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**A STUDY OF THE METALLURGICAL PROPERTIES THAT  
ARE NECESSARY FOR SATISFACTORY BEARING  
PERFORMANCE AND THE DEVELOPMENT OF  
IMPROVED BEARING ALLOYS FOR  
SERVICE UP TO 1000 F**

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## FOREWORD

This report, entitled "A Study of the Metallurgical Properties That Are Necessary for Satisfactory Bearing Performance and the Development of Improved Bearing Alloys for Service up to 1000 F," was prepared by the Crucible Steel Company of America, Pittsburgh, Pennsylvania, under USAF Contract No. AF33(616)-3318, Supplemental Agreement No. 2(57-711). The contract was initiated under Project No. 7351, "Metallic Materials," Task No. 73512, "High Temperature Alloys." The administration of this contract was under the direction of the Materials Laboratory, Directorate of Laboratories, Wright Air Development Center, with Lt. G. St. Pierre and subsequently Lt. E. B. Bitzer acting as project engineer.

One of the authors, A. E. Nehrenberg, initially supervised this project.

The continued interest and technical guidance on the project of P. Payson, Assistant Director of Research and Development, Crucible Steel Company of America, are gratefully acknowledged. Thanks are due to V. K. Chandhok, Richard S. Cremisio, and James Zimmer, Staff Metallurgists, Central Research Laboratory, Crucible Steel Company, for their contributions to the project.

This report covers work conducted from April 1, 1957 to March 31, 1958.

## ABSTRACT

To develop a bearing steel for operating temperatures up to 1000 F, fifty-one experimental compositions were studied. This report describes the theoretical considerations which formed the basis of the formulation of these steels as well as the annealing cycle, austenitizing temperature survey, and the tempering survey. The results of the tempering studies are plotted as "master tempering curves."

From the study of the temper resistance, eight steels were selected for further evaluation tests. These tests consisted of (1) hot hardness determinations, (2) dimensional stability measurements, (3) compression tests at room temperature and elevated temperature, (4) oxidation resistance, and (5) hot oil corrosion resistance.

As a result of the foregoing tests, a steel of the following composition is recommended for use as a bearing steel for operating temperatures up to 1000 F:

<u>C</u>	<u>Cr</u>	<u>V</u>	<u>W</u>	<u>Mo</u>	<u>Co</u>
1.09	4.2	1.9	6.7	3.7	5.2

The steel should be heat-treated as follows: austenitize at 2225 F, oil quench, and temper for two consecutive 2-hour periods at 1050 F.

This work has not only developed a steel for elevated-temperature bearing applications but it has also clarified the effects of alloying elements on secondary hardness, hardness retention at elevated temperatures, and other properties desired of a high-temperature bearing steel.

## PUBLICATION REVIEW

This report has been reviewed and is approved.

FOR THE COMMANDER:



RICHARD R. KENNEDY  
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## I. INTRODUCTION

The steady rise in operating temperatures of present day jet engines and gas turbine power plants has created the demand for bearing materials with superior elevated-temperature physical and mechanical properties. It has been suggested that the bearing should have a minimum hardness of Rockwell "C" 56 to 58 at the operating temperature. The most commonly used bearing steel, SAE 52100, can no longer be considered because of its rapid loss of hardness and dimensional stability at operating temperatures above 400 F.

The use of hot-work steels and high-speed tool steels for elevated-temperature bearing applications has met with varying degrees of success. An evaluation of the mechanical properties of these steels, used or proposed for use in high-temperature bearing applications, was made during the first year's work of this project and was published as WADC TR 57-343. In that report, data of mechanical properties were presented for twenty-two commercial hot-work steels and high-speed tool steels as well as seven experimental steels. The properties investigated included resistance to tempering, hot hardness at 400, 600, 800, and 1000 F after 1000-hour exposures at the respective temperatures, dimensional stability after the above high-temperature runs, and yield strength in compression at the above temperatures. This work showed that none of the steels investigated meet the requirements of aircraft bearing applications above 900 F.

Because there is an urgent need for a bearing steel suitable for service up to 1000 F, the present work was devoted to the development of new high-temperature bearing steel containing a minimum amount of alloying elements.

## II. THEORETICAL CONSIDERATIONS

In order that the bearing may retain a minimum hardness of  $R_c$  56 to 58 at the operating temperature, the steel should be capable of developing very high as-heat-treated hardness. Furthermore, it should retain this hardness during long exposures at elevated temperatures.

In plain-carbon and low-alloy constructional steels, a hard martensitic structure can be produced by heat-treatment. However, when these steels are tempered at elevated temperatures they soften considerably; therefore, they cannot be used for high-temperature bearings. High-speed tool steels on the other hand are known for their hardness retention at elevated temperatures.

The retention of high hardness in alloy steels after tempering at high temperatures, appears to be the result of two concurrent phenomena:

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- (1) an increase in hardness attributable to "secondary hardening" in steels containing tungsten, molybdenum, or vanadium; and
- (2) a "retardation of tempering" in steels containing chromium or cobalt.

The increase in hardness between about 900 and 1050 F, or secondary hardness, has been attributed to the precipitation of complex carbides and the consequent age-hardening reactions that take place in the martensitic matrix of the steels when they are tempered at elevated temperatures.<sup>1-3</sup> A delay of carbide precipitation from the martensite is presumably responsible for the "retardation of tempering."

The alloying elements usually added to steels which are subject to secondary hardening include carbon and the transition elements vanadium, chromium, molybdenum, tungsten, and cobalt. With the exception of cobalt, these transition elements are strong carbide formers. An understanding of the secondary hardening mechanism should be a prerequisite to designing steel compositions for high-temperature applications. This involves an evaluation of the part played by these elements, both individually and collectively, during the tempering operation.

### Carbon

In steels which are to be hardened, carbon is the most important alloying element. An adequate carbon content permits the steel to be quenched to full martensitic hardness and provides sufficient carbon for the formation of the alloy carbides. The percentages of carbon which are required stoichiometrically for each per cent of the alloying elements to form their respective carbides are given on page 6. These factors were used to calculate the required carbon content in each experimental steel.

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1. K. Kuo, "Carbide Precipitation, Secondary Hardening, and Red Hardness of High-Speed Steel," J. Iron Steel Inst. (London) 174, 223 (1953).
  2. Walter Crafts and John Lamont, "Secondary Hardening of Tempered Martensitic Alloy Steel," Trans.AIME 180, 471 (1949).
  3. K. Kuo, "Investigations on Carbides in Alloyed Steel," Jernkontorets Ann. 141, No. 4, 206 (1957).

In a completely annealed high-speed steel, very little carbon is in solution in the matrix; nearly all of it combines with the alloying elements in the form of complex carbides.<sup>4,5</sup>

### Chromium

Throughout the development of high-speed steels, manufacturers have added about 4% chromium to the steel composition. The reason for this percentage addition has never been explained satisfactorily.

In tool steel, two chromium carbides occur; they are  $Cr_7C_3$  and  $Cr_{23}C_6$ .<sup>6-9</sup> Since  $Cr_{23}C_6$  has some solubility for other carbide-forming elements and iron,<sup>4</sup> it is called the  $M_{23}C_6$  type of carbide, where "M" denotes the metal atom in the carbide. This is the most stable chromium carbide and is usually observed in annealed steels or in steels which have been tempered extensively.

At the usual austenitizing temperatures of high-speed steels, nearly all of the chromium carbide goes into solution in austenite<sup>4</sup> and the total carbon from this carbide is therefore available to the martensite formed during the subsequent quenching of the steel. Chromium also helps to retard tempering of the steel, that is, it delays the precipitation of the carbides on tempering.<sup>1</sup> It does not seem that chromium imparts secondary hardness in steels; nevertheless, Crafts and Lamont<sup>2</sup> were able to calculate the hardness of tempered chromium steels on the basis of a  $Cr_7C_3$  carbide composition.

### Tungsten and Molybdenum

Tungsten and molybdenum behave similarly in these steels. They belong to the same group in the periodic table of elements

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4. Francis Kayser and Morris Cohen, "Carbides in High Speed Steel — Their Nature and Quality," *Metal Progr.* 61, No. 6, 79 (1952).
  5. D. J. Blickwede and Morris Cohen, "The Isolation of Carbides from High Speed Steel," *Trans. AIME* 185, 578 (1949).
  6. K. Kuo, "Carbides in Chromium, Molybdenum and Tungsten Steels," *J. Iron Steel Inst. (London)* 173, 363 (1953).
  7. Walter Crafts, "Carbides in Isothermally Transformed Chromium Steels," *Trans. AIME* 185, 957 (1949).
  8. J. R. Lane and N. J. Grant, "Carbide Reactions in High Temperature Alloys," *Trans. ASM* 43, 113 (1952).
  9. H. J. Goldschmidt, "The Structure of Carbides in Alloy Steels, Part I, General Survey," *J. Iron Steel Inst. (London)* 160, 345 (1948).

and their atomic weights differ by a factor of approximately two. Both tungsten and molybdenum form the same type of carbides in which the metallic elements are interchangeable. Several investigators report that  $MC$ ,  $M_2C$ ,  $M_6C$ , and  $M_{23}C_6$  occur in tungsten and/or molybdenum steels.<sup>1,3,6,9</sup> The metal-to-carbon ratio in the steel and the temperature seem to be the factors controlling the formation of these carbides.<sup>1,3</sup>

Kuo<sup>1,3</sup> has shown that  $M_2C$  is mainly responsible for secondary hardening in tungsten and/or molybdenum steels. This transition carbide is formed on tempering the quenched steels. Later it is transformed into  $M_6C$ , which is believed to be the most stable carbide. The usual compositions listed for  $M_6C$  are  $Fe_4(W, Mo)_2C$  to  $Fe_3(W, Mo)_3C$ . Although this is principally a W and/or Mo complex carbide, it has some solubility for other carbide-forming elements in the steel. In high-cobalt steels, some cobalt also appears to dissolve in this carbide.<sup>4</sup>

In the formulation of the compositions of the steels investigated in this program, the stoichiometric carbon requirement in each steel is calculated on the assumption that tungsten and molybdenum, respectively, are present only as  $W_2C$  and  $Mo_2C$  carbides. This is probably a reasonable assumption since  $M_2C$  is the desired carbide and since  $Fe_4M_2C$  is a reported composition of the complex  $M_6C$  stable carbide.

At the austenitizing temperatures of high-alloy tool steels, an appreciable portion of the tungsten and molybdenum carbides remains undissolved in the austenite<sup>4</sup> and, therefore, cannot participate in the age-hardening reactions which take place during tempering of the hardened steel.

#### Vanadium

Vanadium also produces secondary hardening in steels.<sup>2</sup> It is known to form a NaCl type of carbide, reported as  $VC$  or  $V_4C_3$ <sup>1,9,10</sup>; however, the carbide actually observed in steels has been found to have an intermediate composition. As suggested by Goldschmidt,<sup>9</sup> the probable reason for the difference in the reported compositions of the vanadium carbide is that  $VC$  and  $V_4C_3$  form solid solutions. In  $VC$ , 1% vanadium combines with 0.235% carbon while in  $V_4C_3$ , 1% of the element combines with 0.177% carbon. In the stoichiometric calculations of the steel compositions it is assumed that 0.2% carbon combines with each per cent of vanadium to form the carbide (page 6).

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10. Walter Crafts and John Lamont, "Carbides in Long-Tempered Vanadium Steels," Trans. AIME 188, 561 (1950).

The vanadium carbide has slight solubility for other carbide-forming elements in the steel and therefore is called MC carbide.<sup>4</sup> At the usual austenitizing temperatures only a part of this carbide is dissolved in the austenite. The portion that remains undissolved lends wear resistance to the hardened matrix; furthermore, the residual carbide permits the steel to be heated to high austenitizing temperatures without causing pronounced grain growth.

### Cobalt

Cobalt is not a strong carbide-forming element. The element, when present in tool steel, dissolves in the matrix substituting for iron atoms.

Cobalt retards the softening of the steel particularly in the early stages of tempering.<sup>1,3</sup> It also tends to decrease the amount of retained austenite in the steel.<sup>11</sup>

### Niobium

Niobium (columbium) is not usually added to high-speed tool steels. Niobium and vanadium occupy the same group in the periodic table; they form the same NaCl type of refractory carbides, the niobium carbide being reported as NbC or Nb<sub>4</sub>C<sub>3</sub>. The composition of the niobium carbide observed in steels has been found to be intermediate to NbC and Nb<sub>4</sub>C<sub>3</sub>, and Goldschmidt<sup>9</sup> and Binder<sup>12</sup> have proposed that these carbides have mutual solid solubilities.

At the usual austenitizing temperatures of high-speed tool steels, about one-half of the tungsten and molybdenum and an appreciable fraction of the vanadium contents exist as undissolved (residual) carbides in the austenite<sup>4</sup>; therefore, it seems possible to use smaller alloy additions and retain a larger fraction of these elements in solid solution in the austenite by adding refractory carbide-forming elements like niobium to the steel.

In the stoichiometric calculations of the experimental steel compositions containing niobium, a factor of 0.11% carbon is assumed to combine with 1% niobium to form the carbide; this is intermediate to the factors for NbC and Nb<sub>4</sub>C<sub>3</sub>. Niobium dissolved in the austenite may also impart secondary hardness to the steel.

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11. J. H. Holloman and L. D. Jaffe, Ferrous Metallurgical Design, John Wiley and Sons, Inc., New York, 1948, page 47.
  12. W. O. Binder, Columbium in Iron and Steel: Part III, Boron, Calcium, Columbium and Zirconium in Iron and Steel, John Wiley and Sons, Inc., New York, 1957, page 123.

In the formulation of the steel compositions investigated in this program, it was assumed that the optimum carbon content of the steel for maximum secondary hardening is the total of the carbon required by each element to form that carbide which produces secondary hardening. The factor used, namely, weight per cent carbon required for one per cent of the element, is shown in the following tabulation:

<u>Element</u>	<u>Carbide Composition</u>	<u>Factor</u> <u>Wt.% C combining with 1% Element</u>
Chromium	Cr <sub>7</sub> C <sub>3</sub>	0.099
	Cr <sub>23</sub> C <sub>6</sub>	0.060
Tungsten	W <sub>2</sub> C	0.033
Molybdenum	Mo <sub>2</sub> C	0.063
Vanadium	VC or V <sub>4</sub> C <sub>3</sub>	0.200
Niobium	NbC or Nb <sub>4</sub> C <sub>3</sub>	0.110

### III. EXPERIMENTAL WORK AND RESULTS

#### Material

Fifty-two experimental compositions were induction melted in three groups of steels. The steels are designated as WB1 through WB15 in the first group, WB16 through WB34 in the second group, and WB35 through WB52 in the third group.\* The steel compositions are shown in Table I. For the purpose of discussion the fifty-two compositions are divided into 15 sets as shown in Table I.

