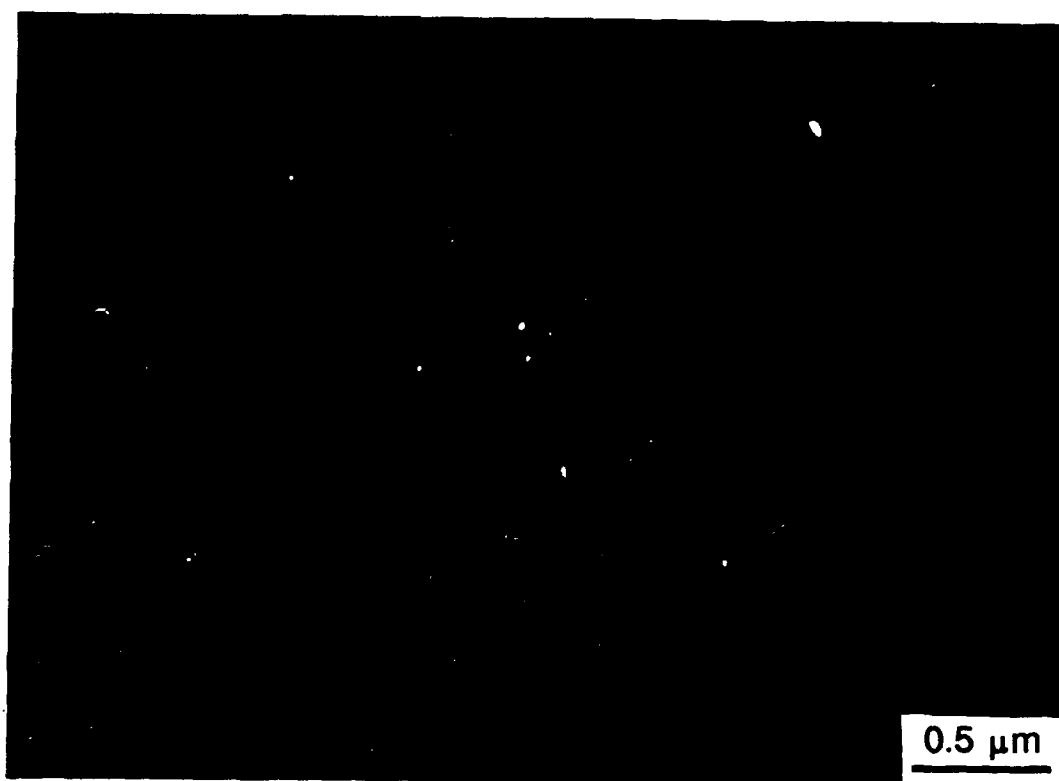


Figure 3.26 DF image showing precipitates observed within a TEM foil. Cooled at 25°C/s, 2X condition.

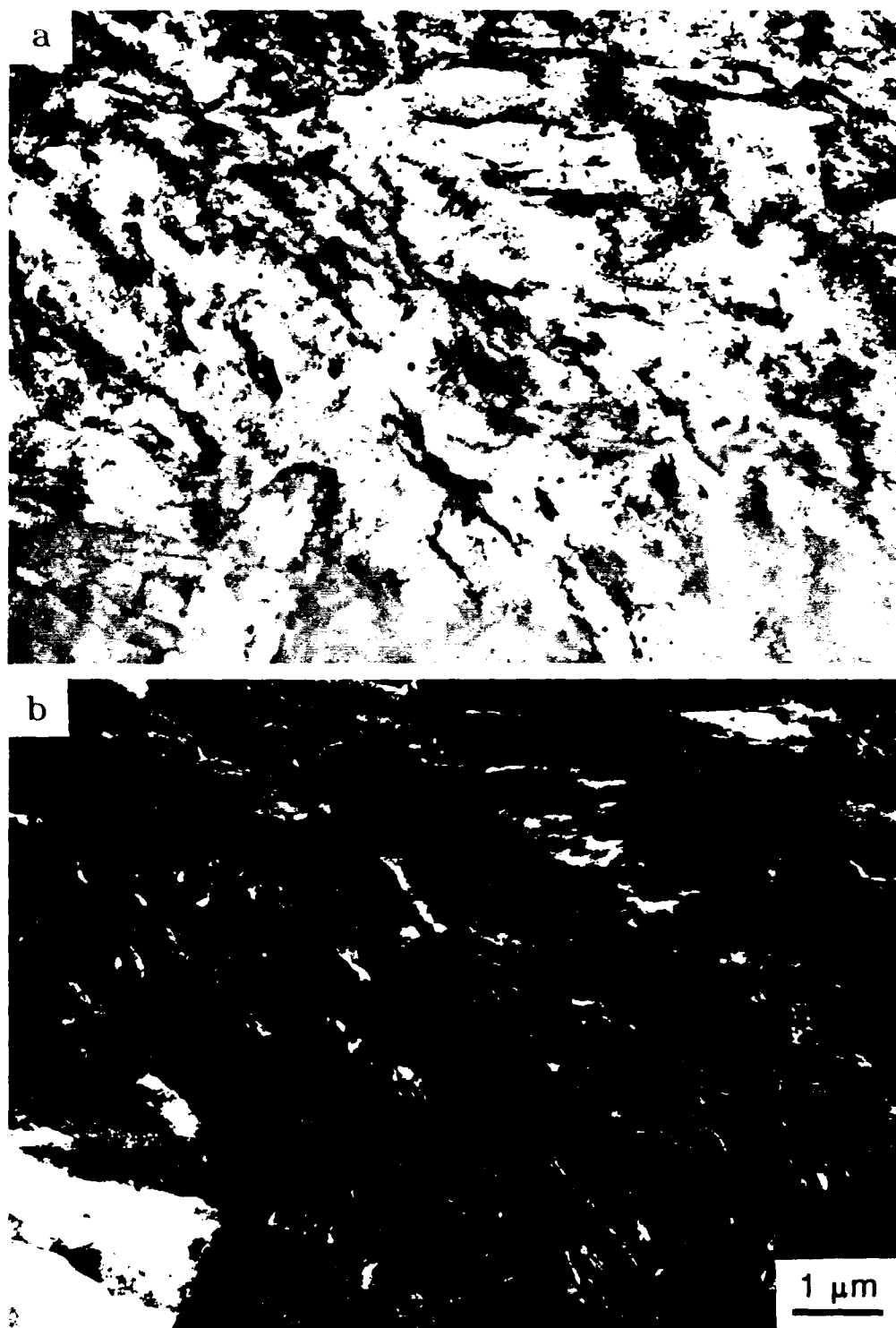


Figure 3.27 Acicular ferrite and retained austenite in specimen cooled at 25°C/s - 2X condition. (a) BF and (b) austenite CDF images.

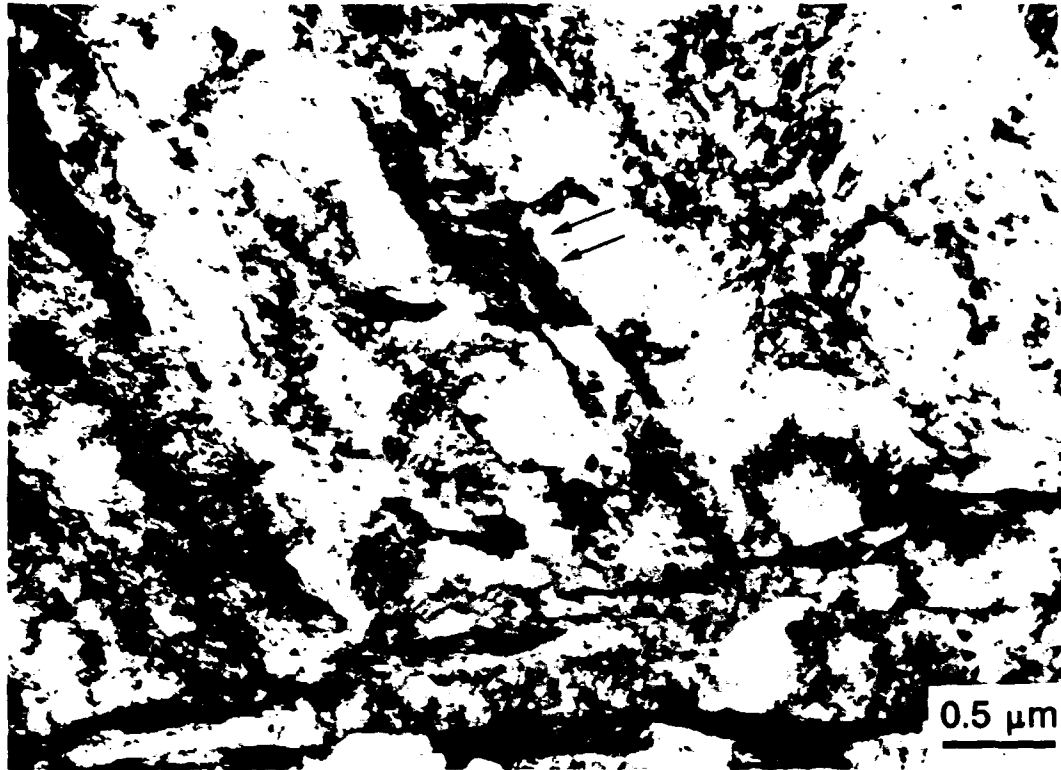


Figure 3.28 A pool of twinned martensite between laths of acicular ferrite in specimen cooled at 25°C/s, 2X condition.

the microstructure of the 2X condition consists mostly of acicular ferrite.