Steels of the first group were formulated with a constant chromium content of about 4.5%. Tungsten, molybdenum, and vanadium contents were varied and the carbon content was adjusted according to the stoichiometric requirements for the formation of the simple alloy carbides. In the first group, no carbon was allocated for chromium carbide formation. The most temper-resistant steel (WB11) of this group was chosen as the base composition for steels of the second group. Chromium, tungsten, and molybdenum contents were changed in addition to that of carbon; cobalt was added to some compositions. Two of the most promising steel compositions (at two carbon levels) of the second group were chosen as the base compositions of the third group of steels. As before, chromium, tungsten, and molybdenum contents were varied and in some steels the vanadium content in the base steel composition was replaced by equivalent amounts of niobium. The last two steels have high silicon additions in their composition.

\*Steel WB40 which contained titanium could not be forged, therefore it could not be included in the experimental work.

The formulation of these experimental steels is more fully described in Section IV.

### Melting and Forging

All steels were air-induction melted as 30-pound heats and forged at the Sanderson-Halcomb Works of the Crucible Steel Company of America. Prior to being forged, the ingots were stripped hot, and slowly cooled in vermiculite; they were then heated to 1900 to 2000 F, soaked for 1 to 2 hours, and clogged with intermediate reheating cycles to 3/4-inch-square bars. The hot workability of Steels WB1 through WB15 and WB35 was considered good, while that of the other steels ranged from fair to poor. However, hot workability ratings based on 30-pound ingots are not necessarily a measure of the forgeability of a production-sized ingot.

All bars were strain-relieved before shipment to Crucible's Central Research Laboratory.

### Annealing

The bars were subjected to the following spheroidizing anneal: (1) austenitized for two hours at 1600 F; (2) furnace-cooled to 1450 F; (3) cooled from 1450 to 1200 F at the rate of 25 degrees per hour; and (4) air-cooled from 1200 F to room temperature.

In the annealed condition, the steels had completely spheroidized structures except the high-niobium steels which showed large idiomorphic niobium carbides (or carbonitrides). Table I also lists the as-annealed hardness of the steels.

### Austenitizing Survey

The object of the austenitizing survey was to determine the highest austenitizing temperatures which would achieve maximum secondary hardening without undue grain growth or grain boundary melting. In this study, 5/8-inch-thick specimens of the annealed bars were used. The oxidized and decarburized surfaces were removed by grinding. Subsequently, the specimens were austenitized at several temperatures in the range 2200 F to 2400 F and then oil quenched.† The as-quenched hardness of each specimen was recorded and the hardness and microstructure were noted after a two-hour tempering treatment at 1050 F. Table II shows the results of this austenitizing survey.

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†Steel WB17 was also austenitized at 2150 F.

From an examination of hardness, microstructure, and grain size, the optimum austenitizing temperature was selected for each steel (Table II).†

### Tempering Survey

A second set of specimens of the steels was austenitized at the selected temperatures and oil quenched. Table III shows the as-quenched hardness and the hardness after cumulative tempering up to 32 hours at 950, 1000, 1050, 1100, and 1200 F.

From these tempering data, a master tempering curve was drawn for each steel (Figures 7 through 24). In these curves, Rockwell "C" hardness is plotted against a tempering parameter  $T(20 + \log t) \times 10^{-3}$ , where T is the absolute temperature in degrees Rankine and t the tempering time in hours.

From the master tempering curve one can determine the room-temperature hardness of a steel after a 1000-hour temper at 1000 F, i.e., at a tempering parameter of 33.6. From an examination of the master tempering curves and especially the hardness at a parameter of 33.6, the following steels were selected for further evaluation: WB18, WB21, WB25, WB39, WB41, WB44, WB45, and WB49. Steels WB51 and WB52 were subjected to limited tests for oxidation and oil corrosion resistance.

### Hot Hardness Measurements

The design of the hot-hardness tester used in this investigation represents a considerable improvement over that used during the first year of this contract. It is believed that the present equipment can reproduce hardness data within  $\pm 1$  Rockwell "C" point. The modifications of the equipment involve the following:

- (a) the removal of the liquid-metal seal,
- (b) a redesign of the cooling system, and
- (c) the construction of a new furnace which allows for specimen positioning without movement of the specimen in relation to the anvil.

A schematic drawing and a photograph of the apparatus are shown in Figures 1 and 2. Essentially, the apparatus consists of five main parts (the numbers in parentheses refer to the numbered legend of Figure 1):

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†It should be noted here that during the subsequent tempering survey some of the steel specimens showed incipient grain boundary melting; this is probably due to inhomogeneity in composition along the length of the bar. The recommended austenitizing temperatures, therefore, are 25 degrees below the temperatures listed in Table II.



- (a) a modified Wilson Rockwell Hardness Tester (1),
- (b) a Marshall Hot-Hardness Testing Furnace (8),
- (c) a cross slide for positioning the specimen (16),
- (d) a one-piece Gilmore integral diamond cone indenter (22), and
- (e) an atmosphere purifying apparatus.

The modifications to the Rockwell testing machine consist mainly of the installation of a furnace for heating the specimens and the cooling coils and jackets (2,5,7,14) to minimize the effect of heat conduction from the furnace to the tester; otherwise, the machine is used in the conventional manner. The furnace consists of a gastight, Inconel heating chamber (11) mounted within a wire-wound furnace (10). The specimens are 0.625 inch round and 0.625 inch high with parallel ground and polished ends. Three specimens were tested at one time, the temperature being controlled and indicated by thermocouples welded to the specimen surface. An electronic temperature controller was used in conjunction with an adjustable power transformer. The furnace is mounted on a cross slide (16) which allows for precise location of the specimen relative to the indenter. An external indicator (19) records the location of all prior indentations on the specimen surface, with respect to the indenter. The indenter (22) consists of a diamond tip ground to a standard C brale and attached to a 6-inch shank by means of a special high-temperature brazing alloy. The indenter is introduced into the furnace through a gastight, water-cooled lid (4) which allows for movement of the furnace. Fifty to sixty individual impressions can be made on each specimen. The furnace parts and specimen are protected from oxidation by purified dry nitrogen atmosphere.

Hardness determinations are made initially at room temperature to ensure seating of all machine components and to compare hardness data with those obtained on other hardness testers. The furnace is then brought uniformly to the desired temperature and held for ten to fifteen minutes prior to making the test indentations. The motorized loading mechanism of the Rockwell hardness tester ensures uniform load application, thus minimizing the effects of creep at temperature. The indenter is maintained at temperature by keeping it in contact with the specimen during the heating period. The room-temperature hardness is measured at the completion of each test.

Specimens, cut from the eight selected steels were heat-treated as indicated in Table IV. Hardness measurements were made at room temperature, 600, 800, and 1000 F, in the heat-treated condition and after exposures for 1000 hours at 600 and 800 F and for 500 hours at 1000 F. The results are shown in Tables IV through VII which also show the room-temperature hardness of specimens after the hot hardness test. Table VIII shows the room-temperature hardness of the steels before and after the elevated-temperature exposures.

## Dimensional Stability Measurements

Specimens used in this study measured 3/8 inch in diameter and  $4.00 \pm 0.001$  inches long, with ends ground to the contour of a 4-inch-diameter sphere. The spherical ends prevent errors in length measurements which might result from slight tilting of a square-end cylindrical specimen.

Oversized specimens were machined from annealed bars of the selected steels and were heat-treated as shown in Table IV. They were then finish ground to the specified dimensions and the ends were given a fine polish.

Precision length measurements were made on a Johansson Comparator, using a jig to ensure that the specimens were vertical at all times. The precision of the measurement was of the order of  $\pm 2.5$  microinches per inch.

The specimens were held at high temperatures in evacuated Vycor tubes which were then placed inside vertical tubes immersed in a salt bath. The open ends of the tubes were closed with asbestos wool and insulating brick to minimize temperature gradients along the specimen length.

Length measurements were made on specimens as-heat-treated and after exposures of 500 and 1000 hours at 600 and 800 F as well as 250 and 500 hours at 1000 F. The length changes of the specimens after elevated-temperature exposures are expressed in microinches per inch in Table IX.

## Compression Tests

Compression tests were made of the selected steels at room temperature, 600, 800, and 1000 F using a Riehle 160,000-pound capacity tensile machine and a fixture constructed for the test. A schematic diagram and photograph of the fixture, mounted on the Riehle machine, are shown in Figures 3 and 4.

The purpose of the fixture is to maintain uniform axial compression loading and to provide a means of continuously recording specimen strain vs. compressive load. The load train consists of the following:

- (a) a top loading plate with a spherical concave seat (1),
- (b) a hardened steel ball (2),
- (c) an upper compression rod with a spherical concave seat at the upper end and a plane surface at the other (3),
- (d) upper and lower cemented carbide pressure plates (4 and 6), and
- (e) a lower compression rod with a plane top surface and a broad base at the lower end.

The ball loading at the top prevents any nonaxial motion resulting from a possible shift of the machine crosshead, with increasing load. The carbide pressure plates forestall any tendency of the compression rod faces to deform plastically. The compression rods are of sufficient diameter and strength to prevent plastic deformation at the temperatures and stresses encountered.

To ensure true axial loading, the specimens must be located precisely at the center of the loading train. An auxiliary locating fixture was used which permitted quick and precise installation of the specimen. Figure 5, shows the locating fixture, with the specimen (1) placed in the small V-notch locating plates of the fixture (2), and held there by the hook and spring arrangement (4). The large V-notch locating plates (3) register on the outer diameter of the upper compression rod, and locate the specimen with its axis coincident with and parallel to the axis of the compression rods.

The three massive guide posts (8) and the upper and lower subpress plates (9) in Figures 3 and 4 ensure the axial travel of the compression rods. This overcomes the effect of a possible heterogeneous yielding of the specimen, which could result in nonaxial loading and columnar buckling of the specimen. The guide posts are bolted rigidly to the lower subpress plate and are bushed to slide freely through the upper subpress plate. The upper compression rod is press-fitted into the geometric center of the upper subpress plate.

The specimen is heated with a split-hinged furnace (12) which is shown in Figures 3 and 4 with one section removed. The hot junction of the control thermocouple rests near the center of the specimen.

The deformation of the specimen under load is recorded continuously through the motion of the upper compression rod in relation to the lower compression rod translated into the output of a linear differential transformer (10). During the elastic deformation of the specimen, the motion measured is the sum of the elastic deformation of the specimen and that of the load train components between the extensometer frame arms (11). However, only the specimen undergoes plastic deformation; the plastic portion of the deformation curve is therefore valid in determining yield strengths. The signal from the linear differential transformer is fed to the strain load recorder of the universal testing machine yielding a record of load versus deformation. The cylindrical specimen was 0.505 inch in diameter and 1 inch long. The temperature of the specimen was measured with thermocouples welded to the top and bottom of the cylindrical surface. Table X shows the 0.1% and 0.2% yield strengths in compression of the heat-treated specimens at room temperature, 600, 800, and 1000 F.

### Oxidation Resistance Studies

To compare the oxidation resistance of selected steels at 1000 F, heat-treated cylindrical specimens (0.510 inch in diameter and 1.5 inches long with surface-ground surfaces) were exposed in an air atmosphere muffle furnace for 1000 hours at 1000 F. The thin scale formed on the specimen surfaces was removed by solution in hot KOH and subsequent thorough cleaning of the specimens. Metal loss per unit area of specimen surface and depth of penetration of the scale were then calculated and are shown in Table XI.

### Corrosion Resistance Studies

Resistance to corrosion of selected steels in lubricating oils at 400 and 600 F was compared with those of SAE 52100 and M2 tool steel. Washer-type specimens with finish-ground surfaces (5/8-inch OD by 1/4-inch ID by 1/8-inch thickness) were used for the tests. Specimens heat-treated and finish ground were sealed in Pyrex capsules (Figure 6) under partial pressure of prepurified nitrogen and were exposed for 400 hours, at 400 F in MIL-7808-C oil (25 ml) and at 600 F in Octa Decyl Tri Decyl Silane (25 ml). The results are presented in Tables XII and XIII.

### Metallographic Examination

A set of the selected steel specimens was austenitized at 2225 F and oil quenched. For Steels WB25, WB41, WB44, and WB49 this is 25 degrees below the austenitizing temperature indicated in Table IV. After a double tempering treatment (2 hours each) at 1050 F, the specimens were polished and examined metallographically. Their microstructures are shown in Figures 26 through 33. The lower austenitizing temperature did not affect the tempered hardness of the steels; furthermore, a finer and more uniform grain size was obtained, but the amounts of residual carbides were somewhat increased. The microstructures show the distribution of the carbides.

Examination, by X-ray diffraction, of the structure of Steel WB49 after the heat-treatment above did not show any retained austenite.

## IV. DISCUSSION OF RESULTS

The master tempering curves for the experimental steels are shown in Figure 7 through 24.

In Steels WB1 through WB6 (Set 1), the vanadium content is held constant and the tungsten and molybdenum contents are increased simultaneously with the carbon content; tungsten and molybdenum are proportioned in a 1:2 ratio of  $W_2C$  to  $Mo_2C$ . Examination of the master tempering curves (Figures 7 and 8) for these steels shows

that with increasing amount of  $M_2C$  (and carbon), the secondary hardness peak is raised to higher hardness levels and the hardness is retained to higher tempering parameters, i.e., longer times at the same temperature. Furthermore, to attain a secondary hardness peak of  $R_C$  67 to 68, a minimum carbon content of 0.85% is indicated.

In steels of Set 2 (WB7, WB8, and WB9), only the vanadium and carbon contents are changed progressively. An examination of the master tempering curves (Figure 9) shows that increasing the vanadium content prevents the abrupt drop in hardness at high tempering parameters which had been observed in steels of Set 1. Vanadium in excess of about 2%, however, does not seem to produce any additional secondary hardening effect. With increasing vanadium contents, greater quantities of residual vanadium carbide are left undissolved in the matrix, and obviously the temper resistance of steels is not enhanced by increasing the amount of residual carbides. The same effect is also shown by the curves of Steels WB10 through WB15 (Figures 10 and 11).

The curves of Figure 9 further suggest that 3% tungsten and 3% molybdenum are insufficient to produce the maximum secondary hardness that can be attained with these elements.

In Figure 10 the master tempering curves are plotted for steels of Set 3 (WB10, WB11, and WB12). Here the VC to  $M_2C$  ratio is changed at a higher tungsten and molybdenum level than in steels of the previous set. The curves show that an increase in VC or VC to  $M_2C$  ratio does not correspondingly increase secondary hardness. In fact, Steel WB11 with 1.8% vanadium retains a higher hardness level at a higher tempering parameter than Steel WB10 which has less vanadium or Steel WB12 which has more vanadium. The secondary hardness peak observed for steels of Set 3 is therefore higher than that for steels of Set 2 because of the higher tungsten and molybdenum contents rather than the increased vanadium contents.

In Set 4 (Steels WB13, WB14, and WB15 of Figure 11), a 1:1 instead of a 1:2 atomic ratio of tungsten to molybdenum is maintained while vanadium and carbon contents are progressively changed. A comparison should therefore be made between the master tempering curves of the following pairs of steels: (a) WB7 with WB13; (b) WB8 with WB14; and (c) WB9 with WB15. Comparison (a) indicates that a 1:1 atomic ratio of tungsten to molybdenum maintains a higher hardness in the steel than when the ratio is 1:2. This observation is not confirmed by comparing pairs (b) and (c) because the tempering curves correspond within the accuracy of the experimental determinations. Additional work is needed to arrive at firmer conclusions in this matter.

As explained in Section II, carbon in the form of  $M_{23}C_6$  chromium carbide is almost completely available for martensite formation on quenching the steel from the usual austenitizing temperature. Furthermore, chromium has not been reported to have

a secondary hardness peak of its own. It was, therefore, thought that carbon allocation for chromium carbide formation was not necessary for high secondary hardness; the first fifteen steels contain no additional carbon to satisfy the stoichiometric formation of chromium carbide.

The composition of Steel WB11 appeared to be the best among the steels of Sets 1 through 4. In Set 5, Steel WB16 is a duplicate of WB11 and Steels WB17, WB18, and WB19 were formulated to check the validity of the foregoing hypothesis. In this instance, additional carbon was allotted in Steel WB17 for  $\text{Cr}_7\text{C}_3$  formation and in Steel WB18 for  $\text{Cr}_{23}\text{C}_6$  formation. Steel WB19 contained no chromium.

The master tempering curves for these steels are shown in Figure 12. When carbon was allocated for  $\text{Cr}_7\text{C}_3$  formation, the total carbon content rose to 1.26% causing a large amount of austenite to be retained in the quenched steel. This steel shows considerable scatter of points to the left of the hardness peak, presumably due to very high austenite retention, therefore, this portion of the master tempering curve is not shown. At all tempering parameters above about 30, Steel WB18 (allowing carbon for the formation of  $\text{Cr}_{23}\text{C}_6$ ) has higher hardness than WB16 (with no carbon allocated to chromium) or WB19, a chromium-free composition.

A carbon addition to account for chromium carbide formation, therefore, develops and maintains very high hardness in these steels; the formation of stable austenite, however, should be avoided.

In Figure 13, Steels WB20, WB21, and WB22 are compared. Steel WB20 has the same base composition as Steel WB16, except that the chromium is lowered from 4.6% to 2.4%. Neither steel has carbon allocated for chromium carbide formation; the master tempering curve for Steel WB16 (4.5% Cr) is at a higher hardness level than that of Steel WB20 (2.5% Cr). In Steel WB21, carbon was allocated for  $\text{Cr}_7\text{C}_3$  formation at an alloy level of 2.5% chromium. Steels WB18 and WB21 have essentially identical compositions except for the chromium content. The master tempering curves for these two steels are almost identical up to a parameter of 33, beyond which Steel WB21 (2.4% Cr) maintains higher hardness. Steels WB21 and WB22 have the same composition except for the carbon content which is higher in WB21. The latter has a higher hardness at all parameter values above 30.

The study of the compositions of Steels WB16 through WB22 thus shows that to develop and maintain very high as-tempered hardness in these steels, (a) chromium is an essential alloying constituent (2.5% to 4.5%), and (b) the total carbon content must be between 1.0 and 1.1%.

In Steels WB23 and WB24 (Set 7), 5 and 10% cobalt, respectively, were added to the base composition of Steel WB20. The master tempering curves of these two steels (Figure 14) are almost identical.

A comparison of the curves for Steels WB23 and WB20 shows that 5% cobalt aids in maintaining a high level of hardness at all tempering parameters.

In Steels WB25 and WB26 of Set 8 (Figure 15), 5 and 10% cobalt, respectively, were added to the base composition of Steel WB16. A comparison of the curves for Steels WB25 and WB23, and those of Steels WB26 and WB24, shows that the steels with 4.5% chromium (WB25 and WB26) at a 0.85% carbon level maintain higher hardness up to a parameter of about 33.75. Comparing the tempering curves of WB25 and WB16 (base composition) again shows that cobalt addition helps to maintain higher hardness at all tempering parameters.

In the initial stages of tempering (up to about 1000 F for 4 hours) steels containing 10% cobalt show slightly higher hardness than those containing 5%. This can probably be explained by the observation of less retained austenite in steels with higher cobalt content. However, with respect to hardness retention, a 10% cobalt addition shows no advantage over a 5% cobalt addition. Furthermore, high cobalt contents lowered the grain coarsening temperature of the steel (Table II). The optimum cobalt content in the bearing steels therefore appears to be 5%.

Set 9 (Steels WB27 through WB34) was produced to study the effects on secondary hardness, at three carbon levels, of tungsten with no molybdenum, or molybdenum with no tungsten, or a combination of the two elements when they are present in a 1:2 atomic ratio; the other carbide-forming elements are kept constant. In this set, the carbon contents of these compositions for WB28 and WB31 did not meet the desired stoichiometric balance with the alloying elements; therefore a comparison of the steels in the intended manner is not possible. Nevertheless a comparison may be made among the following groups: (a) WB16 and WB27, (b) WB28, WB29, and WB30, and (c) WB32, WB33, and WB34. A study of the master tempering curves of these steels (Figures 16, 17, and 18) indicates that resistance to tempering is not significantly affected by substituting tungsten for molybdenum or vice versa. However, steels containing molybdenum and no tungsten show a secondary hardness peak earlier than the other steel compositions and steels containing tungsten and no molybdenum maintain a slightly higher hardness at high tempering parameters. As shown in Table II, the exclusive use of molybdenum lowers the austenitizing temperature.

A comparison of Steels WB16, WB29, and WB32 shows that the secondary hardness peak increases progressively with increasing carbon content and alloying elements. However, Steel WB32 is not superior to WB18 which contains a smaller percentage of alloying elements. Equivalent secondary hardness was thus achieved by the correct balance of carbon and smaller additions of alloying elements rather than by an increase in the amount of one alloying element alone.

The study of Steels WB17, WB18 and WB19 gave ambiguous results regarding the contribution made by chromium in the development and maintenance of a high level of secondary hardness. To re-evaluate the significance of chromium, Steels WB35 through WB38 (Set 10) were made at a carbon level of 1.1%. In Steel WB35 chromium was not added intentionally while the other steels contain increasing amounts of chromium up to 6.7%. A comparison of the master tempering curves of these steels (Figure 19) reveals the following:

(1) Increasing the chromium content increases the amount of retained austenite in the steel. The as-quenched hardness of these steels decreased from  $R_C$  68 in Steel WB35 to  $R_C$  47 in Steel WB38. The latter contained about 80 to 90% retained austenite. On the other hand, no retained austenite could be detected under the microscope in Steel WB35.

(2) At all tempering parameters, Steel WB36 (2.5% Cr) showed a higher hardness than either Steel WB35 (no chromium) or Steel WB38 (6.7% Cr). Steel WB35 is not capable of developing high secondary hardness, as compared with the other steels; furthermore, it does not retain this hardness at high tempering parameters. Both Steels WB36 and WB37 develop very high secondary hardness, but the former (2.5% Cr) shows a more gradual drop in hardness at parameters above 34.5. The curve of Steel WB37 maintains a hardness of 1 Rockwell "C" point above that of WB36 between parameters of 30.5 and 34.5; however, this was not observed in the same steel compositions of the previous group. (See Steels WB18 and WB21.)

(3) Steel WB38 has a secondary hardness peak of  $R_C$  67. The peak is quite sharp and the master tempering curve declines rapidly on both sides of this peak. At parameters to the left of the hardness peak, the curve shows a considerable scatter of points, as in Steel WB17 (Figure 12). Since Steel WB38 shows excessive austenite retention in the as-quenched condition, it is not suited for high-temperature bearing applications.

In Steel WB39 of Set 11 (Steels WB39, WB41, WB42, and WB43 in Figure 20), 5% cobalt was added to the composition of Steel WB37. A comparison of the master tempering curves for these two steels (Figures 19 and 20) shows that cobalt addition flattens the curve at higher hardness levels in the early stages of tempering. Cobalt furthermore appears to reduce the amount of retained austenite.

In Steels WB41, WB42, and WB43, niobium replaces approximately equivalent amounts of vanadium in Steel WB39. Up to a tempering parameter of about 32.5 the master tempering curves of Steels WB39 and WB41 are identical, but at higher parameters, Steel WB39 retains higher hardness. In the master tempering curves of Steels WB42 and WB43, there is a considerable scatter of points between parameters



of 29.5 and 34, the range in which the secondary hardness peaks occur due to the precipitation of tungsten, molybdenum, and vanadium carbides. Hardness values obtained by tempering at 950, 1000, and 1200 F fall upon a smooth curve; however, values obtained by tempering at 1050 and 1100 F follow an erratic pattern and produce points of lower hardness than would correspond to the original curve. This erratic tempering behavior of niobium bearing steels is not quite understood.

The compositions of Steels WB39 and WB44 are essentially the same except that WB44 has 2.7% chromium while WB39 contains 4.6% chromium. Between tempering parameters of 29 and 33, the master tempering curves for the two steels (Figures 20 and 21) are identical; however, outside these limits the lower chromium steel (WB44) shows higher hardness. Again this furnishes proof that a lower chromium level can impart greater temper resistance at very high tempering parameters.

It is interesting to note that in Steel WB39, which contains 4.6% chromium, there is enough carbon for the formation of  $\text{Cr}_{23}\text{C}_6$ ; on the other hand, in Steel WB44, with the lower chromium content (2.7%), the carbon satisfies the stoichiometric requirement for the formation of  $\text{Cr}_7\text{C}_3$ . It may be recalled that in Steel WB17 (4.5% Cr) the carbon content satisfied the stoichiometric requirement for  $\text{Cr}_7\text{C}_3$  formation and that on quenching, this steel contained excessive amounts of retained austenite.

In Set 12 (WB44 and WB45) and Set 13 (WB46, WB47, and WB48), niobium replaces equivalent vanadium contents in the respective base compositions of WB44 and WB46 at two carbon and chromium levels. Steels WB44 and WB45 have almost identical master tempering curves up to a parameter of about 31.5 (Figure 21); at higher parameters, Steel WB45, containing niobium, has lower hardness. A study of the master tempering curves of these steels confirms the previous observation that the higher carbon and lower chromium contents promote hardness retention to higher tempering parameters. It further shows that niobium addition at a lower carbon level causes the steel to lose hardness rapidly with increasing tempering parameters.

In contrast to the master tempering curves for Steels WB42 and WB43, there is relatively little scatter of points in the curves for Steels WB45, WB47, and WB48, which also contain niobium.

Steels WB49 and WB50 constitute Set 14. In all preceding steels of this group (Sets 10 through 12) a 1:2 atomic ratio of tungsten to molybdenum had been maintained. However, in Steel WB49 this ratio is 1:1 and in Steel WB50 it is 2:3. A comparison of the curves in Figure 23 shows that the steel with a 1:1 atomic ratio of tungsten to molybdenum maintains higher hardness at higher tempering parameters than the steel in which this ratio is 2:3.

Comparison of curves for Steel WB49 (W:Mo = 1:1) and Steel WB39 (W:Mo = 1:2) shows that the hardness of the two steels is almost identical up to a tempering parameter of about 34. Beyond this parameter, the higher tungsten alloy maintains a higher hardness.

Steel WB44 (W:Mo = 1:2 and 2.7% Cr) and Steel WB49 (W:Mo = 1:1 and 4.2% Cr) have identical master tempering curves, except in the very early stages of tempering.

In Set 15, Steels WB51 and WB52 have the same composition as Steel WB46 except for the addition of 0.7% and 1.5% silicon, respectively. The curve for Steel WB46 (Figure 22) has a higher hardness at all tempering parameters than either Steel WB51 or WB52.

The addition of silicon does not promote a high secondary hardness level as compared with some of the better steels in this group. The effect of silicon on the resistance to oxidized oil corrosion will be described later (page 19).

As previously mentioned, eight steels, namely, WB18, WB21, WB25, WB39, WB41, WB44, WB45, and WB49, were selected for further evaluation.

A study of the hardness at room temperature and the hot hardness at 600, 800, and 1000 F of the selected steels after different thermal treatments shows the following:

- (a) The steels have excellent hardness retention at elevated temperatures.
- (b) Within the limits of reproducibility of hardness measurements, exposures up to 1000 hours at 600 and 800 F do not lower the hardness of the steels. A 500-hour exposure of the steel at 1000 F reduces the hardness by approximately 3 Rockwell "C" points.
- (c) The hardness drop from room temperature to 600, 800, or 1000 F is nearly equivalent for the different steels. The average hardness drop from the room-temperature value is 4.1 Rockwell "C" points to 600 F, 6.6 Rockwell "C" points to 800 F, and 9.6 Rockwell "C" points to 1000 F (Figure 25). Therefore, in these steels, the hot hardness (up to 1000 F) can be predicted from the room-temperature hardness.
- (d) Upon cooling the steel specimens to room temperature after the hot hardness test at 1000 F, the original room-temperature hardness of the steel is restored. This shows that hardness changes which are due entirely to an increase in testing temperature (between room temperature and 1000 F) are completely reversible.

Of the eight steels, WB44 and WB49 retain the highest hardness at 1000 F after a 500-hour exposure at 1000 F.

In this work "dimensional stability" refers to the contraction or expansion of the quenched and tempered specimen, resulting from prolonged exposures at elevated temperatures (Table IX). The decomposition of retained austenite causes an expansion while tempering of martensite (carbide precipitation) causes a contraction of the steel. High-temperature bearing components should not expand or contract beyond specified limits during service at the operating temperatures.

Steel WB39 showed the largest proportion of retained austenite in the microstructure as well as the maximum expansion at 1000 F. Steels WB18 and WB21, which are cobalt-free compositions showed large positive length changes. This indicates that cobalt additions promote dimensional stability in the steels. Steels WB44 and WB49 are comparatively stable at 1000 F up to 500 hours of exposure.