An example of a mixed microstructure is shown in the SEM micrograph of Figure 3.29. This microstructure represents the cooling rate of 7°C/s for a 1X condition. Although it is difficult to judge about the presence of martensite packets in this micrograph, the following reasons can explain why this microstructure contains both acicular ferrite and martensite. The acicular nature of the microstructure, the lath width and the M-A pool sizes are consistent with the previous observations. The hardness value for this cooling rate on the 1X hardness profile is not on the lower or upper plateau, thereby providing more evidence that this microstructure may be a mixture of acicular ferrite and martensite. The amount of acicular ferrite in the 2X condition compared to that of the 25°C/s -1X condition has increased, and that explains the improved etching behavior and the decrease in the hardness. At slower cooling rates (3.3°C/s), the microstructures shown in Figure 3.30 resemble those of the as-received specimens. Etching behavior in these samples was relatively good, as shown by the many clear features in the light micrographs. The number of retained austenite/martensite islands has increased, and the acicularity of the microstructure is

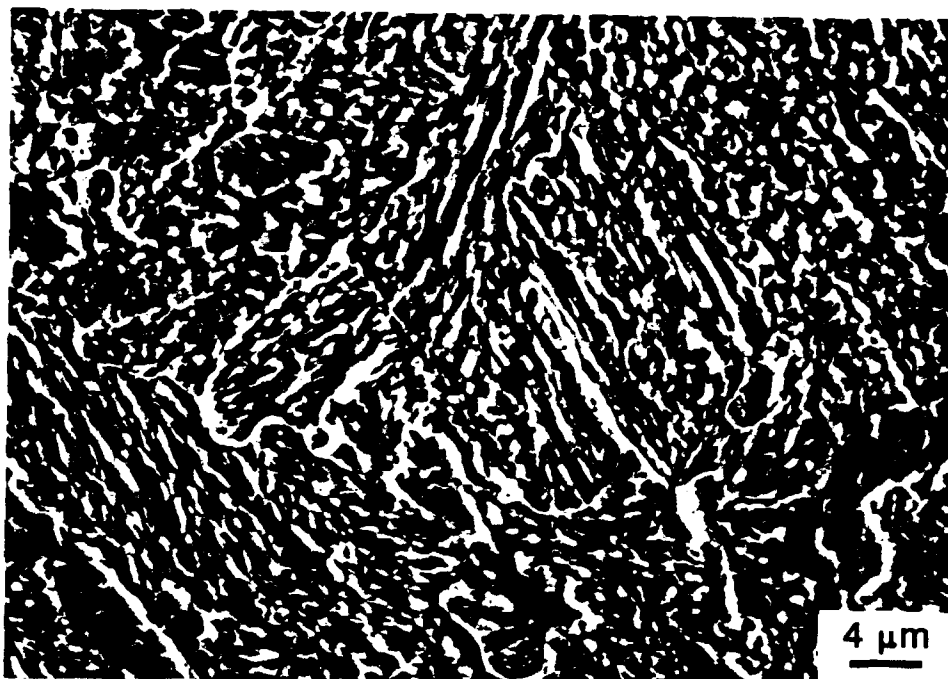


Figure 3.29 Acicular microstructure in a specimen cooled at 7°C/s, 1X condition. SEM micrograph.

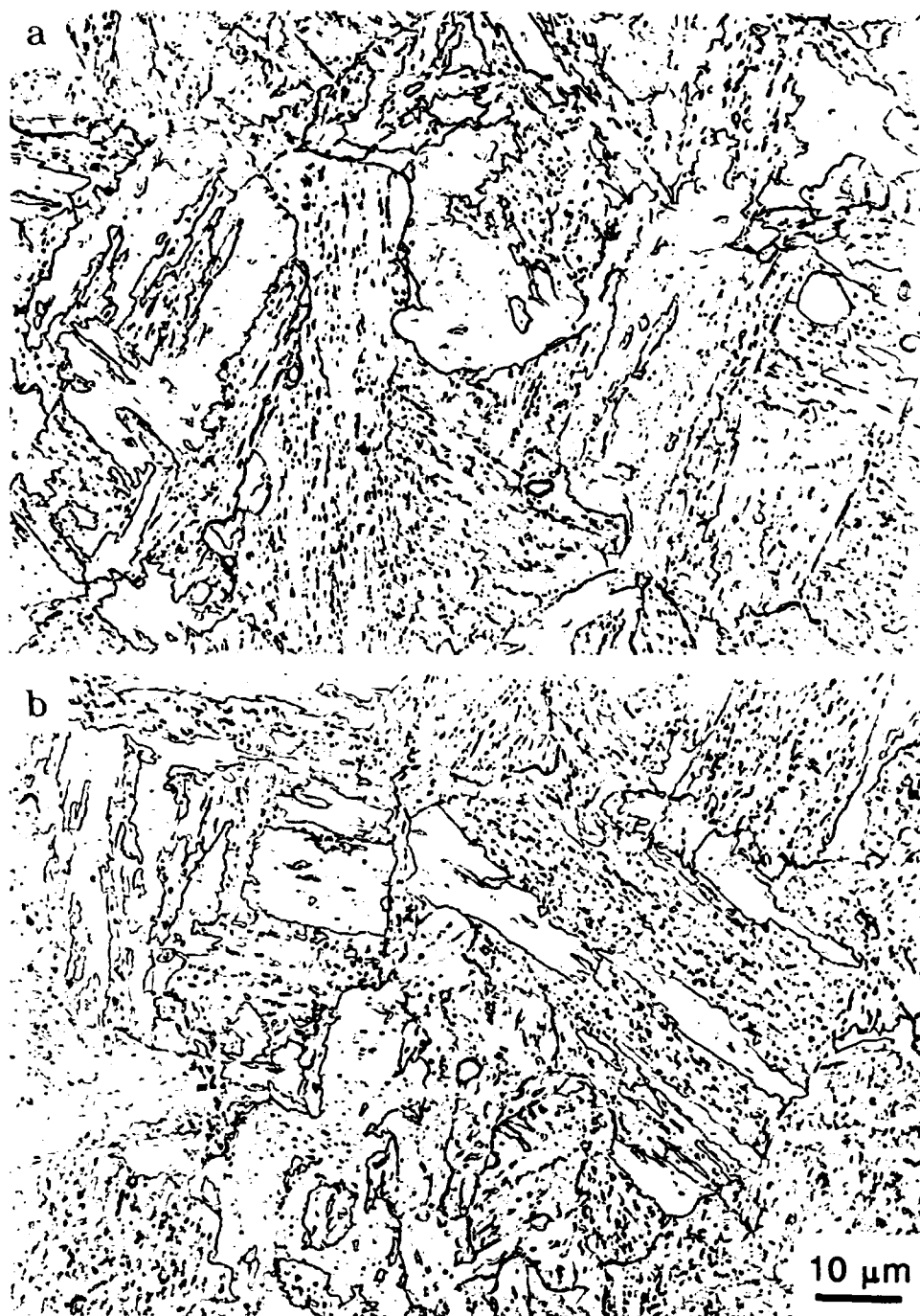


Figure 3.30 Light micrographs of the specimens cooled at 3.3°C/s. (a) 1X condition and (b) 2X condition.

still evident. There is no evidence of a granular structure, and this observation is consistent with the as-received structure. The hardness profiles for both the 1X and 2X conditions show that the hardness values of these specimens are in the lower plateau, which represents the acicular ferrite region.

Observations with the transmission electron microscope support the above conclusions. Figure 3.31 shows a representative micrograph of the specimen for the 1X condition. The ferrite laths are elongated and the dislocation density is high. Although the dislocation density was not measured, it seems to be consistent with the previous observations for the acicular ferrite microstructure. Some pools of retained austenite/martensite are present in this microstructure. In addition, as with the as-received microstructure, the jagged edges of some ferrite laths are evident (arrowed). This phenomenon was observed in the as-received samples and the possible reasons for that were explained in association with Figure 3.13c. The montage in Figure 3.32 represents a bigger picture of this microstructure. The acicular nature of ferrite laths is very obvious, and the islands of the M-A constituent are elongated between the ferrite laths. Considering this montage and the light micrograph for this cooling rate, it



Figure 3.31 The edges of "bainitic" ferrite subunits.
Specimen cooled at 3.3°C/s, 1X condition.



Figure 3.32 Acicular ferrite and M-A islands in specimen cooled at 3.3°C/s, 1X condition.

is clear that the rows of small dark islands in the light micrograph represent both acicularity of the microstructure and the retention of some austenite between ferrite laths. The measured average lath width for these microstructures is about $0.45\text{ }\mu\text{m}$ and the size of M-A islands are about 1 to $1.5\text{ }\mu\text{m}$, consistent with the other cooling rates.

It is interesting to notice the same directions of the ferrite laths within different packets. The low-angle features of the ferrite boundaries make them insensitive to etching, therefore only packet boundaries and interfaces between ferrite crystals and M-A islands can be revealed in the light microscope.

Figures 3.33 and 3.34 represent two different regions of a specimen cooled at 3.3°C/s for the 2X condition. There is not much difference between the 1X and the 2X conditions. The presence of small round particles indicates the same redeposition effect observed at the higher cooling rates. As mentioned before, reaustenitizing treatments may be responsible for this phenomenon. In particular, precipitation and coarsening may occur at 905°C .

The light micrographs in Figure 3.35 represent the microstructure of both 1X and 2X conditions for specimens cooled at 0.1°C/s . The relatively good etching behavior of these specimens is similar to that of the as-received and



Figure 3.33 Acicular ferrite in specimen cooled at 3.3°C/s, 2X condition.



Figure 3.34 Acicular ferrite in specimen cooled at 3.3°C/s, 2X condition.

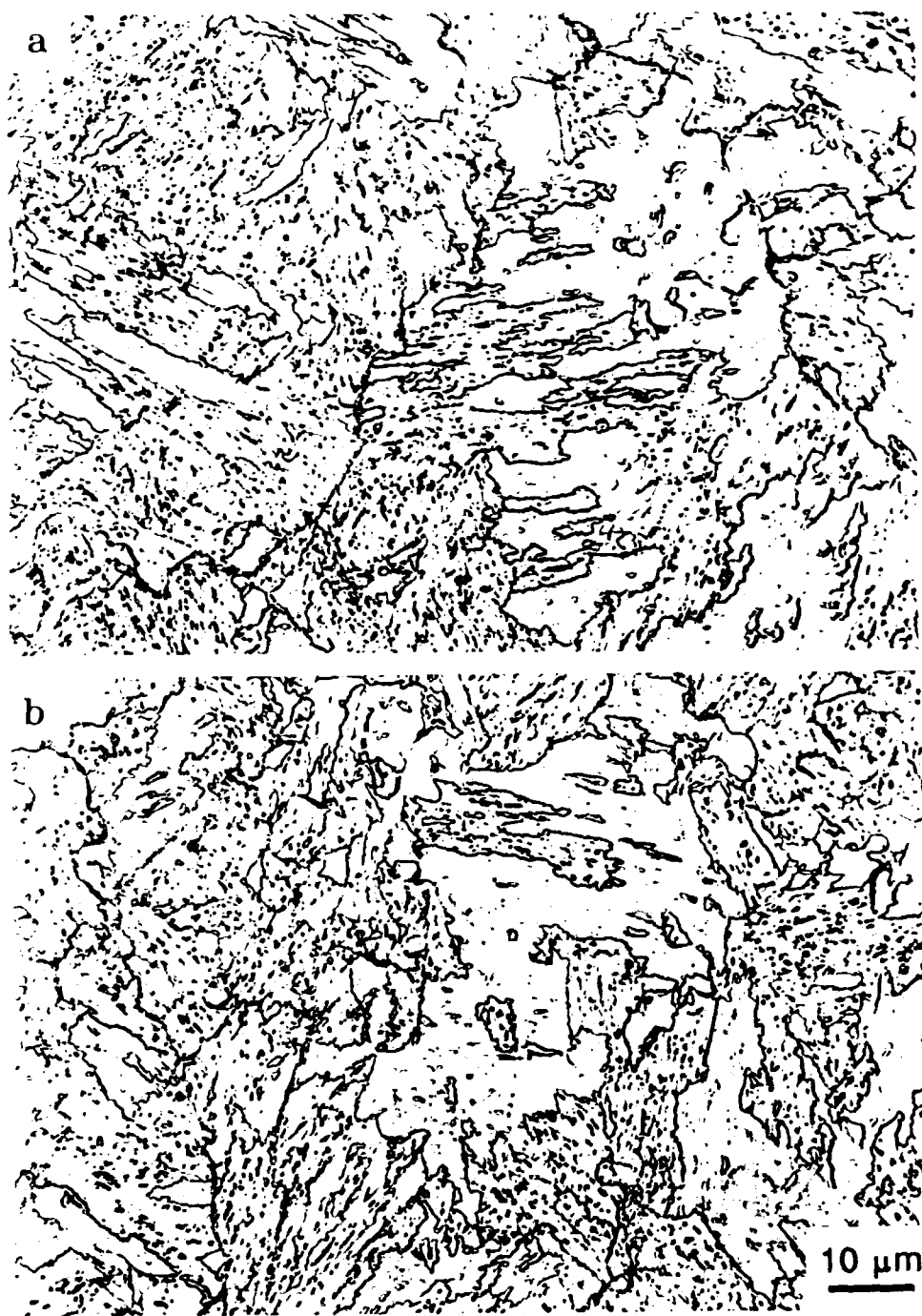


Figure 3.35 Light micrographs of the specimens cooled at 0.1°C/s. (a) 1X condition and (b) 2X condition.