All steels tested have high compressive yield strengths (Table X). The 0.2% yield strengths of Steels WB44 and WB49 which have the highest hot hardness at 1000 F decreases from about 500,000 psi at room temperature to about 340,000 psi at 1000 F. According to opinions expressed by bearing manufacturers, these strength levels are adequate for high-temperature bearing steels.

Table XI compares the resistance to oxidation of the selected steels. Two high-silicon steels were included in this test. The test results show that decreasing the chromium content of steel from 4.5% to 2.5% (WB21 and WB44) decreases the oxidation resistance. Similarly, increasing the silicon content increases oxidation resistance. A comparison of WB44 and WB49 shows the latter to have somewhat better oxidation resistance.

In Tables XII and XIII, the resistance of the selected steels to corrosion in lubricating oils at 400 F and 600 F is compared with that of SAE 52100 and M2 high-speed steels. At 400 F, SAE 52100 shows the least corrosion resistance. At 600 F in MIL-7808-C oil the oil corrosion resistance of the steels developed in this program is comparable to that of M2 steel. At 600 F, Steel WB49 shows no weight loss during the oil corrosion test and is therefore superior to Steel WB44 or M2.

Microstructures of the heat-treated experimental steels show residual carbides and tempered martensite. A uniform distribution of the carbides in a tempered martensitic matrix will not only increase the wear resistance of the steel, but should improve its fatigue properties. Such a distribution of the carbides is more easily achieved in production-sized ingots which had been subjected to severe hot reductions than in 30-pound experimental heats.

## V. CONCLUSIONS

From the results of this investigation for the development of a bearing steel capable of operating between room temperature and 1000 F, the following conclusions were drawn.

The steel should be capable of developing very high as-heat-treated hardness and should retain this high hardness at the operating temperatures. To meet these requirements, the steel must have a minimum carbon content of 1.0 to 1.1%. In this work it was assumed that the highest hardness is developed and that hardness retention at high temperatures is optimum when the carbon content in the steel satisfies the stoichiometric requirements for the formation of those alloy carbides which have been known to produce secondary hardness. The results indicate this is probably a valid assumption.

Chromium does not show a secondary hardness peak of its own, however, in the experimental steels chromium is essential to develop and maintain high secondary hardness. The data indicate that the optimum chromium content is between 2.5 and 4.5%. The results of this work also show that to attain and maintain high hardness, carbon should be allocated for the formation of  $M_{23}C_6$ .

The addition of 5% cobalt improves the temper resistance at high hardness levels. Increasing the cobalt content from 5 to 10% contributes little, if any, additional temper resistance.

Increasing tungsten and molybdenum, the  $M_2C$ -forming elements in the steel, to a certain optimum limit raises the secondary hardness peak as well as hardness retention. Varying the proportion of tungsten and molybdenum in  $(W,Mo)_2C$  appears to produce the following effects: When approximately equal volumes of the  $M_2C$  carbides are present at equal carbon and other alloy contents, molybdenum or tungsten alone produces master tempering curves that do not differ significantly from each other. Steels containing only molybdenum tend to produce a hardness peak at lower tempering parameters; tungsten alone, on the other hand, tends to increase temper resistance at high tempering parameters. A combination of the two elements in an atomic ratio of 1:1 (WB13 vs. WB7; and WB49 vs. WB39 and WB50) produces better resistance to tempering at high parameters than other ratios of tungsten to molybdenum or the use of these elements alone.

Vanadium increases the temper resistance of these steels. The optimum vanadium content appears to be about 2%; a larger proportion of this element does not impart greater hardness retention.

Niobium failed to raise the secondary hardness properties of any composition studied in this program, and it adversely affected resistance to tempering when added in amounts ranging from 1.0 to 2.8%. The addition of about 2.5 to 3% niobium produces idiomorphic

carbides (or carbonitrides) in the steel which cannot take part in the secondary hardness reactions.

Silicon up to 1.5% decreases secondary hardness in this type of steel.

The results of this work also suggest that a proper alloy balance with carbon is essential for maximum hardness retention with the most economical utilization of alloying elements.

When the experimental steels are heated from room temperature to 600, 800, and 1000 F, the average hardness is lowered by 4.1, 6.6, and 9.6 Rockwell "C" points, respectively. The elevated-temperature hardness of these steels could, therefore, be predicted from the room-temperature hardness. The hardness change in these steels with change in temperature (from room temperature to 1000 F) is completely reversible. The data also indicate that the steels suffer no measurable drop in hardness by exposures up to 1000 hours at 600 and 800 F. A 500-hour exposure at 1000 F lowers the hardness by about 3 Rockwell "C" points.

The presence of retained austenite in heat-treated specimens of the experimental steels adversely affects dimensional stability during prolonged high-temperature exposure. The data further indicate that cobalt improves dimensional stability.

Steels with high hardness also showed high compressive yield strength; however insufficient data are available to correlate the compressive yield strength with any other property of the steel.

The steels tested in this program possess fair resistance to oxidation at 1000 F and good resistance to oil corrosion at 400 and 600 F. At 400 F the corrosion resistance of all experimental steels in MIL-7808-C oil is superior to that of SAE 52100. At 600 F in Octa Decyl Tri Decyl Silane the corrosion resistance of the steel is about equivalent to that of M2 steel.

The oxidation resistance and oxidized oil corrosion resistance appear to be increased by increasing the chromium content from 2.5 to 4.5% or the silicon content up to 1.5%. However, an increase in the silicon content lowers hardness retention at elevated temperatures. Even though Steels WB44 and WB49 have comparable elevated-temperature hardness retention, dimensional stability and high-temperature compressive strengths, Steel WB49 is preferable as a bearing steel on the basis of increased resistance to oxidation and oxidized oil corrosion.

## VI. RECOMMENDATIONS AND FUTURE WORK

Steel WB49, with the following composition, is recommended as a bearing steel for use up to 1000 F:

	<u>C</u>	<u>Cr</u>	<u>V</u>	<u>W</u>	<u>Mo</u>	<u>Co</u>
WB49	1.09	4.2	1.9	6.7	3.7	5.2
Recommended Melting Range	1.0/1.1	4.0/4.5	1.8/2.1	6.5/7.0	3.5/4.0	5.0/5.5

The heat-treatment of this steel should include austenitizing at 2225 F and tempering for two 2-hour periods at 1050 F. This treatment reduces the amount of retained austenite to a low level.

A 600-pound melt of this steel will be vacuum induction melted and cast at the Vacuum Metals Division of the Crucible Steel Company, Syracuse, New York. The ingot will be forged to a 4-1/2-inch-square billet, hot rolled to 13/22-inch bars, and annealed.

SKF Industries, Philadelphia, Pennsylvania, has consented to manufacture 1/2-inch balls from the hot-rolled and annealed bar stock. These balls will be fatigue-tested by SKF and several other companies interested in high-temperature ball bearings.

The corrosion resistance of this steel, however, may not be adequate under anticipated conditions of engine operation. Type 440 C stainless has been considered for high-temperature bearings, but it has inadequate hardness retention on tempering at about 1000 F. The current development program, under this contract, will be devoted to the development and testing of modifications of Type 440 C steel. The aim is to develop a bearing steel with good oxidation and corrosion resistance as well as secondary hardening characteristics for use between room temperature and 1000 F.

Table I

Chemical Composition and Annealed Hardness  
of Experimental Steels<sup>a</sup>

Set No.	Steel Designation	C	Cr	V	W	Mo	Co	Nb	Ta	Rockwell "C" Hardness After Annealing
1	WB1	0.45	4.29	0.50	3.62	3.67	--	--	--	13
	WB2	0.55	4.29	0.51	4.81	4.77	--	--	--	14
	WB3	0.65	4.60	0.59	5.80	5.34	--	--	--	17
	WB4	0.70	4.93	0.54	6.59	6.57	--	--	--	18
	WB5	0.82	4.24	0.48	7.66	7.78	--	--	--	23
	WB6	0.97	4.61	0.54	8.76	9.00	--	--	--	26
2	WB7	0.53	4.56	1.02	3.10	3.11	--	--	--	14
	WB8	0.74	4.59	1.85	3.21	3.07	--	--	--	16
	WB9	0.87	4.69	2.94	3.11	3.07	--	--	--	16
3	WB10	0.67	4.60	1.10	4.94	4.58	--	--	--	15
	WB11	0.86	4.79	1.79	5.02	4.89	--	--	--	18
	WB12	1.02	4.68	2.91	4.95	4.71	--	--	--	19
4	WB13	0.52	4.48	0.94	5.43	2.67	--	--	--	13
	WB14	0.75	4.47	1.99	5.24	2.51	--	--	--	16
	WB15	0.89	4.85	2.96	5.23	2.64	--	--	--	18
5	WB16	0.85	4.54	1.99	4.75	4.46	--	--	--	16
	WB17	1.26	4.12	1.87	4.72	4.46	--	--	--	22
	WB18	1.07	4.61	2.11	4.75	4.49	--	--	--	19
	WB19	0.79	0.04	1.99	4.75	4.43	--	--	--	11
6	WB20	0.84	2.40	2.07	4.72	4.43	--	--	--	12
	WB21	1.08	2.35	1.90	4.72	4.46	--	--	--	19
	WB22	0.96	2.55	2.08	4.72	4.46	--	--	--	18
7	WB23	0.83	2.57	1.90	4.75	4.46	4.99	--	--	17
	WB24	0.84	2.57	1.80	4.80	4.42	9.65	--	--	19
8	WB25	0.88	4.67	1.90	4.77	4.46	5.11	--	--	20
	WB26	0.89	4.54	1.80	4.62	4.49	10.1	--	--	22

<sup>a</sup>All steels contained about 0.20 to 0.40 Mn and about 0.20 to 0.40 Si and the usual residual amounts of P, S, and Ni except Steels WB51 and WB52 which contained 0.67 and 1.51 Si, respectively.

Table I (continued)

Chemical Composition and Annealed Hardness  
of Experimental Steels<sup>a</sup>

Set No.	Steel Designation	C	Cr	V	W	Mo	Co	Nb	Ta	Rockwell "C" Hardness After Annealing
9	WB27	0.84	4.16	1.78	--	6.99	--	--	--	15
	WB28	0.93	4.53	1.99	14.95	0.12	--	--	--	19
	WB29	0.93	4.44	1.85	5.77	5.65	--	--	--	19
	WB30	0.91	4.28	1.78	--	8.45	--	--	--	16
	WB31	1.01	4.33	1.91	17.08	0.18	--	--	--	21
	WB32	1.09	4.81	1.89	6.80	6.76	--	--	--	20
	WB33	1.03	4.60	1.89	--	10.20	--	--	--	19
	WB34	1.05	4.87	2.02	20.1	0.25	--	--	--	24
10	WB35	1.01	0.07	1.91	4.51	4.55	--	--	--	18
	WB36	1.10	2.54	1.97	4.62	4.62	--	--	--	18
	WB37	1.11	4.74	1.91	4.67	4.65	--	--	--	21
	WB38	1.08	6.73	1.96	4.67	4.62	--	--	--	21
11	WB39	1.10	4.58	1.91	4.80	4.77	5.03	--	--	22
	WB41	1.08	4.75	1.45	4.59	4.58	5.23	0.82	0.17	22
	WB42	1.07	4.71	1.02	4.90	4.65	5.23	1.13	0.31	22
	WB43	1.07	4.72	0.45	4.52	4.65	5.24	2.52	0.33	22
12	WB44	1.08	2.68	1.90	4.67	4.62	5.15	--	--	19
	WB45	1.07	2.75	0.97	4.59	4.65	5.21	1.67	0.23	22
13	WB46	0.86	4.61	1.75	4.67	4.62	5.15	--	--	20
	WB47	0.88	4.72	1.45	4.60	4.58	5.18	0.74	0.20	25
	WB48	0.86	4.68	0.98	4.68	4.62	5.17	1.46	0.26	19
14	WB49	1.09	4.24	1.91	6.72	3.70	5.18	--	--	23
	WB50	1.07	4.72	1.96	5.01	4.27	5.23	--	--	22
15	WB51	0.82	4.68	1.94	4.51	4.62	5.21	--	--	20
	WB52	0.84	4.68	1.94	4.51	4.65	5.23	--	--	23

<sup>a</sup>All steels contained about 0.20 to 0.40 Mn and about 0.20 to 0.40 Si and the usual residual amounts of P, S, and Ni except Steels WB51 and WB52 which contained 0.67 and 1.51 Si, respectively.



Table II  
Austenitizing Survey of Experimental Steels<sup>a</sup>

Steel Designation	Austenitizing Treatment												Selected Austenitizing Temperature (OF)
	2200 F, 15 min		2250 F, 12 min		2300 F, 10 min		2325 F, 5 min		2350 F, 5 min		Grain Size		
	Hardness (Rc) Q	T	Hardness (Rc) Q	T	Grain Size	Hardness (Rc) Q	T	Grain Size	Hardness (Rc) Q	T			
WB1	64	60	--	--	61	63	60	c	62	59	c	2350	
WB2	64	63	--	--	64	63	64	c	63	64	*	2325	
WB3	65	64	--	--	64	63	65	f	63	64	c*	2325	
WB4	64	63	--	--	64	63	65	mf	63	66	c*	2325	
WB5	64	66	--	--	66	64	66	f	62	67	c*	2325	
WB6	64	67	--	--	66	61	67	f	59	67	c*	2325	
WB7	64	62	--	--	62	63	62	mf	62	62	c	2375	
WB8	64	64	--	--	64	62	64	mf	63	65	c	2300	
WB9	65	64	--	--	64	64	64	f	63	65	c	2325	
WB10	64	64	--	--	66	62	65	c	62	64	vc	2300	
WB11	64	65	--	--	67	61	--	c*	--	--	--	2275	
WB12	64	65	--	--	66	63	--	c*	--	--	--	2275	
WB13	64	62	--	--	63	63	63	c	63	63	vc	2375	
WB14	64	64	--	--	66	64	65	mf	64	65	vc*	2300	
WB15	65	64	--	--	63	63	65	f	63	66	c*	2325	
WB16	64	66	--	66	67	62	67	c	62	66	vc†	2275	
WB17 <sup>b</sup>	44	58	--	60	57	42	64	vc†	53	65	vc†	2150	
WB18	61	67	--	67	67	59	68	mf*	61	68	vc†	2275	
WB19	66	64	--	65	65	66	66	mf	67	66	vc*	2325	

<sup>a</sup>In the column headings, "Q" denotes the Rockwell "C" hardness after oil quenching and "T" denotes the hardness after a 2-hour temper at 1050 F.

<sup>b</sup>Steel WB17 was also austenitized at 2150 F; the Rockwell "C" hardness was 41 after oil quenching and 56 after a 2-hour temper at 1050 F; grains were medium fine.

<sup>c</sup>The following is the approximate Snyder-Graff grain size: vf = very fine (above 18); f = fine (14 to 18); mf = medium fine (9 to 13); c = coarse (5 to 8); and vc = very coarse (less than 5).

\*Specimen showing beginning of grain boundary melting.

†Considerable grain boundary melting in the sample.

Table II (continued)

Austenitizing Survey of Experimental Steels

Steel Designation	Austenitizing Treatment												Selected Austenitizing Temperature (°F)			
	2200 F, 15 min		2250 F, 12 min		2300 F, 10 min		2325 F, 5 min		2350 F, 5 min		Grain Size					
	Hardness (Rc) Q	Hardness (Rc) T	Grain Size	Hardness (Rc) Q	Hardness (Rc) T	Grain Size	Hardness (Rc) Q	Hardness (Rc) T	Grain Size	Hardness (Rc) Q		Hardness (Rc) T				
WB20	66	64	mf	66	65	f	66	65	mf	66	66	mf*	66	66	vc†	2300
WB21	63	67	mf	63	67	mf	62	67	c	68	68	c*	68	67	vc†	2275
WB22	65	66	f	64	67	f	65	67	mf	67	67	vc*	67	67	vc†	2300
WB23	66	65	f	67	66	mf	67	67	f	67	67	vc*	67	67	vc†	2300
WB24	66	66	f	66	67	mf	67	67	mf	67	67	vc*	67	66	vc†	2300
WB25	64	66	vf	64	67	vf	62	67	mf*	67	67	vc†	67	66	vc†	2275
WB26	64	66	f	63	67	mf	62	67	mf*	67	67	vc†	67	66	vc†	2275
WB27	63	66	f	63	66	vc*	63	66	mf	66	66	ct	66	66	ct	2225
WB28	63	65	f	65	67	vf	64	67	f	67	67	f	67	67	c	2325
WB29	64	67	f	65	67	f	62	67	mf	68	68	c*	68	67	vc†	2300
WB30	61	67	mf	62	68	mf*	63	67	ct	67	67	vc†	67	67	vc†	2225
WB31	65	66	mf	65	67	f	63	67	f	68	68	mf	68	67	c	2325
WB32	62	68	mf	62	67	vf	58	68	mf*	68	68	c*	68	68	ct	2275
WB33	61	67	mf	61	68	f	59	67	c*	67	67	ct	67	67	ct	2250
WB34	64	67	vf	64	68	f	63	68	mf	68	68	mf	68	68	c	2325
WB35	68	64	vf	67	66	mf	67	66	vc*	66	66	vc†	67	66	vc†	2250
WB36	64	67	vf	63	67	mf	63	68	c*	68	67	ct	67	67	vc†	2250
WB37	56	65	f	56	67	c	59	67	c*	67	67	ct	60	67	vc†	2250
WB38	47	59	mf	48	62	c*	56	67	vc†	66	66	vc†	57	66	vc†	2200
WB39	61	68	f	59	68	c*	60	68	vc†	68	68	vc†	62	68	vc†	2225
WB41	63	68	vf	60	68	f	61	68	c*	68	68	ct	62	68	vc†	2275
WB42	64	67	vf	61	68	f	60	67	c*	67	67	ct	62	68	vc†	2275
WB43	64	67	vf	61	68	f	60	67	c*	67	67	ct	62	67	vc†	2275

Table II (continued)  
Austenitizing Survey of Experimental Steels<sup>a</sup>

Steel Designation	Austenitizing Treatment																		Selected Austenitizing Temperature (OF)
	2200 F, 15 min			2250 F, 12 min			2300 F, 10 min			2325 F, 5 min			2350 F, 5 min						
	Hardness (Rc)		Grain Size <sup>c</sup>	Hardness (Rc)		Grain Size	Hardness (Rc)		Grain Size	Hardness (Rc)		Grain Size	Hardness (Rc)		Grain Size				
	Q	T		Q	T		Q	T		Q	T		Q	T					
WB44	66	67	vf	64	68	c*	65	68	ct	65	68	c*	65	68	c*	2275			
WB45	66	67	vf	64	69	f	65	68	c*	65	68	c*	65	68	ct	2275			
WB46	65	66	vf	63	67	f	63	67	ct	63	67	ct	63	67	c*	2275			
WB47	65	65	vf	64	66	mf*	64	67	ct	64	67	ct	63	66	ct	2275			
WB48	65	65	vf	65	65	f	62	65	mf	62	65	mf	63	64	mf	2350			
WB49	63	68	vf	60	69	c*	62	68	ct	62	68	ct	62	68	vc†	2275			
WB50	62	68	vf	59	68	c*	61	68	vc†	61	68	vc†	62	68	vc†	2275			
WB51	65	66	vf	64	66	f	63	66	ct	63	67	ct	63	66	vc†	2275			
WB52	64	65	vf	64	65	vf	63	66	c*	62	66	c*	63	64	ct	2300			

<sup>a</sup>In the column headings, "Q" denotes the Rockwell "C" hardness after oil quenching and "T" denotes the hardness after a 2-hour temper at 1050 F.

<sup>c</sup>The following is the approximate Snyder-Graff grain size: vf = very fine (above 18); f = fine (14 to 18); mf = medium fine (9 to 13); c = coarse (5 to 8); and vc = very coarse (less than 5).

\* Specimen showing beginning of grain boundary melting.

† Considerable grain boundary melting in the sample.





Table IV

Hot Hardness Measurements of Steels as-Heat-Treated

Steel Designation	Heat-Treatment <sup>a</sup>	Rockwell "C" Hardness at				
		Room Temp.	600 F	800 F	1000 F	Room Temp. After Test
WB18	Aust. 2225 F/12 min	68	63	60	58	68
WB21	Aust. 2225 F/12 min	68	63	61	58	68
WB25	Aust. 2250 F/10 min	68	63	60	58	68
WB39	Aust. 2225 F/12 min	69	64	--	59	69
WB41	Aust. 2250 F/10 min	68	64	62	59	68
WB44	Aust. 2250 F/10 min	68	64	61	58	68
WB45	Aust. 2270 F/10 min	68	64	62	59	68
WB49	Aust. 2250 F/10 min	69	64	61	59	69

<sup>a</sup> All specimens were oil quenched from the austenitizing temperature and tempered; the tempering involved two consecutive 2-hour tempers at 1050 F.

Table V

Hot Hardness Measurements after a 1000-Hour Exposure at 600 F<sup>a</sup>

Steel Designation	Rockwell "C" Hardness at				Room Temp. After Test
	Room Temp.	600 F	800 F	1000 F	
WB18	68	63	60	58	68
WB21	67	63	60	57	67
WB25	67	63	60	58	67
WB39	68	64	62	60	68
WB41	68	64	62	61	68
WB44	68	64	61	58	68
WB45	68	64	61	58	68
WB49	69	65	62	59	69

<sup>a</sup>Prior to the 600 F exposure, all specimens were heat-treated as indicated in Table IV.

Table VI

Hot Hardness Measurements after a 1000-Hour Exposure at 800 F<sup>a</sup>

Steel Designation	Rockwell "C" Hardness at				
	Room Temp.	600 F	800 F	1000 F	Room Temp. After Test
WB18	68	63	62	59	68
WB21	67	63	61	57	67
WB25	67	64	62	59	67
WB39	69	65	62	59	69
WB41	68	64	62	59	68
WB44	68	64	62	59	68
WB45	68	64	62	59	68
WB49	69	65	62	59	69

<sup>a</sup>Prior to the 800 F exposure, all specimens were heat-treated as indicated in Table IV.



Table VII

Hot Hardness Measurements after a 500-Hour Exposure at 1000 F<sup>a</sup>

Steel Designation	Rockwell "C" Hardness at				
	Room Temp.	600 F	800 F	1000 F	Room Temp. After Test
WB18	65	61	59	55	65
WB21	65	61	58	55	65
WB25	63	59	56	53	63
WB39	66	62	59	56	66
WB41	65	61	58	56	65
WB44	65	62	60	57	65
WB45	65	60	57	54	65
WB49	66	62	60	57	66

<sup>a</sup>Prior to the 1000 F exposure, all specimens were heat-treated as indicated in Table IV.

Table VIII

Room-Temperature Hardness of Specimens  
before and after the Elevated-Temperature Exposures

Steel Designation	Rockwell "C" Hardness			
	As-Heat-Treated <sup>a</sup>	After 1000 hr at 600 F	After 1000 hr at 800 F	After 500 hr at 1000 F
WB18	68	68	68	65
WB21	68	67	67	65
WB25	68	67	67	63
WB39	69	68	69	66
WB41	68	68	68	65
WB44	68	68	68	65
WB45	68	68	68	65
WB49	69	69	69	66

<sup>a</sup>Prior to the elevated-temperature exposures, all specimens were heat-treated as indicated in Table IV.

Table IX

Results of Dimensional Stability Measurements<sup>a</sup>

Steel Designation	Length Change (μin./in.)											
	600 F				800 F				1000 F			
	500 hours		1000 hours		500 hours		1000 hours		250 hours		500 hours	
	Measured	Aver- age	Measured	Aver- age	Measured	Aver- age	Measured	Aver- age	Measured	Aver- age	Measured	Aver- age
WB18	-46 -45	-46	-53 -51	-52	+6 +55	+31	+16 --b	+16	+201 +159	+180	+185 +140	+163
WB21	+26 +28	+27	none +14	+7	-28 -26	-27	-33 --b	-33	+160 +194	+177	+140 +159	+150
WB25	-1 -3	-2	-40 none	-20	-94 +16	-39	-83 +21	-31	+39 +19	+29	+45 +23	+34
WB39	-3	-3	-4	-4	--	--	--	--	+63	+63	+289	+289
WB41	-5 --	-5	-5 --	-5	+4 -25	-11	-3 -23	-13	+59 --	+59	+66 --	+66
WB44	+10 +6	+8	+5 +9	+7	none -6	-3	+3 --b	+3	+35 +26	+31	+19 +4	+12
WB45	-5 -11	-8	-8 -11	-10	-1 +99	+49	+14 --b	+14	+25 -11	+7	+9 -19	-5
WB49	-5 -4	-5	-9 -3	-6	-6 -10	-8	-1 -8	-5	+25 +58	+42	+13 +51	+32

a prior to the elevated-temperature exposures, all specimens were heat-treated as indicated in Table IV.

b these specimens scaled; hence the measurements were discarded.

Table X  
Compressive Yield Strength of Selected Experimental Steels<sup>a</sup>

Steel Designation	Compressive Yield Strength (psi)											
	Room Temp.		600 F			800 F			1000 F			
	0.1% offset	0.2% offset	0.1% offset	0.2% offset	0.1% offset	0.2% offset	0.1% offset	0.2% offset	0.1% offset	0.2% offset		
WB18	317,000 <sup>b</sup>	384,000 <sup>b</sup>	311,000 <sup>b</sup>	360,000 <sup>b</sup>	314,000	348,000	269,000 <sup>b</sup>	301,000 <sup>b</sup>				
WB21	311,000 <sup>b</sup>	374,000 <sup>b</sup>	253,000 <sup>b</sup>	324,000 <sup>b</sup>	--	-- <sup>c</sup>	284,000 <sup>b</sup>	308,000 <sup>b</sup>				
WB25	304,000	384,000	303,000 <sup>b</sup>	356,000 <sup>b</sup>	--	-- <sup>c</sup>	273,000 <sup>b</sup>	311,000 <sup>b</sup>				
WB39	--	--	--	--	--	--	301,000	330,000				
WB41	545,000	555,000	301,000	367,000	--	-- <sup>c</sup>	326,000	352,000 <sup>b</sup>				
WB44	463,000 <sup>b</sup>	500,000 <sup>b</sup>	336,000	384,000	--	-- <sup>c</sup>	317,000	342,000 <sup>b</sup>				
WB45	472,000	507,000	347,000	395,000	--	-- <sup>c</sup>	305,000	333,000 <sup>b</sup>				
WB49	468,000	503,000	371,000	405,000	331,000	370,000	299,000	332,000				

<sup>a</sup>All specimens were heat-treated as shown in Table IV.

<sup>b</sup>Average value from two tests.

<sup>c</sup>Unreliable test results due to defect in the strain recording train.

Table XI

Oxidation Test Results at 1000 F for 1000 Hours

Steel Designation	Metal Loss (mg/sq.in. surface area)		Average Depth of Oxide Layer ( $\mu$ in.)	
	Observed	Average	Calculated	Average (Calculated)
WB18	7.9	7.1	60.2	54
	6.2		47.5	
WB21	25.3	23.7	192.5	180
	22.1		168.1	
WB25	5.8	5.8	44.2	44
WB41	6.7	5.9	51.0	44
	5.0		37.7	
WB44	11.9	11.3	90.4	86
	10.6		80.6	
WB45	12.2	11.7	92.8	81
	11.2		70.0	
WB49	6.7	6.5	50.3	49
	6.2		47.1	
WB51	2.8	4.2	20.8	32
	5.6		42.3	
WB52	4.4	3.5	33.3	26
	2.5		19.2	

Table XII

Corrosion Test Results in MIL-7808-C Oil  
at 400 F for 400 Hours

Steel Designation	Change of Weight <sup>a</sup>		Surface Appearance
	(mg)	(%)	
SAE 52100	-1.5	-0.036	Pitted; most corroded steel
M2	+0.3	+0.007	Fairly bright and clean
WB18	+0.3	+0.007	Slight pitting; brown but clean
WB21	-0.1	-0.002	Specimen bright
WB25	+0.1	+0.002	Specimen bright
WB41	-0.4	-0.010	Slight pitting; dull surface
WB44	-0.7	-0.017	Fine pits; dark specimen
WB45	+0.2	+0.005	Specimen bright
WB49	+0.4	+0.009	Specimen bright
WB51	-0.1	-0.002	Specimen quite bright
WB52	none	none	Specimen quite bright

<sup>a</sup>The specimen weight varied from 4.09 to 4.26 g.