the 3.3°C/s cooling rate specimens. All the features in these microstructures are similar to what was observed for the 3.3°C/s cooling rate. Although the cooling rate is very slow, the rows of retained austenite pools are indicative of the acicular nature of these microstructures. The hardness values of both the 1X and 2X conditions are in the lower plateau, an implication of non-martensitic structure. There is no indication of a granular structure.

TEM micrographs show the acicularity of these microstructures in more detail. Figure 3.36 is a montage (with a very slight gap) from the 2X condition. Notice the acicular ferrite laths and their interwoven nature. A large packet of ferrite "C" has separated the two sections of packets "A" and "B". A BF montage for the 1X condition is shown in Figure 3.37, and the CDF corresponding to the same area is represented in Figure 3.38. Notice the similarity of this microstructure with that of the as-received specimens. The different directions of ferrite packets are clearly shown in Figure 3.38.

Polishing behavior is the reason that some areas in these micrographs do not show the elongated laths. As mentioned before, the etching and polishing behaviors of these specimens act inversely. The electropolishing and thinning behavior is very good for the martensitic



Figure 3.36 Acicular ferrite in specimen cooled at 0.1°C/s, 2X condition.



Figure 3.37 Acicular ferrite in specimen cooled at 0.1°C/s, 2X condition. BF image.



Figure 3.38 CDF image of the same region shown in Figure 3.37.

specimens, but they do not etch well. In opposition to this behavior, the acicular ferrite structure responds better to the etching, but does not polish well. This behavior makes the thin areas on the TEM foil very small. Such behavior also explains why the etching behavior of the specimens cooled at different cooling rates can be a basis for comparison of the microstructures corresponding to each cooling rate.

In addition, an obscure but interesting feature was occasionally observed in these microstructures. Figure 3.39 shows an example of this observation. A block of martensite about $5.5\text{ }\mu\text{m}$ long and $0.6\text{ }\mu\text{m}$ thick seems to be perpendicular to the direction of ferrite laths. An interlath island of retained austenite/martensite which is almost perpendicular to the direction of this block indicates the direction of ferrite laths. The exact reason for this phenomenon is not clear. The lighter marks on this block suggest a martensitic structure. The accumulation of carbon in this region could be responsible for this phenomenon.



Figure 3.39 Angular martensite island in specimen cooled at 0.1°C/s, 2X condition.

3.6 Mechanical Properties

3.6.1 Tensile Properties

Tensile data for as-cooled (1°C/s , 1X and 2X) and as-cooled-plus tempered steels A, B and C are given in the appendix. These data include a 0.2% offset yield strength, tensile strength and total elongation.

Figure 3.40 represents these data in graphical form as a function of manganese content. As it is seen in this figure, the tensile strength and the yield strength increase with manganese content. The total elongation decreases slightly with an increase in manganese content, which is expected to demonstrate an inverse relationship with the ultimate tensile strength behavior. The increase in tensile strength is a reflection of the increase in the hardness values as a function of manganese content, as described before.

There are two features about the tensile behavior of these steels to discuss. First is the behavior of the steels as a function of reaustenitizing cycles. The second feature is the effect of tempering on the tensile behavior of these steels.

The data show that both the yield strength and the

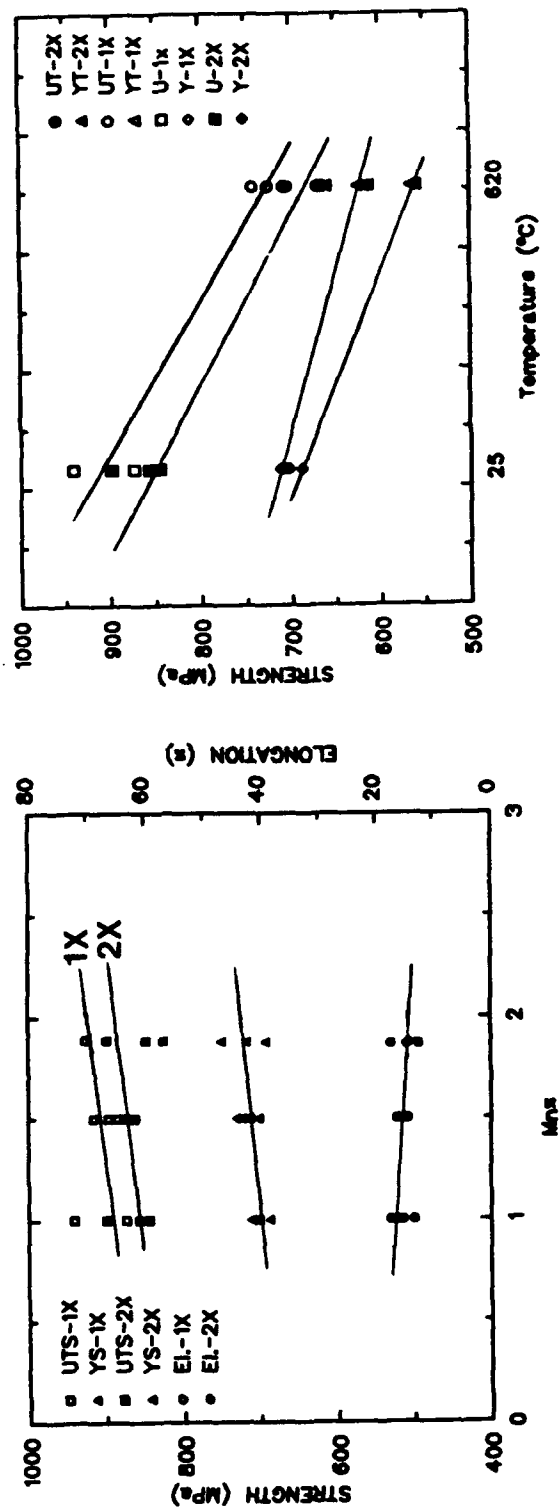


Figure 3.40 (a) Tensile data for steels A, B and C (1X and 2X conditions).
 (b) Tensile properties of steel A before and after tempering.

ultimate tensile strength are higher for the 1X condition. This is consistent with the hardness values as a function of cooling rate, Figure 3.9. Since tensile tests were conducted on specimens cooled at about 1°C/s, the specimens are expected to have a microstructure of acicular ferrite and M-A islands. The effect of grain size on the strength of steel has been discussed before. In acicular ferrite, several microstructural features play important roles in addition to the prior austenite grain size. The lath and packet size (also the degree of misorientation between the packets), the average dislocation density and the M-A island size and number density are the major factors that influence the strength of steels with this microstructure (10,12,60). These features are influenced by the austenite transformation temperature and steel composition. The lower the transformation temperature, the more dislocated the structure, and the finer the lath size. This discussion provides the basis for explaining the difference in the tensile strengths of 1X and 2X conditions. Since the 1X condition has a lower transformation temperature, finer packet size and higher dislocation density should be expected.

Coldren et al. showed that the presence of M-A constituents increases the tensile strength of steel (126).

This phenomenon could be explained by the role of austenite-martensite islands as nucleation sources for mobile dislocations during deformation. It is possible that there is some difference in the average volume fraction of M-A islands which could lead to the difference in tensile strength between the 1X versus 2X conditions. Since the volume fraction of M-A islands has not been measured, this hypothesis is subject to verification.

The effects of tempering on the tensile properties of steel A for both the 1X and 2X conditions are shown in Figure 3.40b. Tempering at 620°C for 35 minutes caused a noticeable drop in the yield strength, but the effect on the tensile strength was dramatic. It has been shown (128) that tempering of low-carbon bainitic steels decreases their tensile strength and improves the fracture toughness. The increase in total elongation values for both 1X and 2X conditions is in agreement with previous observations.

3.6.2 Charpy Impact Properties

Charpy impact data for the as-cooled and the as-cooled-plus-tempered steels A, B and C are given in the appendix. These data include both the test temperature and the absorbed energy.

Figure 3.41 shows a graphical representation of these data. The data reveal that the 2X condition shows a better impact toughness and a lower transition temperature than the 1X condition. This is not true for steel C which contains 1.76% manganese. Generally, the effect of manganese on impact toughness is opposite of its effect on the strength of these steels. Steel A, with the lowest manganese level, shows the best impact toughness, while steel C has the lowest toughness among all of the three steels. This observation is true for both the 1X and 2X conditions.

The superiority of the 2X condition in impact properties over the 1X condition is due to the smaller prior austenite grain size and lower strength of the 2X condition. It is well known that grain refinement improves the impact properties of steel (3,128). The high-angle prior austenite grain boundaries are effective obstacles to propagation of cracks. When a crack reaches a high-angle grain boundary, a change in the direction of the crack is unavoidable. The propagation of the crack at this stage requires more energy. In discussion of the difference in tensile strengths between the 1X and 2X conditions, it was pointed out that this difference could be partly due to the presence of more M-A constituents in the 1X microstructure. Here, the higher transition temperature of the 1X condition could be related

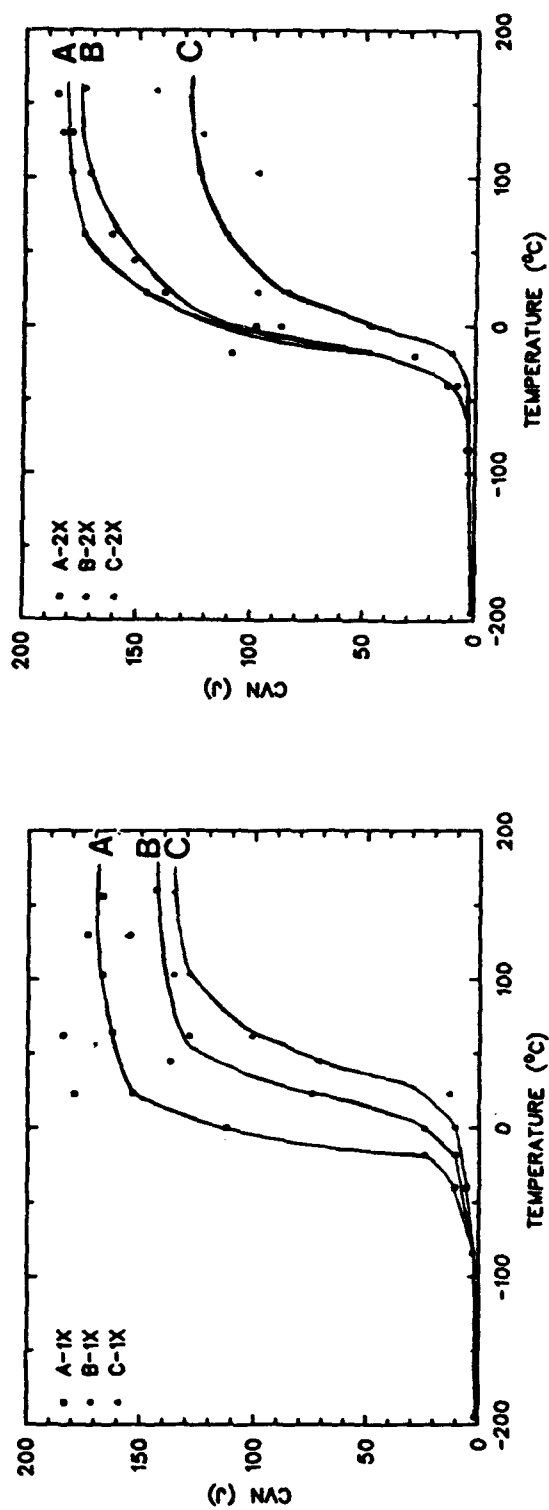


Figure 3.41 (a) Energy Absorbed versus temperature for steels A, B and C, 1X condition.
(b) Energy Absorbed versus temperature for steels A, B and C, 2X condition.

to the deleterious effect of the higher volume fraction of M-A islands (126). Figure 3.41c shows the impact properties of steel A for the 1X and 2X conditions which were tempered at 620°C for 35 minutes.

According to Coldren et al. (126), the deleterious influence of M-A islands can be eliminated by tempering at 540-650°C. DeArdo et al. (22) also showed that tempering of steels very similar to those in the present study improved the fracture toughness of those steels. In another investigation, DeArdo and coworkers (129) stated that ductile fracture characteristics in the transverse direction are related to the amount, size and shape of the nonmetallic inclusions present in the steel. The resistance to brittle fracture has been related to the metallurgical condition of the parent austenite (the grain size of the austenite is the most important factor), the cleanliness of the steel and the strength level. They also concluded that the presence of hard particles, either inclusions, carbides and/or high-carbon martensite islands, could have an adverse effect on brittle fracture resistance.

Upon tempering, some of the high-carbon metastable austenite could have transformed isothermally to ferrite-carbide aggregates. The embrittlement effect of the carbides and the variation in the carbide dispersion can

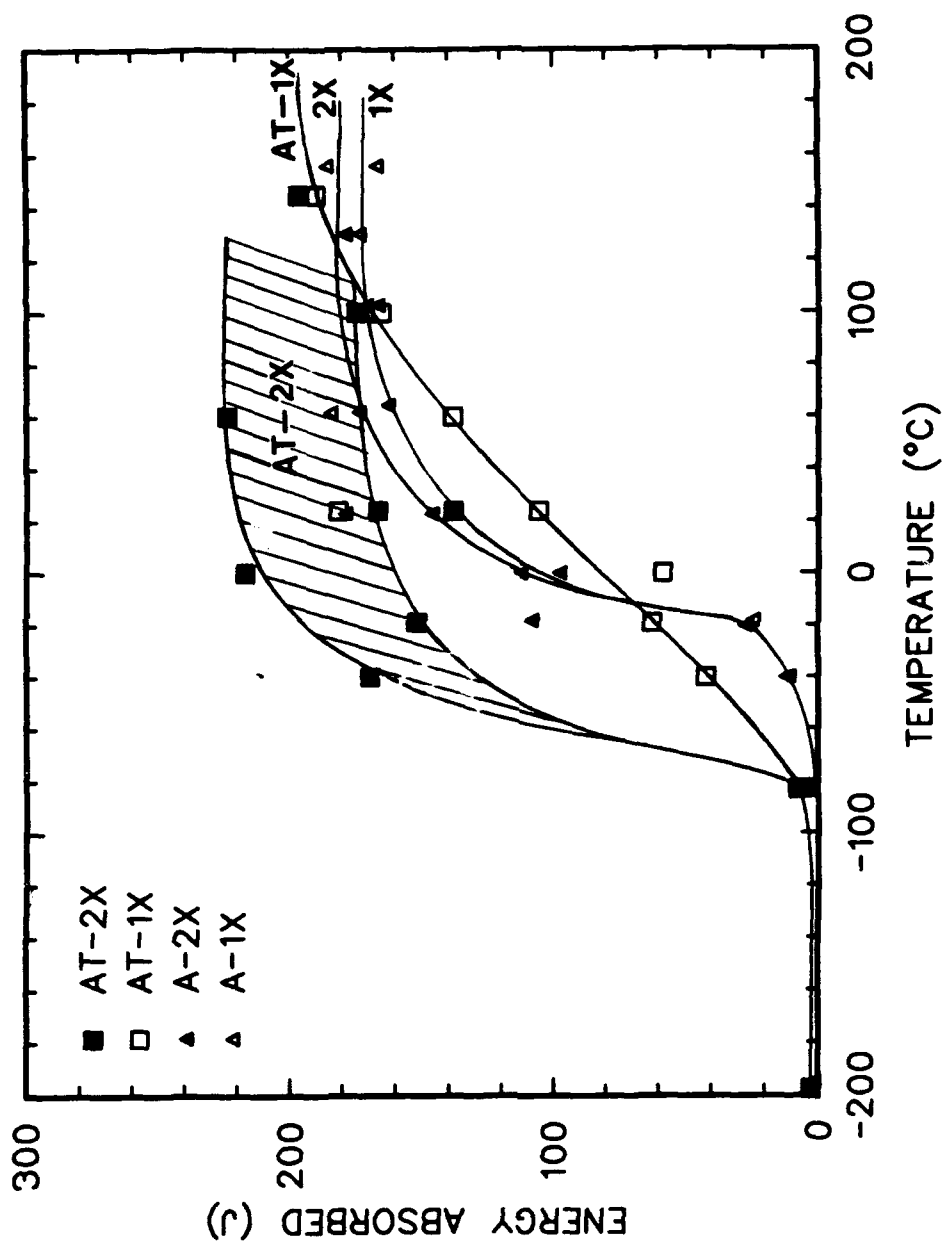


Figure 3.41 (c) Energy Absorbed versus temperature for steel A in the as-cooled and 620°C-tempered treatments. Includes data for the 1X and 2X conditions.

lead to lower toughness. This has been confirmed by Smith et al. (16).

With this in mind, tempering of the steels with the presence of M-A constituents may have mixed results. Although the transition temperature of the 2X condition has shifted to lower temperatures, the effect of tempering on the upper shelf energy is not very clear. The appearance of a "dual upper shelf" could be the result of the mixed effect of tempering when M-A constituents are present. There is another possibility for this behavior. A delamination effect, which will be shown in the fractography of the non-tempered specimen, could be responsible. One can also argue that the so-called delamination did not have such an effect on the as-cooled specimens, which may make the above argument invalid.

The effects of tempering showed an absolute inverse effect on the transition temperature of the 1X condition. Here again, the possibility of more M-A islands for the 1X condition arises. As indicated above and according to the previous observations, the presence of M-A islands could have negative effect on the toughness of steel after tempering.

3.6.3 Fractography

An investigation of the fracture surfaces of steel B for the 1X and 2X conditions was conducted using SEM. Three samples, from the lower shelf (-195°C), the transition region (23°C) and the upper shelf (100°C), were examined for each condition. Figure 3.42 shows the fracture surfaces of specimens tested at all three temperatures for the 1X condition. At 15X magnification, the transition from the brittle fracture to the ductile fracture is clear. Notice the "delamination" in Figure 3.42c.

Figure 3.43 represents the fracture surfaces for the 1X and 2X conditions at -195°C . It is clear from these pictures that a cleavage fracture developed in both conditions. The energy absorbed was about 3 J for both samples. A comparison between the facet sizes in Figures 3.43a and b clearly shows the larger facet size associated with the 1X condition.

Figure 3.44 shows the fracture surfaces for both the 1X and 2X conditions at room temperature. Both of the samples showed a mixed mode of fracture, although there was more ductile rupture in the 2X condition compared to the 1X. The energy absorbed for the 1X condition was 74 J while that of the 2X condition was 84 J. The higher strength and larger