Table XIII

Corrosion Test Results in Octa Decyl Tri Decyl Silane  
at 600 F for 400 Hours

Steel Designation	Change of Weight <sup>a</sup>		Surface Appearance
	(mg)	(%)	
M2	-0.3	-0.007	Dark; uniform fine pits
WB18	none	none	Dark, but brighter than M2
WB21	-0.3	-0.007	Slight pitting; slightly darkened surface
WB25	none	none	Dark, but clean
WB41	none	none	Dark, but bright
WB44	-0.1	-0.002	Slightly dark
WB45	-0.3	-0.007	Slight pitting; fairly bright but dark
WB49	none	none	Dark, but clean and shiny
WB51	-0.1	-0.002	Slightly dark, but shiny
WB52	-0.2	-0.005	Slightly dark, but shiny

<sup>a</sup>The specimen weight varied from 4.15 to 4.27 g.

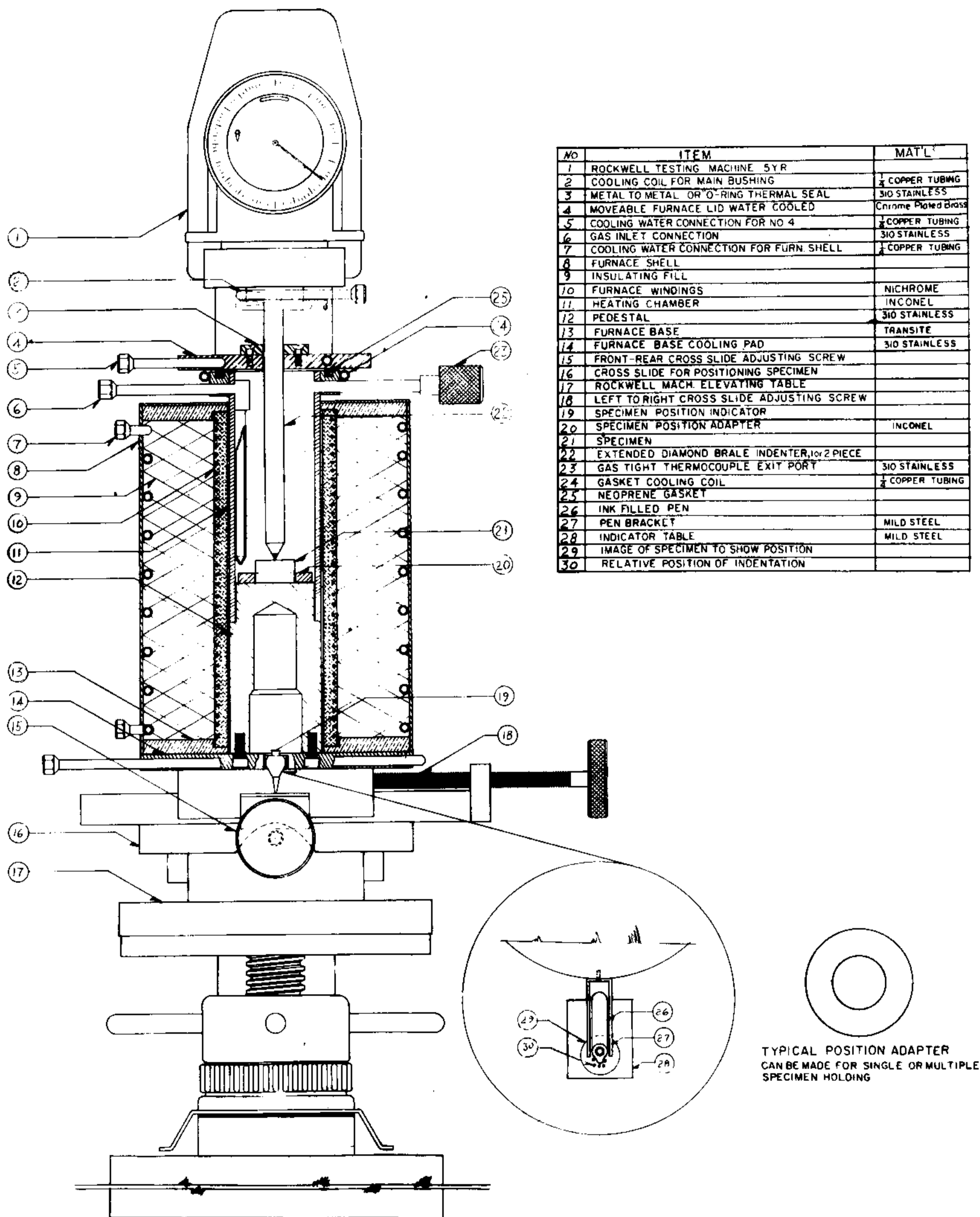


Figure 1. Schematic Diagram of the Hot Hardness Tester



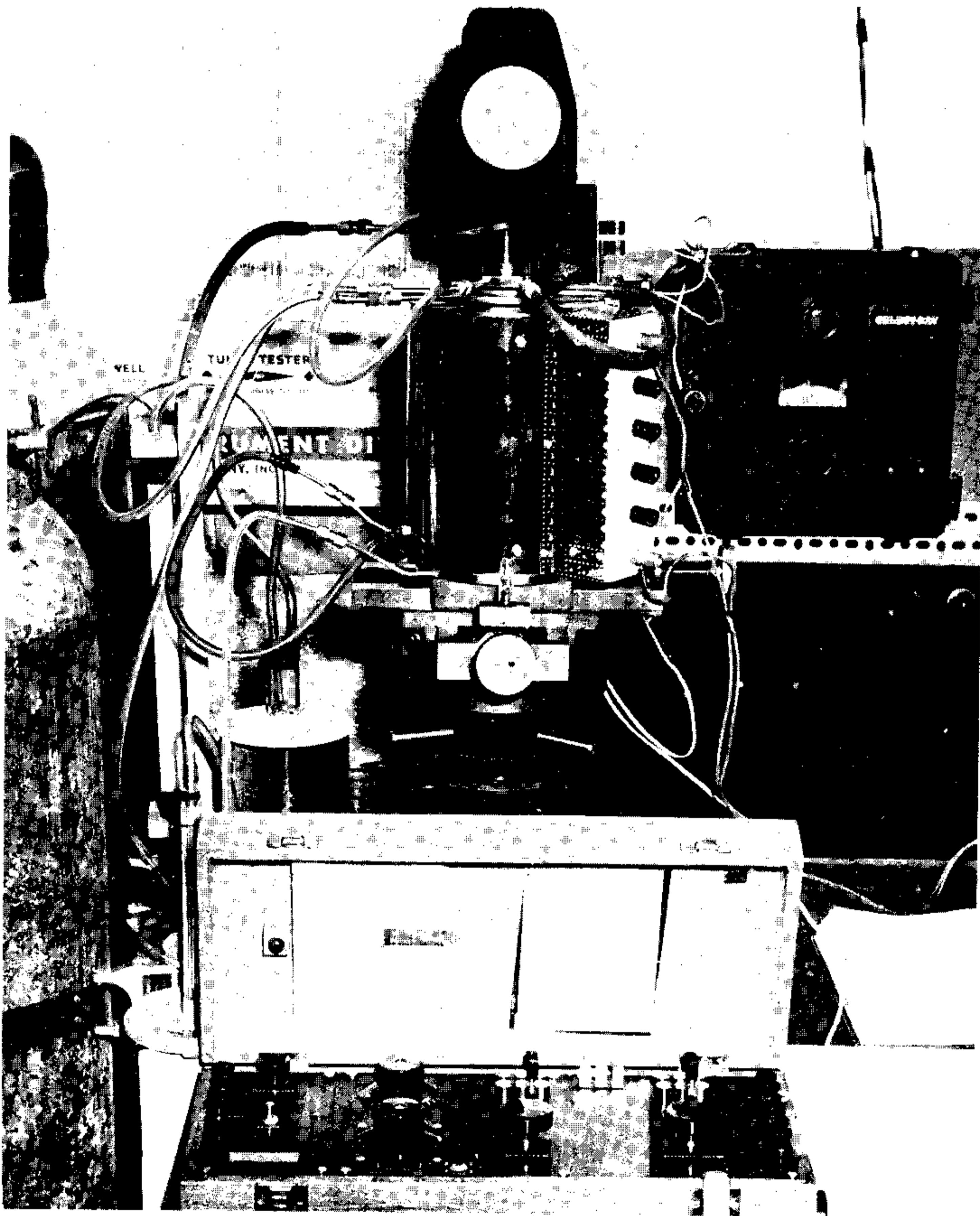


Figure 2. Hot Hardness Tester

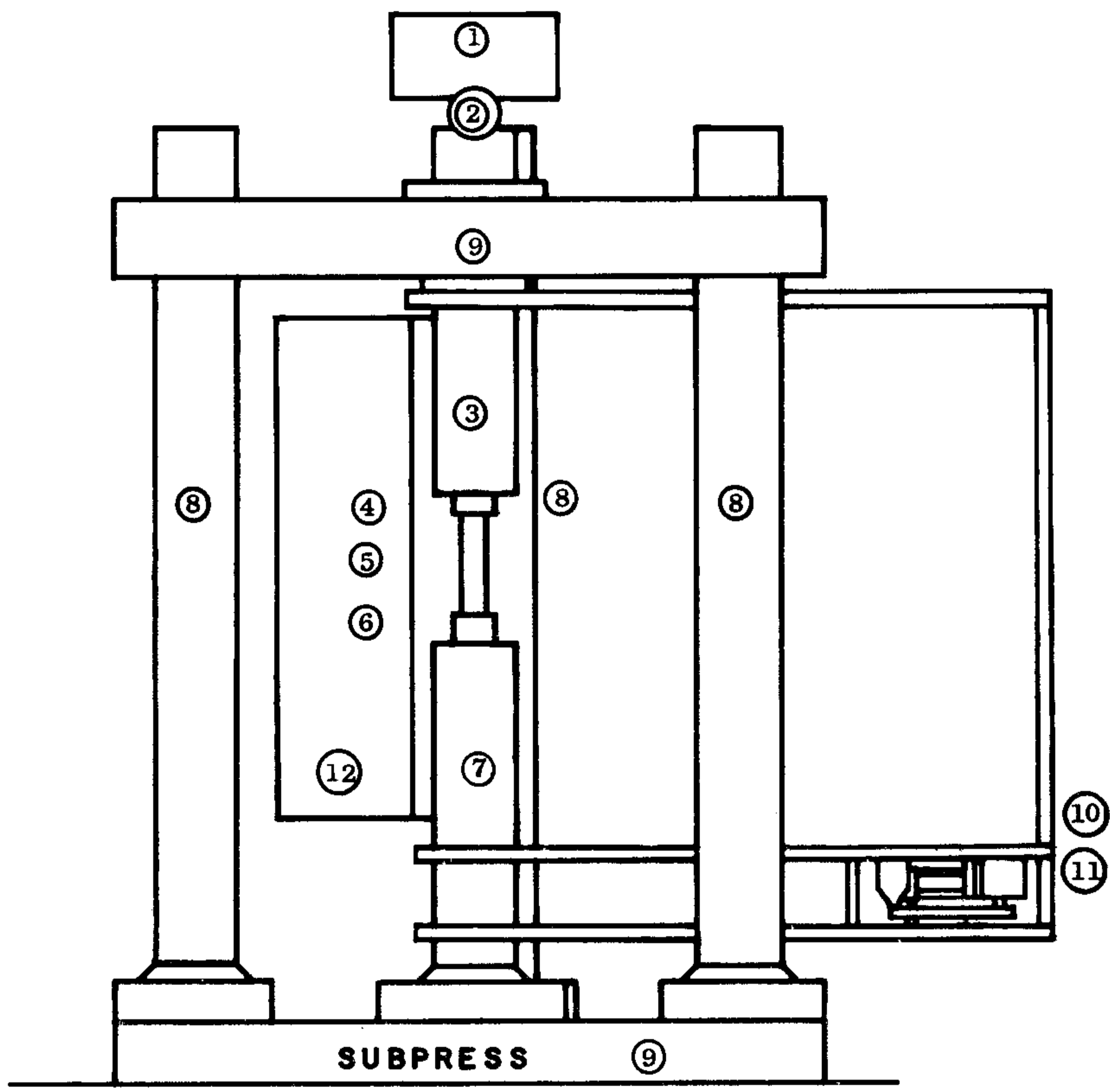


Figure 3. Schematic Diagram of Compression Test Fixture

Legend

- |  |                                     |
|--|-------------------------------------|
| 1. Top Loading Plate                     | 7. Lower Compression Rod            |
| 2. Hardened Steel Ball                   | 8. Guide Posts                      |
| 3. Upper Compression Rod                 | 9. Upper and Lower Subpress Plates  |
| 4. Upper Cemented Carbide Pressure Plate | 10. Linear Differential Transformer |
| 5. Test Specimen                         | 11. Extensometer Frame Arms         |
| 6. Lower Cemented Carbide Pressure Plate | 12. Furnace                         |