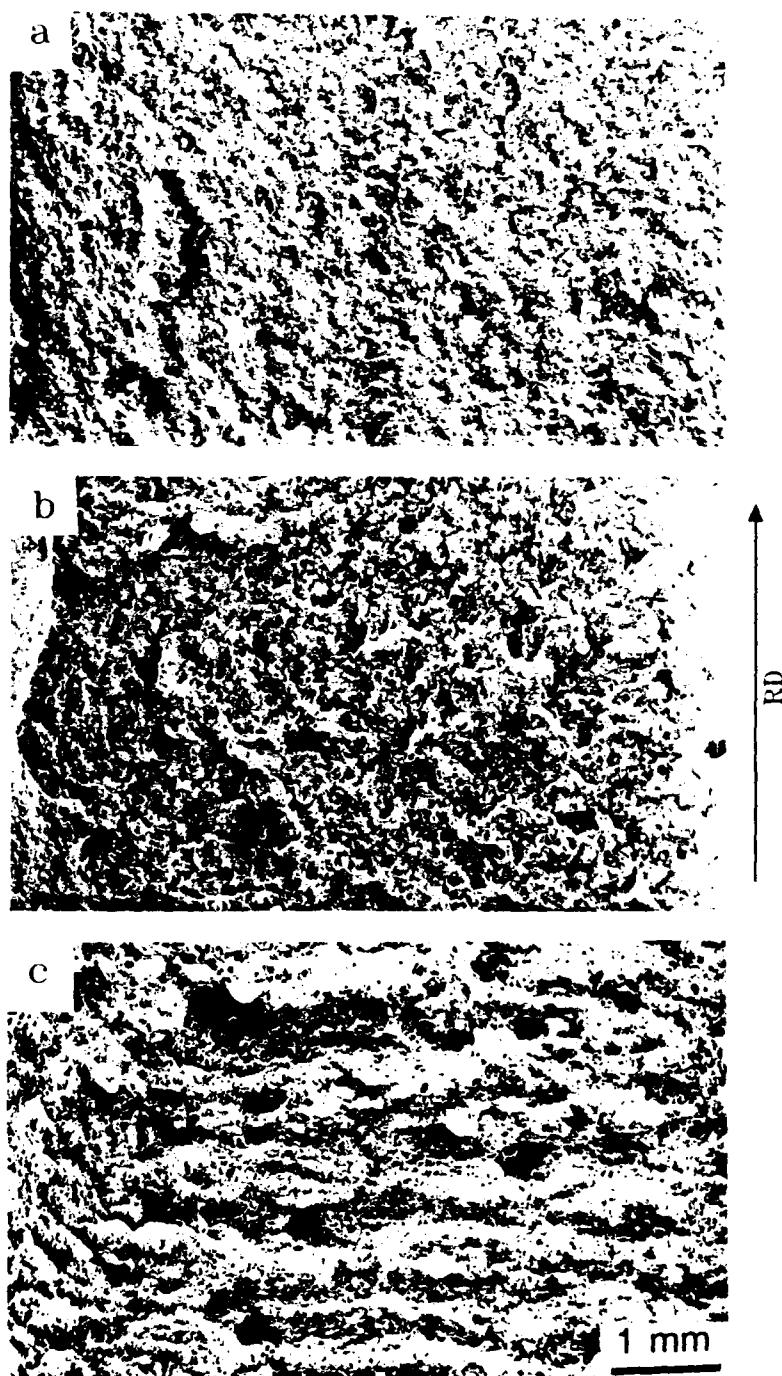


Figure 3.42 SEM micrographs of the fracture surfaces after testing at (a) -195°C , (b) 23°C and (c) 100°C . Steel B, 1X condition.



Figure 3.43 Fracture surfaces for steel B tested at -195°C .
(a) 1X and (b) 2X conditions.

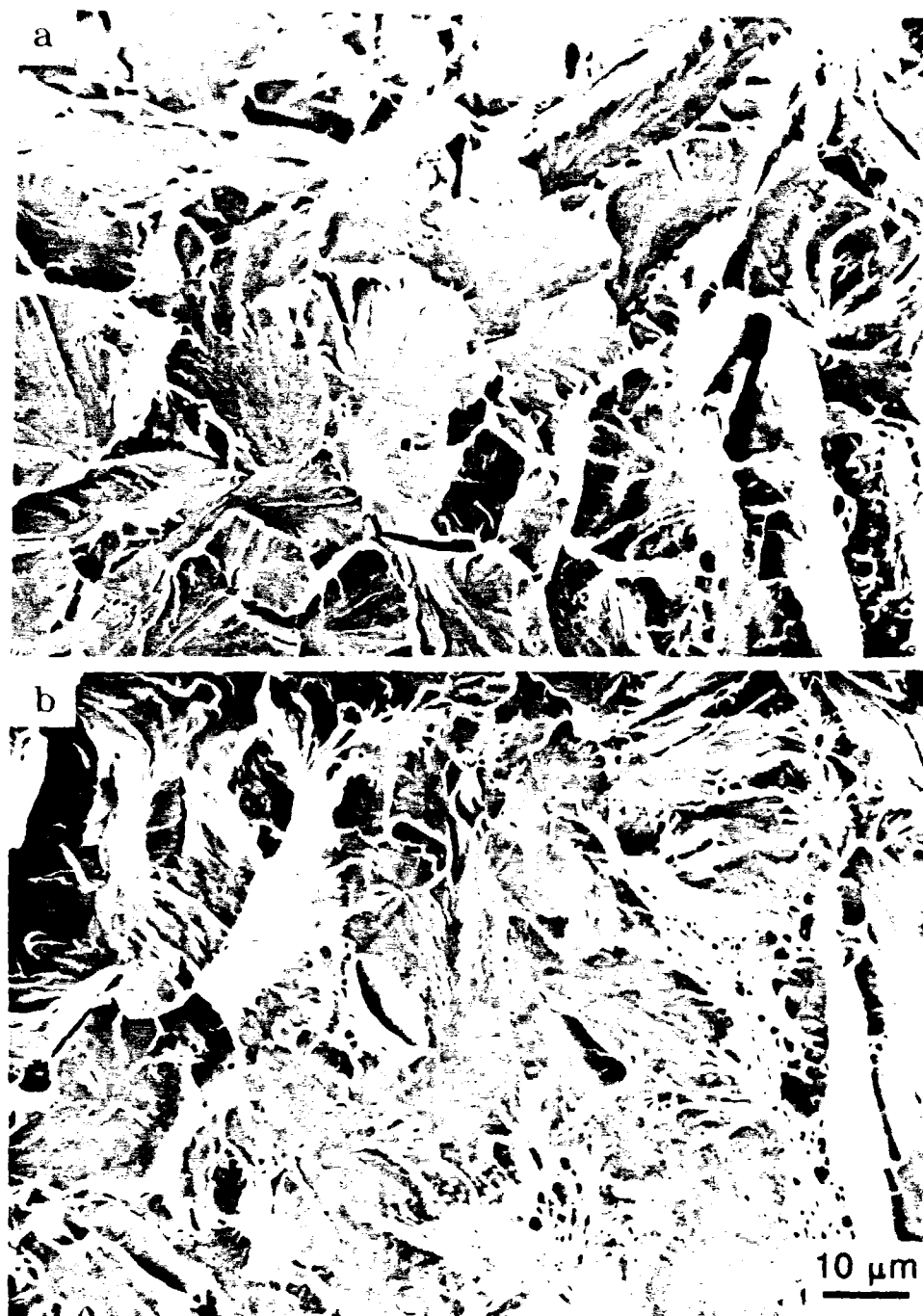


Figure 3.44 Fracture surfaces for steel B tested at 23°C.
(a) 1X and (b) 2X conditions.

grain size of the 1X condition explain this difference. The presence of small particles are apparent where ductile rupture has occurred.

The fracture surfaces of the specimens tested at about 100°C are shown in Figure 3.45. The energy absorbed by each sample was 167 and 179 J, for the 1X and 2X conditions, respectively. Ductile microvoid nucleation, coalescence and growth produce this type of fracture surface. The presence of inclusions within each dimple is in agreement with the role of the hard particles, including inclusions, in the initiation of fracture (128). Elongated inclusion particles in the rolling direction caused the elongated shape of the dimples, while round inclusions do not show the same effect. These observations reinforce the idea that cleanliness is an important factor in determining the fracture toughness of the steel (128,129). Irvine and Pickering (128) suggested that the presence of particles in every dimple is not necessary for the formation of dimples. Most of the dimples in these micrographs contain inclusions. This indicates the high inclusion content of these steels, which is probably responsible for their having a lower toughness than what was expected.

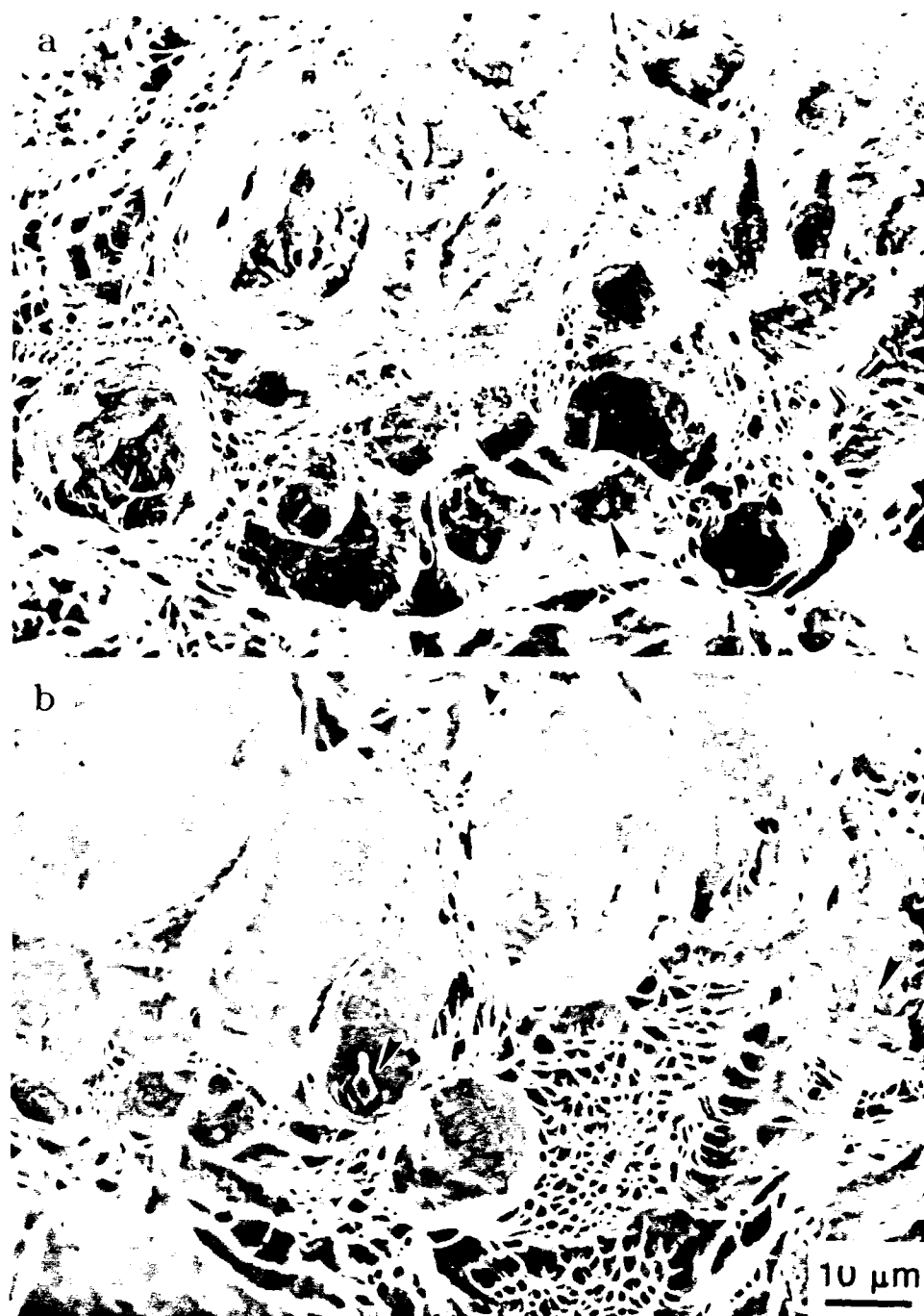


Figure 3.45 Fracture surfaces for steel B tested at 100°C.
(a) 1X and (b) 2X conditions.

3.7 Summary and Conclusions

The continuous cooling transformation of these very low-carbon steels revealed interesting features of their transformation behavior. Although the broad, flat-topped CCT diagrams do not present any noticeable indication of changes in the transformation products upon cooling at different cooling rates, the hardness profiles are clearly indicative of important changes. These changes are coupled with differences in the etching and electropolishing behavior of the specimens cooled at different cooling rates to prove the existence of changes in austenite transformation behavior.

The light microscopy and transmission electron microscopy of the specimens examined for generation of the CCT diagrams, as well as the as-received specimens, showed the following features:

1. A martensitic microstructure is the product of fast cooling rates. At cooling rates faster than about 25°C/s , the 1X condition (i.e. prior austenite grain size of about $30\text{ }\mu\text{m}$) exhibits predominantly martensitic microstructure. For the 2X condition (i.e., prior austenite grain size of about $7\text{ }\mu\text{m}$), martensite is formed at cooling rates faster than about 37°C/s . These structures do not respond well to the chemical etching, but their response to electropolishing

(thinning for TEM) is quite good. The upper plateau on the hardness profiles corresponds to the martensitic structure. Very fine, elongated and sometimes interwoven lath structures, in addition to some twinned martensite, are characteristics of this structure.

2. Acicular ferrite is the dominant austenite decomposition product for the cooling rates below about 7°C/s for the 1X condition. For the 2X condition this range is below about 25°C/s . Elongated ferrite laths, interlath retained austenite and/or martensite and high dislocation density are characteristics of this microstructure. The lowest cooling rate examined was 0.05°C/s . The etching response of these structures is much better than the martensitic structures, but their response to electropolishing is very poor. The lower plateau on the hardness profiles corresponds to the acicular ferrite structure.

3. At cooling rates intermediate to those at which the martensitic structure and the acicular ferritic structure form, mixed microstructures consisting of acicular ferrite and martensite form during cooling. The higher the cooling rate, the more martensite is in the microstructure. This is consistent with the higher hardness values for higher cooling rates.

4. The ferritic lath width increases with decreasing

cooling rate. This value is about $0.2 \mu\text{m}$ for the martensitic structure produced at a 150°C/s cooling rate, and increases to about $0.55 \mu\text{m}$ for the acicular ferrite formed upon cooling at 0.1°C/s . An anomaly was found in the as-received steel ($0.8 \mu\text{m}$), but this was probably due to an insufficient number of measurements.

5. Both the yield strength and the ultimate tensile strength increased with manganese content. However, the 1X condition shows higher strength than the 2X condition regardless of the manganese content. This is consistent with the hardness data. Tempering at 620°C for 35 minutes decreased both the tensile and the yield strengths of steel A.

6. The Charpy impact behavior of all three steels is consistent with other data. The 2X condition showed higher impact toughness and a lower transition temperature than the 1X condition. Generally, the effect of manganese on impact toughness is opposite its effect on the strength of these steels.

Tempering lowered the transition temperature and increased the toughness of steel A for the 2X condition. For the 1X condition, the transition temperature increased after tempering. This is attributed to the possible higher number density of M-A islands in the 1X condition than the 2X condition.

REFERENCES

1. E.E. Fletcher, High-Strength, Low-Alloy Steels: Status, Selection and Physical Metallurgy, Battelle Press, Columbus, Ohio, 1979.
2. F.B. Pickering, "High-Strength, Low-Alloy Steels - A Decade of Progress", Proceedings of an International Symposium on High-Strength, Low-Alloy Steels, Microalloying 75, Union Carbide Corp., New York, N.Y., 1977, pp. 9-31.
3. F.B. Pickering, Physical Metallurgy and the Design of Steels, Applied Science Publishers, London, England, 1978.
4. W.C. Leslie, The Physical Metallurgy of Steels, McGraw-Hill, Tokyo, Japan, 1982.
5. J.M. Rigsbee and P.J. VanderArend, "Laboratory Studies of Microstructures and Structure-Property Relationships in 'Dual-Phase' HSLA Steels", in Conference Proceedings: Formable HSLA and Dual-Phase Steels, ed. A.T. Davenport, AIME, New York, N.Y., 1977, pp. 56-86.
6. R.L. Brockenbrough, "Design Trends for High-Strength, Low-Alloy Steels", Processing and Properties of Low Carbon Steel, ed. J.M. Gray, AIME, New York, N.Y., 1973, pp. 389-397.
7. I. Takahashi, T. Kato, T. Tanaka and T. Mori, "Development of a High Strength, Hot Rolled Steel With 100 ksi Yield Strength", in Conference Proceedings: Toughness Characterization and Specifications for HSLA and Structural Steels, ed. P.L. Mangonon Jr., AIME, New York, N.Y., 1979, pp. 308-337.
8. J.D. Evensen, J. Lereim and J.D. Embury, "Microstructural Aspects of the Fracture Parameters of Controlled Rolled HSLA Steels", in Conference Proceedings: Toughness Characterization and Specifications for HSLA and Structural Steels, ed. P.L. Mangonon Jr., AIME, New York, N.Y., 1979, pp. 187-215.

9. High-Strength, Low-Alloy and High-Strength Intermediate Manganese Steel, AISI Steel Products Manual 17, 1967. From ref. 4.
10. F.B. Pickering, "The Structure and Properties of Bainite in Steels", in Symposium: Transformation and Hardenability in Steels, Climax Molybdenum Co., Ann Arbor, Michigan, 1967, pp. 109-132.
11. R.W.K. Honeycombe and F.B. Pickering, "Ferrite and Bainite in Alloy Steels", Metallurgical Transactions, Vol. 3, 1972, pp. 1099-1112.
12. A.J. McEvily, R.G. Davies, C.L. Magee and T.L. Johnston, "Structure, Hardenability and Toughness of Low-Carbon High-Strength Steels", Symposium: Transformation and Hardenability in Steels, Climax Molybdenum Co., Ann Arbor, Michigan, 1967, pp. 179-194.
13. A.J. McEvily and C.L. Magee, "Transformation-Structure-Property Characteristics of Low-Carbon Alloy Steels", in Low-Alloy Steels, ISI Publication 114, London, England, 1968, pp. 111-117.
14. K.J. Irvine and F.B. Pickering, "Low-Carbon Bainitic Steels", JISI, London, England, 1957.
15. A.P. Coldren, Y.E. Smith and R.L. Cryderman, "Microstructure and Properties of Low-Carbon Mn-Mo-Cb Steels", Processing and Properties of Low Carbon Steel, ed. J.M. Gray, AIME, New York, N.Y., 1973, pp. 163-189.
16. Y.E. Smith, A.P. Coldren and R.L. Cryderman, "Manganese-Molybdenum-Niobium Acicular Ferrite Steels With High Strength and Toughness", Symposium: Toward Improved Ductility and Toughness, Climax Molybdenum Co., Ann Arbor, Michigan, 1971, pp. 119-142.
17. R.R. Irving, "A Cost Effective Replacement for HY-80", Iron Age, May 16, 1986, pp. 53-55.
18. A.P. Coldren and T.B. Cox, "Development of 100 ksi Yield Strength HSLA Steel", AMAX Report No. CPR-20, 1986.
19. L.G. Kvidahl, "An Improved High Yield Strength Steel for Shipbuilding", Welding Journal, Vol. 64, 1985, pp. 42-48.

20. D.J. Colvin, "The Physical Metallurgy of Some Low-Carbon, Cu-Bearing HSLA Plate Steels", M.Sc. Thesis, T 3600, Colorado School of Mines, Golden, CO, 1988.
21. Sherri A. Richardson, "The Effects of Thermal Processing on the Microstructure and Mechanical Properties of a HSLA-100 Plate Steel", M.Sc. Thesis, T 3949, Colorado School of Mines, Golden, CO, 1990.
22. M.R. Blicharski, C.I. Garcia, S. Pytel and A.J. DeArdo, "Structure and Properties of ULCB Plate Steels for Heavy Section Applications", in Processing, Microstructure and Properties of HSLA Steels, ed. A.J. DeArdo, AIME, New York, N.Y., 1988, pp. 317-329.
23. H.K.D.H. Bhadeshia and D.V. Edmonds, "The Bainite Transformation in a Silicon Steel", Metallurgical Transactions A, Vol. 10A, 1979, pp. 895-907.
24. H.K.D.H. Bhadeshia and D.V. Edmonds, "Bainite in Silicon Steels: New Composition-Property Approach Part 1", Metal Science, Vol. 17, 1983, pp. 411-419.
25. H.K.D.H. Bhadeshia and D.V. Edmonds, "Bainite in Silicon Steels: New Composition-Property Approach Part 2", Metal Science, Vol. 17, 1983, pp. 420-425.
26. V.T.T. Miihkinen and D.V. Edmonds, "Microstructural Examination of Two Experimental High-Strength Bainitic Low-Alloy Steels Containing Silicon", Materials Science and Technology, Vol. 3, 1987, pp. 422-431.
27. H.I. Aaronson and H.J. Lee, "Another Visit to the Three Definitions of Bainite", Scripta Metallurgica, Vol. 21, 1987, pp. 1011-1016.
28. First Progress Report of Subcommittee XI: "Electron Microstructure of Steel", ASTM Committee E-4, Proceedings ASTM, Vol. 50, 1950, pp. 444-492.
29. Second Progress Report by Subcommittee XI of Committee E-4, "Electron Microstructure of Bainite in Steel", Proceedings ASTM, Vol. 52, 1952, pp. 543-572.
30. D.N. Shackleton and P.M. Kelly, "Morphology of Bainite", Physical Properties of Martensite and Bainite, ISI, Special Report 93, London, England, 1965, pp. 126-134.

31. J.M. Oblak and R.F. Hehemann, "Structure and Growth of Widmanstätten Ferrite and Bainite", Symposium: Transformation and Hardenability in Steels, Climax Molybdenum Co., Ann Arbor, Michigan, 1967, pp. 15-38.
32. H.K.D.H. Bhadeshia and A.R. Waugh, Proc. Int. Conf. on Solid-Solid Phase Transformations, ed. by Aaronson, Laughlin, Sekerka and Wayman, AIME, New York, N.Y., 1982, p. 1581.
33. G.R. Purdy and M. Hillert, "On the Nature of the Bainite Transformation in Steels", *Acta Metallurgica*, Vol. 32, No. 6, 1984, pp. 823-828.
34. R.H. Goodenow, R.H. Barkalow and R.F. Hehemann, "Bainite Transformations in Hypoeutectoid Steels", Physical Properties of Martensite and Bainite, ISI, Special Report 93, London, England, 1965, pp. 135-152.
35. L.J. Habraken and M. Economopoulos, "Bainitic Microstructures in Low-Carbon Alloy Steels and Their Mechanical Properties", Symposium: Transformation and Hardenability in Steels, Climax Molybdenum Co., Ann Arbor, Michigan, 1967, pp. 69-108.
36. S.W. Thompson and G. Krauss, "Structure and Properties of Continuously Cooled Bainitic Ferrite-Austenite-Martensite Microstructures", in Mechanical Working and Steel Processing Proceedings, 1989, pp. 467-481.
37. S.W. Thompson, D.J. Colvin, and G. Krauss, "Continuous Cooling Transformations and Microstructures in a Low-Carbon, High-Strength Low-Alloy Plate Steel", *Metallurgical Transactions A*, Vol. 21A, 1990, pp. 1493-1507.
38. B.S. Kloberdanz, ongoing M.S. thesis work, Colorado School of Mines, 1992.
39. J.H. Woodhead and J.A. Whiteman, "Structure and Properties of Apolygonal Ferrites", Processing and Properties of Low Carbon Steel, ed. J.M. Gray, AIME, New York, N.Y., 1973, pp. 145-162.
40. S.H. Avner, Introduction to Physical Metallurgy, Second Edition, McGraw-Hill, Singapore, 1985.