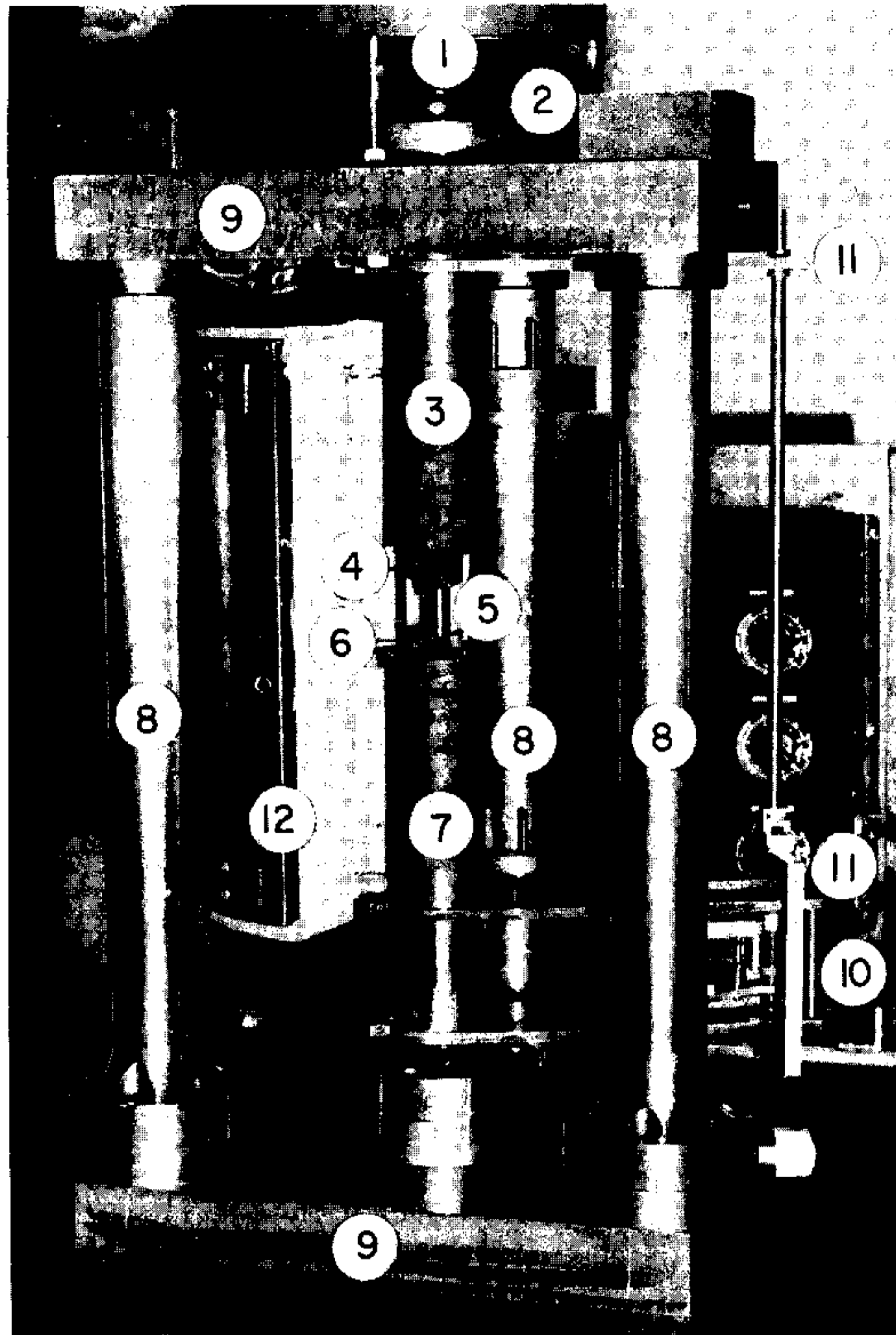


Figure 4. Compression Test Fixture

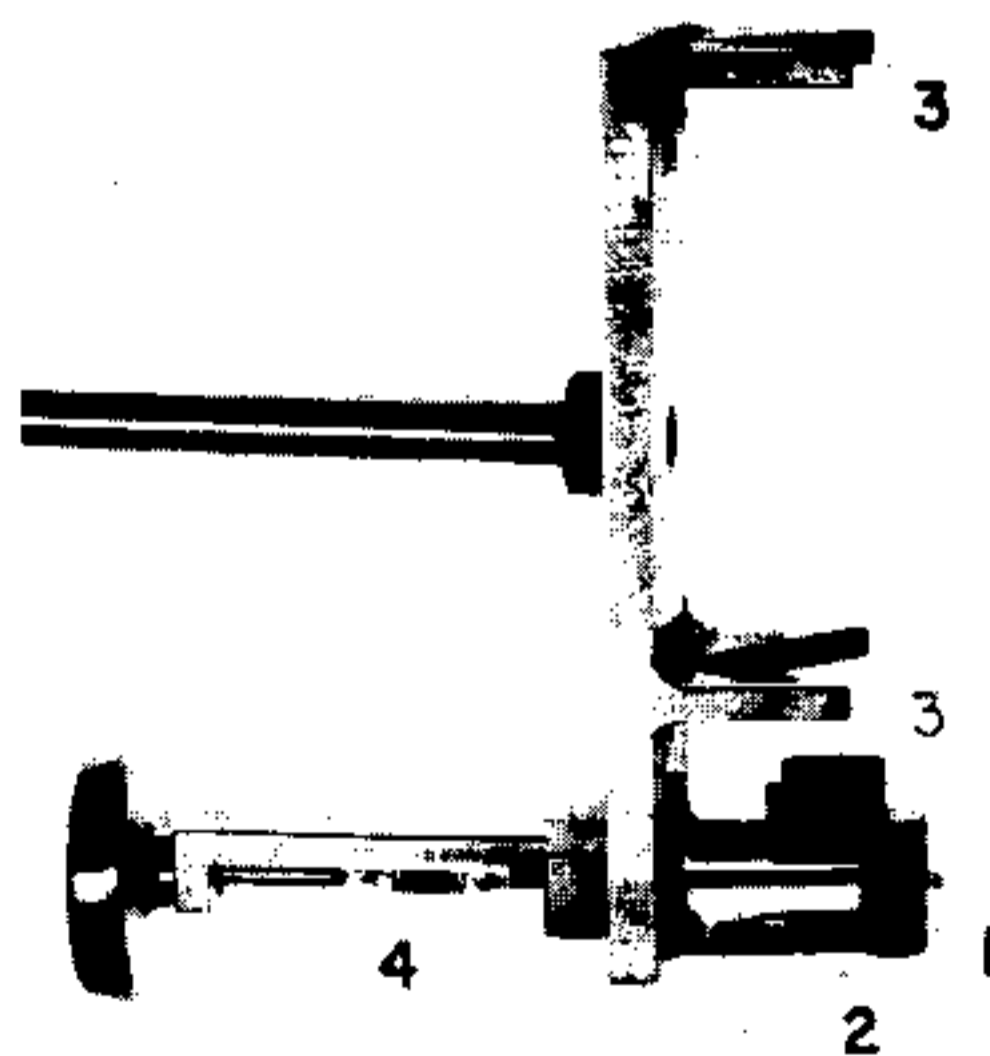


Figure 5. Auxiliary Specimen Locating Fixture

Legend

- |                                  |                                  |
|----------------------------------|----------------------------------|
| 1. Specimen                      | 3. Large V-Notch Locating Plates |
| 2. Small V-Notch Locating Plates | 4. Hook and Spring Arrangement   |