41. E.C. Bain and H.W. Paxton, Alloying Elements in Steels, Second Edition, Third Revised Printing, ASM, Metals Park, Ohio, 1966.
42. S.W. Poole and J.E. Franklin, "High-Strength Structural and High-Strength Low-Alloy Steels", Metals Handbook, 9th Edition, ASM, Metals Park, Ohio, pp. 403-420.
43. R.J. Jessman and G.J. Murphy, "Precipitation Hardening Alloy Steels Provides Strength and Low Temperature Toughness Under Severe Service Conditions", Industrial Heating, Sept. 1979, pp. 27-32.
44. George Krauss, Steels: Heat Treatment and Processing Principles, ASM, Metals Park, Ohio, 1989.
45. Y.J. Park and F.B. Fletcher, "Effects of Manganese, Chromium, and Molybdenum on the Isothermal Transformation of Austenite in Eutectoid Steels", J. Heat Treating, Vol. 4, No. 3, 1986, ASM, pp. 247-252.
46. P. Henry, The Importance of Molybdenum in Heavy Steel Sections, Climax Molybdenum Co., London, England, 1972.
47. J.C. Greenbank, "Carbon Solute Interactions in Fe-Cr-C, Fe-Mo-C and Fe-W-C Alloys", Journal of the Iron and Steel Institute, 1971, pp. 986-990.
48. H. Wada, "Thermodynamics of the Fe-Mo-C System at 985K", Metallurgical Transactions A, Vol. 17A, 1986, pp. 391-398.
49. J.A. Straatmann, D.V. Doane and Y.J. Park, "Using CCT Diagrams in the Development of Low-Carbon, High Strength Steels" in Processing, Microstructure and Properties of HSLA Steels, ed. A.J. DeArdo, AIME, New York, N.Y., 1988, pp. 287-303.
50. F.G. Berry and R.W.K. Honeycombe, "The Isothermal Decomposition of Austenite in Fe-Mo-C Alloys", Met. Trans., Vol. 1, 1970, pp. 3279-3286.
51. R.A. DePaul and A.L. Kitchin, "The Role of Nickel, Copper, and Columbium (Niobium) in Strengthening a Low-Carbon Ferritic Steel", Metallurgical Transactions, Vol. 1, Feb. 1970, pp. 389-393.

52. A.P. Coldren and G.T. Eldis, "Using CCT Diagrams to Optimize the Composition of an As-Rolled Dual-Phase Steel", *Journal of Metals*, March 1980, pp. 41-48.
53. M.R. Krishnadev and A. Galibois, "Development of a High-Strength Low-Alloy Steel Strengthened by Transformation Substructure and Precipitation of Copper and Niobium", *Metals Technology*, Vol. 1, 1974, pp. 300-302.
54. K.J. Irvine, *JISI*, 1962, p. 820.
55. R.L. Smith, G. Spangler and R.M. Brick, *Trans Am. Soc. Metals*, Vol. 46, 1954, p. 973.
56. E.O. Hall, *Proc. Phys. Soc.*, London, Vol. B64, 1951, p. 747.
57. N.J. Petch, *JISI*, Vol. 174, 1953, p. 25.
58. R.W.K. Honeycombe, Structure and Strength of Alloy Steels, Climax Molybdenum Co., London, England.
59. R.W. Armstrong, "The Influence of Polycrystal Grain Size on Several Mechanical Properties of Materials", *Met. Trans.*, Vol. 1, 1970, pp. 1169-1176.
60. K.J. Irvine, "A Comparison of the Bainite Transformation With Other Strengthening Mechanisms in High-Strength Structural Steels", Symposium: Steel-Strengthening Mechanisms, Climax Molybdenum Co., Greenwich, Connecticut, 1969, pp. 55-65.
61. J. Malcolm Gray, "Columbium (Niobium) as a Grain Refiner in Hot-Rolled Steels", Processing and Properties of Low Carbon Steel, ed. J.M. Gray, AIME, New York, N.Y., 1973, pp. 225-247.
62. M. Durbin and P.R. Krahe, "Controlled Rolling and the Properties of Very Low Carbon, High Manganese Steels Containing Strong Carbide Formers", in Processing and Properties of Low Carbon Steel, ed. by J.M. Gray, AIME, New York, N.Y., 1973, pp. 109-131.
63. I. Kozasu and T. Osuka, "Processing Conditions and Properties of Control-Rolled Steel Plates", in Processing and Properties of Low Carbon Steel, ed. by J.M. Gray, AIME, New York, N.Y., 1973, pp. 47-67.

64. Tomo Tanaka, N. Tabata, T. Hatomura and C. Shiga, "Three Stages of the Controlled-Rolling Process", Proceedings of an International Symposium on High-Strength, Low-Alloy Steels, Microalloying 75, Union Carbide Corp., New York, N.Y., 1977, pp. 107-119.
65. B. Karlsson, "Grain Refinement in Fe-C Alloys by Thermal Cycling", Materials Science and Engineering, Vol. 11, 1973, pp. 185-193.
66. J.Y. Koo and G. Thomas, "Thermal Cycling Treatments and Microstructures for Improved Properties of Fe-0.12% C-0.5% Mn Steels", Materials Science and Engineering, Vol. 24, 1976, pp. 187-198.
67. R.A. Grange, Trans. Am. Soc. Metals, Vol. 59, 1966, p. 26.
68. R.B. Nicholson, "Strong Microstructures From the Solid State" in Strengthening Mechanisms in Crystals ed. A. Kelly and R.B. Nicholson, John-Wiley and Sons, New York, N.Y., 1971, p. 542.
69. J.H. Bucher, "The Toughness of Ferritic Steel Strengthened by Precipitation of CbC", Trans. AIME, Vol. 239, 1967, pp. 478-485.
70. E.C. Hamre and A.M. Gilroy-Scott, "Properties of Acicular-Ferrite Steel for Large-Diameter Line Pipe", in Proceedings of an International Symposium on High-Strength, Low-Alloy Steels, Microalloying 75, Union Carbide Corp., New York, N.Y., 1977.
71. Paul E. Repas, "Control of Strength and Toughness in Hot-Rolled Low-Carbon Manganese-Molybdenum-Columbium-Vanadium Steels", Proceedings of an International Symposium on High-Strength, Low-Alloy Steels, Microalloying 75, Union Carbide Corp., New York, N.Y., 1977, pp. 387-398.
72. T. Gladman, D. Dulieu and Ian D. McIvor, "Structure-Property Relationships in High-Strength Microalloyed Steels", in Proceedings of an International Symposium on High-Strength, Low-Alloy Steels, Microalloying 75, Union Carbide Corp., New York, N.Y., 1977, pp. 32-58.

73. P.R. Kirkwood, "Welding of Niobium Containing Microalloyed Steels", in Niobium: Proceedings of the International Symposium, edited by Harry Stuart, AIME, New York, N.Y., 1981, pp. 761-802.
74. A.L. Mincher, R.A. Bosch and P.R. Morgan, "Commercial Production of High-Strength, Titanium-Bearing Hot-Rolled Steel", Proceedings of an International Symposium on High-Strength, Low-Alloy Steels, Microalloying 75, Union Carbide Corp., New York, N.Y., 1977, pp. 251-260.
75. B. Aronsson, "The Microstructure and Mechanical Properties of Some Martensitic Steels With Special Reference to the Influence of Niobium and Molybdenum", Symposium Steel-Strengthening Mechanisms, Climax Molybdenum Co., Greenwich, Connecticut, 1969, pp. 77-87.
76. S.W. Thompson and G. Krauss, "Precipitation and Fine Structure in Medium-Carbon Vanadium and Vanadium/Niobium Microalloyed Steels", Metallurgical Transactions A, Vol. 20A, 1989, pp. 2279-2288.
77. T. Tanaka, "Controlled Rolling of Steel Plate and Strip", International Metals Review, 1981, pp. 185-212.
78. N.P. Allen, "Effect of Solute Element on the Mechanical Properties of Iron", in Proceedings of a Conference on Iron and Its Dilute Solid Solutions, ed. C.W. Spencer and F.E. Werner, Interscience Publishers, New York, N.Y., 1963, pp. 271-314.
79. R.E. Read-Hill, Physical Metallurgy Principles, Van Nostrand Co., New York, N.Y., 1964.
80. George E. Dieter, Mechanical Metallurgy, Second Edition, McGraw-Hill, New York, N.Y., 1984.
81. P.M. Kelly and J. Nutting, "Strengthening Mechanisms in Martensite", Physical Properties of Martensite and Bainite, ISI, Special Report 93, London, England, 1965, pp. 166-170.
82. G.R. Speich and P.R. Swann, "Yield Strength and Transformation Substructure of Quenched Iron-Nickel Alloys", JISI, 1965, pp. 480-485.

83. G.E. Hicho, S. Singhal, L.C. Smith and R.J. Fields, "Effect of Thermal Processing Variations on the Mechanical Properties and Microstructure of a Precipitation Hardening HSLA Steel", Journal of Heat Treating, Vol. 3, No. 3, 1984, pp. 205-212.
84. W.E. Creswick, "Commercial Development of a Rimmed Low-Alloy Precipitation-Hardening High-Strength Steel", Strong Tough Structural Steels, ISI No. 104, London, England, 1967, pp. 86-92.
85. P. Morcinek, V. Smid, T. Heczko and T. Prenka, "Structural Steels With Acicular Ferrite", in Proceedings of an International Symposium on High-Strength, Low-Alloy Steels, Microalloying 75, Union Carbide Corp., New York, N.Y., 1977, pp. 272-278.
86. M. Cohen and W. S. Owen, "Thermo-mechanical Processing of Microalloyed Steels" Proceedings of an International Symposium on High-Strength, Low-Alloy Steels, Microalloying 75, Union Carbide Corp., New York, N.Y., 1977, pp. 2-8.
87. M.G. Gemmill, "The Merits and Demerits of Strengthening Mechanisms in Low-Alloy Steels", Symposium: Steel-Strengthening Mechanisms, Climax Molybdenum Co., Greenwich, Connecticut, 1969, pp. 67-75.
88. J.W. Christian, "Military Transformations: An Introductory Survey", Physical Properties of Martensite and Bainite, ISI, Special Report 93, London, England, 1965, pp. 1-25.
89. R.F. Hehemann, K.R. Kinsman and H.I. Aaronson, "A Debate on the Bainite Reaction", Metallurgical Transactions, Vol. 3, 1972, pp. 1077-1094.
90. K.R. Kinsman and H.I. Aaronson, "Influence of Molybdenum and Manganese on the Kinetics of the Proeutectoid Ferrite Reaction", Symposium: Transformation and Hardenability in Steels, Climax Molybdenum Co., Ann Arbor, Michigan, 1967, pp. 39-55.
91. H.I. Aaronson, "On the Problem of the Definitions and the Mechanisms of the Bainite Reaction", The Mechanism of Phase Transformations in Crystalline Solids, pp. 270-312.

92. J.W. Christian, "The Origin of Surface Relief Effects in Phase Transformations", in Decomposition of Austenite by Diffusional Processes, ed. by V.F. Zackay and H.I. Ironstone, InterScience Publishers, Easton, Pennsylvania 1962, pp. 371-386.
93. S.K. Liu and G.Y. Zhang, "The Effect of Mn and Si on the Morphology and Kinetics of the Bainite Transformation in Fe-C-Ti Alloys", Metallurgical Transactions A, Vol. 21A, June 1990, pp. 1509-1515.
94. H.K.D.H. Bhadeshia and J.W. Christian, "Bainite in Steels", Metallurgical Transactions A, Vol. 21A, April 1990, pp. 767-797.
95. S.W. Thompson, Private communication based on unpublished research.
96. H.K.D.H. Bhadeshia, Proc. Int. Conf. Solid-Solid Phase Transformations, ed. by G.W. Lorimer, Institute of
97. Bainite Committee of Iron & Steel Institute of Japan, Atlas for Bainitic Microstructures, Vol. 1, 1992. Metals, London, England, 1988, pp. 321-386.
98. G.R. Speich and H. Warlimont, "Yield Strength and Transformation Substructure of Low-Carbon Martensite", JISI, 1968, pp. 385-392.
99. W.D. Swanson and J. Gordon Parr, "Transformations in Iron-Nickel Alloys", JISI, Feb. 1964, pp. 104-106.
100. D.W. Gomersall and J. Gordon Parr, "Transformations in Iron-Manganese Alloys," JISI, March 1965, pp. 275-279.
101. J.M. Wallbridge and J. Gordon Parr, "Transformations in Fe-Cr Alloys," JISI, Feb. 1966, pp. 119-121.
102. J. Gordon Parr, "Transformations in Fe-Co Alloys," JISI, April 1967, pp. 426-427.
103. E.A. Wilson, "The $\gamma \rightarrow \alpha$ Transformation in Iron," Scripta Metallurgica, Vol. 12, 1978, pp. 961-968.
104. E.A. Wilson, "The $\gamma \rightarrow \alpha$ Transformation in Iron and its Dilute Alloys," Scripta Metallurgica, Vol. 4, 1970, pp. 309-312.

105. A. Gilbert and W.S. Owen, "Diffusionless Transformation in Iron-Nickel, Iron-Chromium and Iron-Silicon Alloys," *Acta Metallurgica*, Vol. 10, 1962, pp. 45-54.
106. A.R. Marder and G. Krauss, "The Morphology of Martensite in Iron-Carbon Alloys," *Trans. ASM*, Vol. 60, 1967, pp. 651-660.
107. R.H. Goodenow and R.F. Hehemann, "Transformations in Iron and Fe-9 Pct Ni Alloys," *Trans. AIME*, Vol. 233, 1965, pp. 1777-1786.
108. J.M. Marder and A.R. Marder, "The Morphology of Iron-Nickel Massive Martensite," *Trans. ASM*, Vol. 62, 1969, pp. 1-10.
109. G. Krauss and A.R. Marder, "The Morphology of Martensite in Iron Alloys," *Met. Trans.*, Vol. 2, 1971, pp. 2343-2357.
110. Lee Dillinger, "Hardness Testing", Met-Tips, Ideas for Metallographic Procedures, Leco Corp., Michigan, 1980.
111. George T. Eldis, "A Critical Review of Data Sources for Isothermal Transformation and Continuous Cooling Transformation Diagrams", in Proceedings of an International Conference on Solid-Solid Phase Transformations, TMS-AIME, Warrendale, PA, 1982, pp. 126-157.
112. "Standard Test Methods for Determining Average Grain Size", ASTM Designation: E112-88, Annual Book of ASTM Standards, Vol. 03.01, ASTM, Philadelphia, PA, 1989, pp. 284-306.
113. Halle Abrams, "Grain Size Measurement by Intercept Method", *Metallography*, Vol. 4, 1971, pp. 59-78.
114. "Standard Test Methods for Characterizing Duplex Grain Sizes", ASTM Designation: E1181-87, Annual Book of ASTM Standards, Vol. 03.01, ASTM, Philadelphia, PA, 1989, pp. 828-841.
115. E.E. Underwood, "Applications of Quantitative Metallography", Metals Handbook, ASM, Metals Park, Ohio, 1988, pp. 37-47.

116. Dynamic Thermal/Mechanical Metallurgy Using the Gleeble 1500, Second Edition, Duffers Scientific Inc., Troy, N.Y., 1986.
117. G.R. Speich and T.M. Scoonover, "Continuous-Cooling-Transformation Behavior and Strength of HSLA-80 (A710) Steel Plates", in Processing, Microstructure and Properties of HSLA Steels, ed. A.J. DeArdo, AIME, New York, N.Y., 1988, pp. 263-286.
118. W.W. Cias, Phase Transformation Kinetics and Hardenability of Medium-Carbon Alloy Steels, Climax Molybdenum Co., Greenwich, Connecticut.
119. "Standard Test Method for Vickers Hardness of Metallic Materials", ASTM Designation: E92-82, Annual Book of ASTM Standards, Vol. 03.01, ASTM, Philadelphia, PA, 1989, pp. 264-272.
120. "Standard Test Methods of Tension Testing of Metallic Materials", ASTM Designation: E8-89, Annual Book of ASTM Standards, Vol. 03.01, ASTM, Philadelphia, PA, 1989.
121. R.F. Johnson, "The Measurement of Yield Stress", Strong Tough Structural Steels, ISI Publication 104, London, England, 1967, pp. 51-60.
122. "Standard Test Methods for Notched Bar Impact Testing of Metallic Materials", ASTM Designation: E23-88, Annual Book of ASTM Standards, Vol. 03.01, ASTM, Philadelphia, PA, 1989.
123. K.W. Andrews, "Empirical Formulae for the Calculation of Some Transformation Temperatures", JISI, London, England, 1965, pp. 721-727.
124. W. Steven and A.G. Haynes, "The Temperature of Formation of Martensite and Bainite in Low-Alloy Steels: Some Effects of Chemical Composition", JISI, London, England, 1956, pp. 349-359.
125. V. Biss and R.L. Cryderman, "Martensite and Retained Austenite in Hot-Rolled, Low-Carbon Bainitic Steels", Metallurgical Transactions, Vol. 2, 1971, pp. 2267-2276.

126. A.P. Coldren and R.L. Cryderman, "Strength and Impact Properties of Low-Carbon Structural Steels Containing Molybdenum", in Symposium: Steel-Strengthening Mechanisms, Climax Molybdenum Co., Greenwich, Connecticut, May 1969, pp. 17-44.
127. Written discussion by B.V. Narashima Rao, following G. Krauss, "Martensitic Transformation, Structure and Properties in Hardenable Steels," proceedings Hardenability Concepts With Applications to Steel, editors D.V. Doane and J.S. Kirkaldy, AIME,
128. K.J. Irvine and F.B. Pickering, "The Impact Properties of Low Carbon Bainitic Steels", JISI, London, England, 1963, pp. 518-531.
129. C.I. Garcia, A.K. Lis, S.M. Pytel and A.J. DeArdo "Ultra -Low Carbon Bainitic Plate Steels: Processing, Microstructure and Properties" I&SM, October 1991, pp. 97-106.

Table 1. Charpy Impact Data

STEEL	TEMPERATURE (°C)	ENERGY ABSORBED (J)		TEMPERATURE (°C)	ENERGY ABSORBED (J)	
		1X			2X	
A	-195	1.35		-195	1.35	
A	-84	2.71		-84	2.71	
A	-40	10.84		-40	12.20	
A	-18	24.40		-20	27.11	
A	0	112.53		-18	108.46	
A	23	178.97		0	97.62	
A	61.7	184.39		23	146.43	
A	64.5	162.70		23	138.30	
A	103	166.76		61.7	173.54	
A	130	173.54		103	170.83	
A	156	166.76		130	178.97	
				156	185.75	
B	-195	1.35		-195	1.35	
B	-84	2.70		-84	4.07	
B	-40	6.78		-40	8.13	
B	-18	10.85		-18	47.45	
B	0	24.45		0	86.77	
B	23	74.57		23	84.06	
B	44.5	136.94		44.5	151.85	
B	62	128.80		62	161.34	
B	103	135.58		103	178.96	
B	130	154.56		130	183.03	
B	160	143.71		160	173.54	
C	-195	1.35		-195	1.35	
C	-84	2.70		-84	4.07	
C	-40	5.42		-40	4.07	
C	-18	10.85		-18	10.85	
C	0	10.85		0	47.45	
C	23	13.56		23	97.62	
C	44.5	71.86		61.7	111.17	
C	61.7	101.69		103	97.62	
C	103	128.80		103.5	123.38	
C	129	155.92		129	122.03	
C	159	135.61		159	142.36	
AT*	-197	2.70		-197	2.70	
AT	-83	7.46		-83	6.10	
AT	-40	42.03		-40	169.47	
AT	-19	62.37		-19	151.85	
AT	0	58.30		0	216.93	
AT	24	181.60		24	138.29	
AT	60	138.29		60	223.71	
AT	100	165.41		100	174.90	
AT	145	189.81		145	196.59	

Table 2. Tensile Data

STEEL	Mn%	YS(MPa)	UTS(MPa) 1X	EL(%)	YS(MPa)	UTS(MPa) 2X	EL(%)
A-1	1.00	712	899	13.35	703	857	13.25
A-2	1.00	707	874	15.21	711	898	17.40
A-3	1.00	688	941	15.35	703	845	16.24
B-1	1.45	723	916	14.72	702	872	14.62
B-2	1.45	731	885	14.37	731	862	15.94
B-3	1.45	716	896	16.40	714	876	16.23
C-1	1.76	751	927	13.10	752	899	12.47
C-2	-	-	-	-	719	848	14.61
C-3	-	-	-	-	694	826	17.40
AT-1	1.00	612	704	19.71	559	664	25.14
AT-2	1.00	625	725	19.80	618	708	23.60
AT-3	1.00	660	742	20.74	565	670	24.20

* TEMPERED