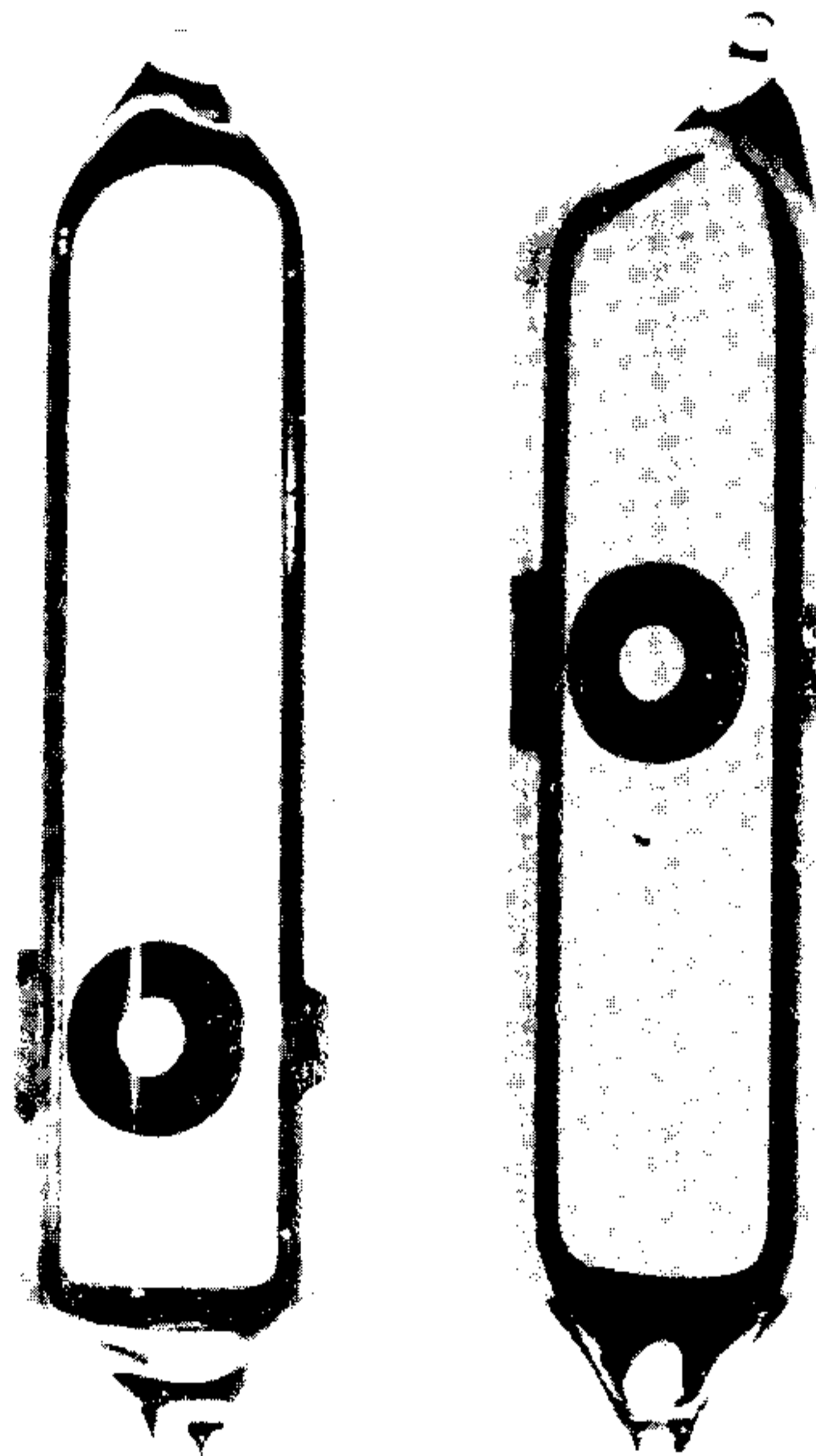


Figure 6. Oil Corrosion Test Specimens  
Inside Pyrex Capsules

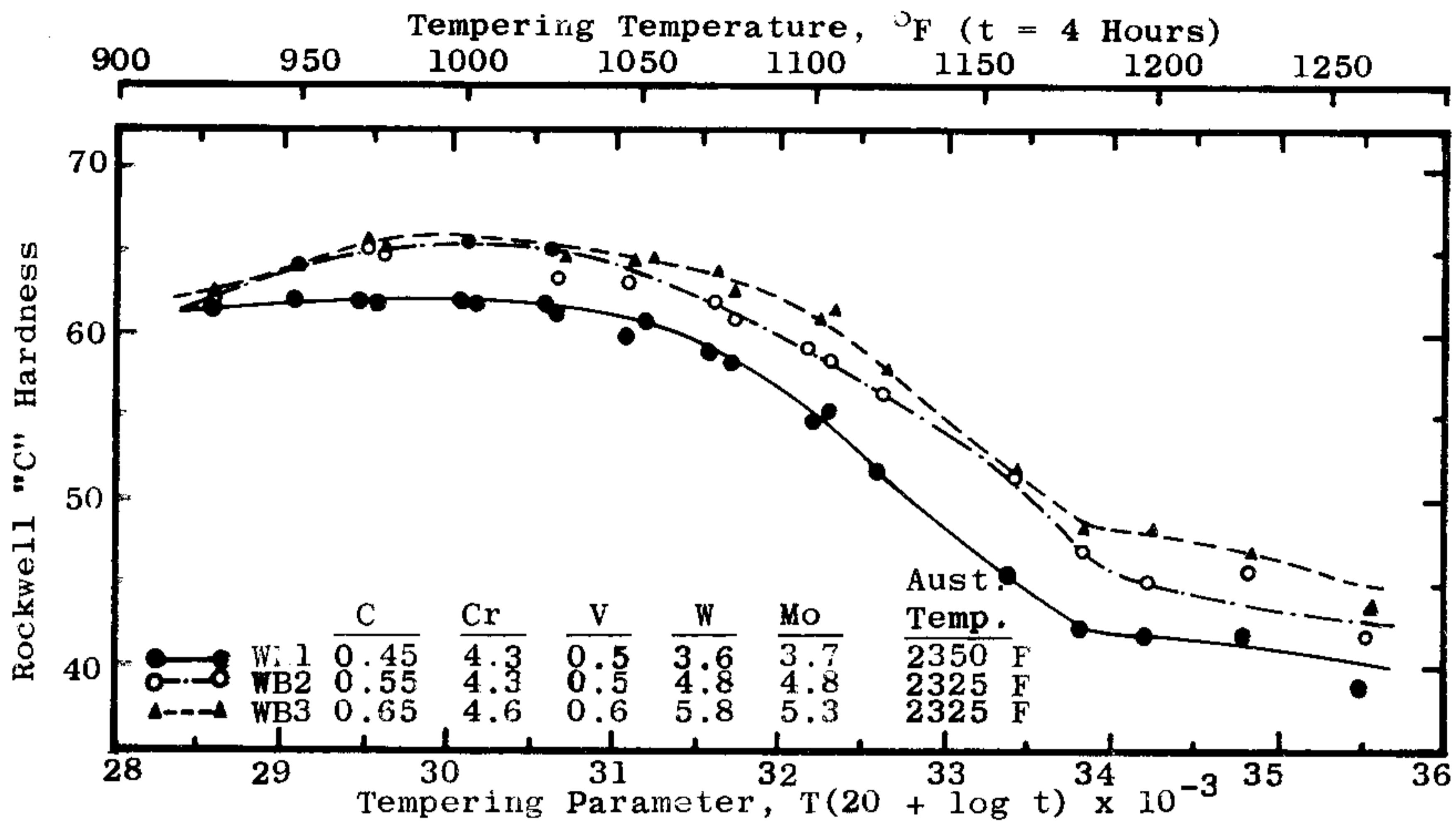


Figure 7. Master Tempering Curves for Steels WB1, WB2, and WB3

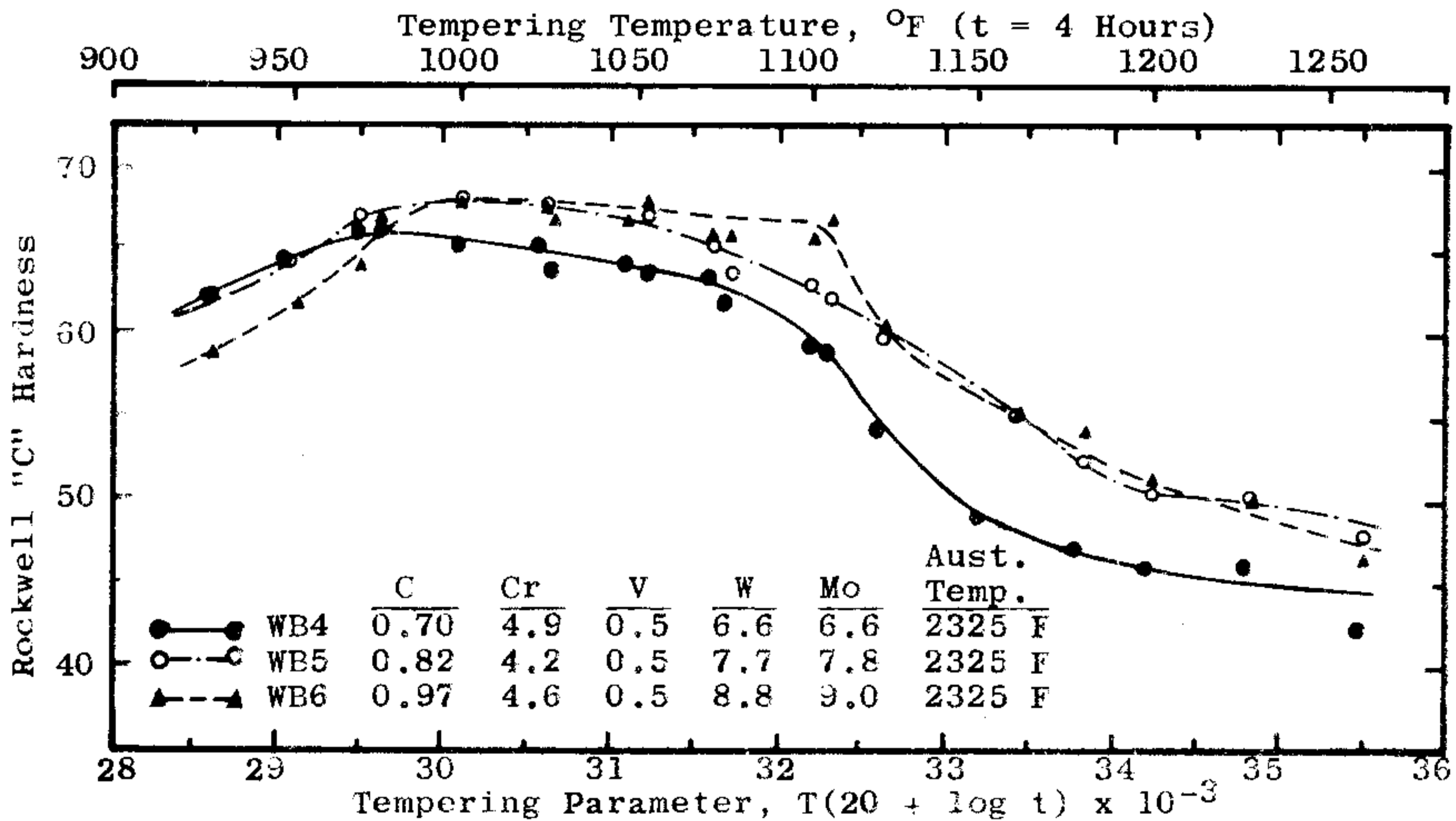


Figure 8. Master Tempering Curves for Steels WB4, WB5, and WB6

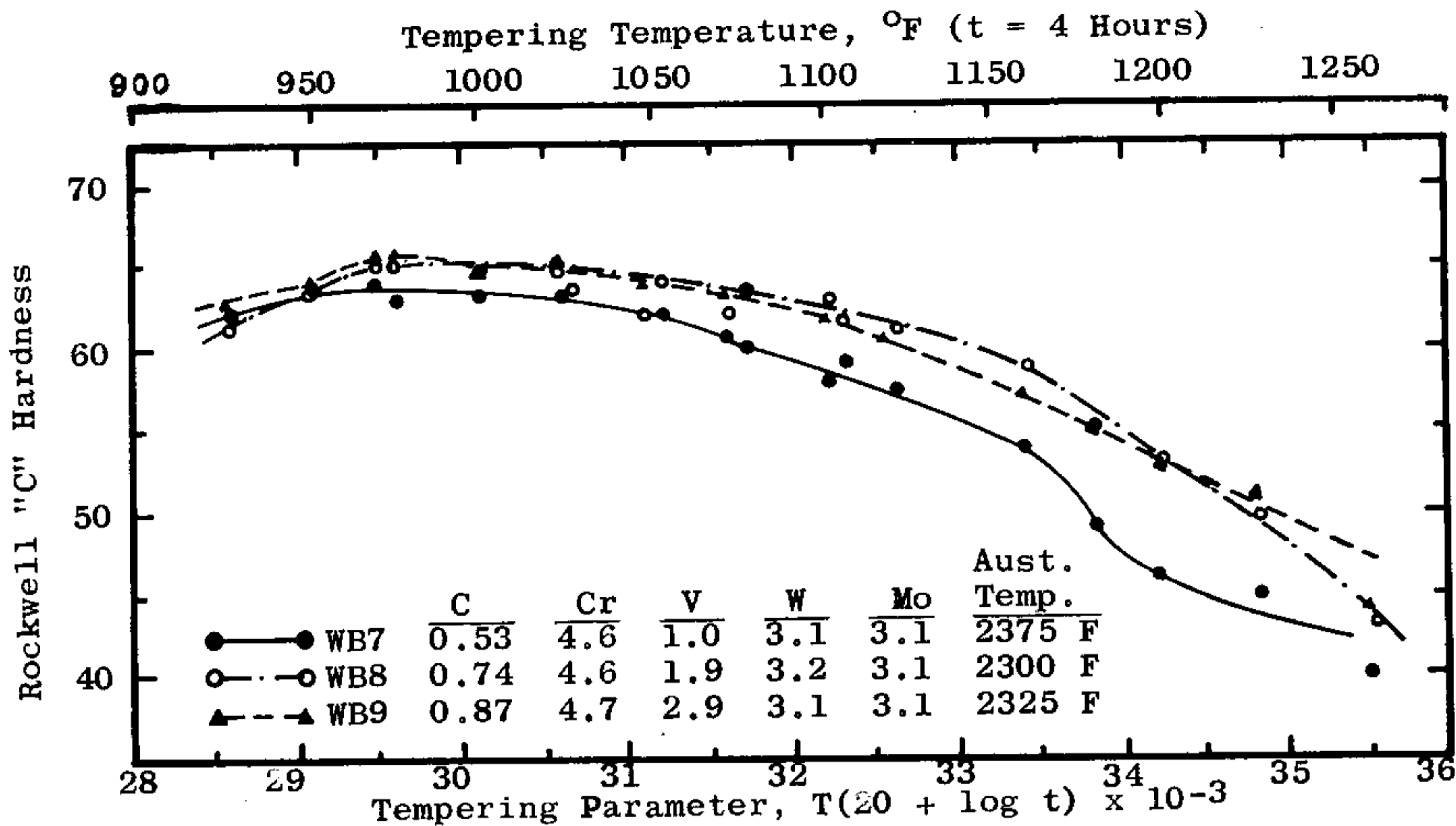


Figure 9. Master Tempering Curves for Steel WB7, WB8, and WB9

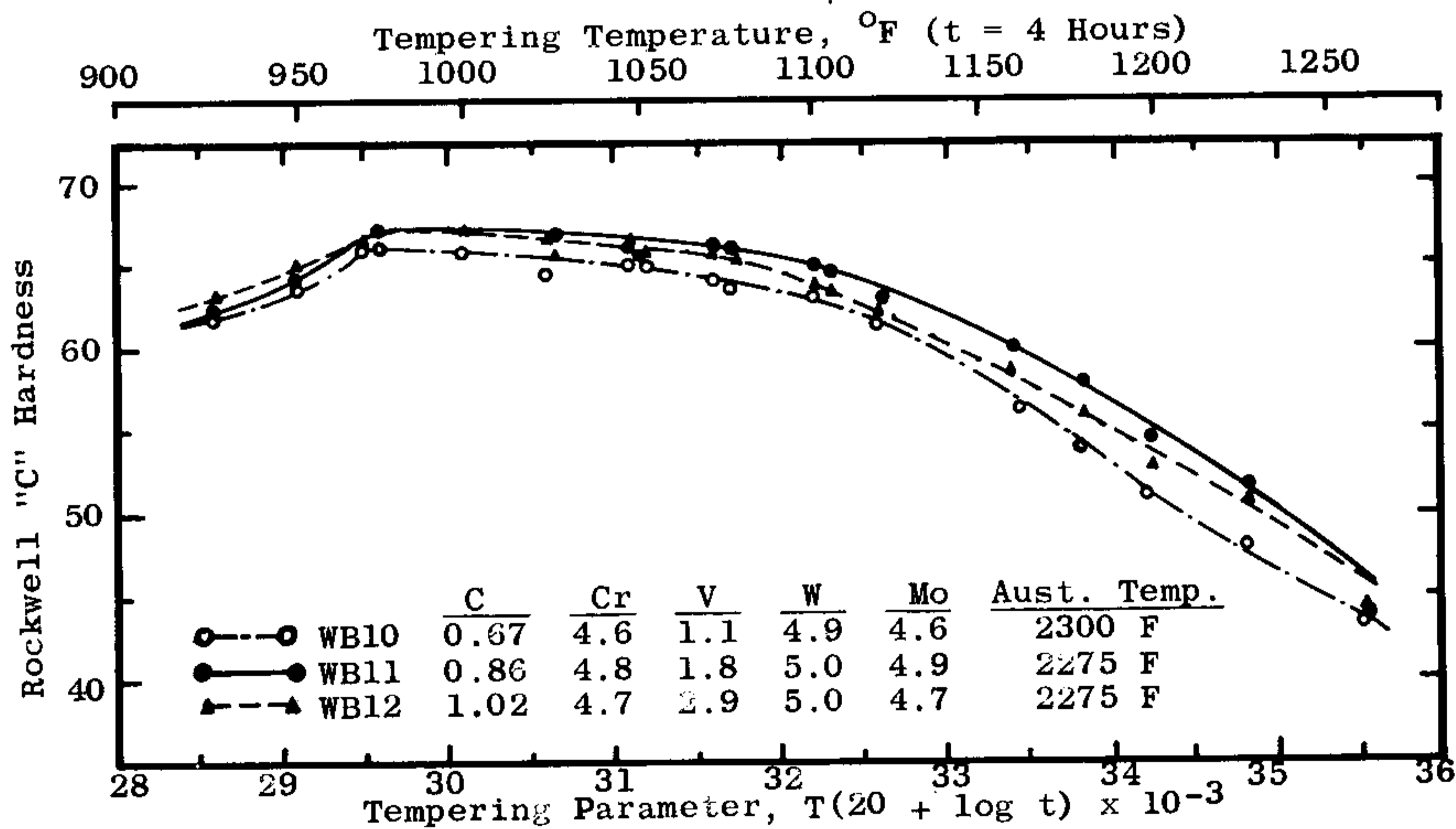


Figure 10. Master Tempering Curves for Steels WB10, WB11, and WB12

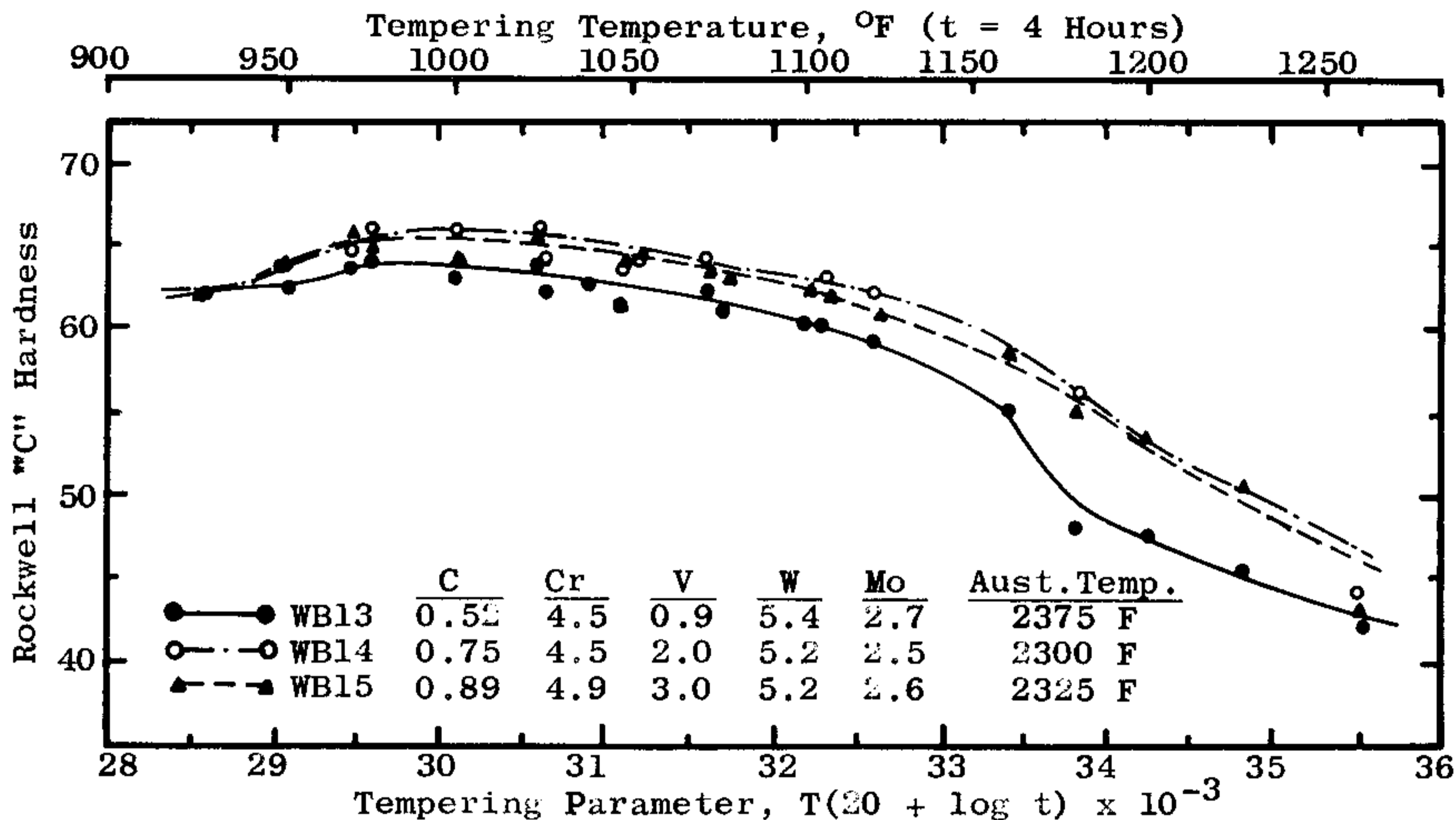


Figure 11. Master Tempering Curves for Steels WB13, WB14, and WB15

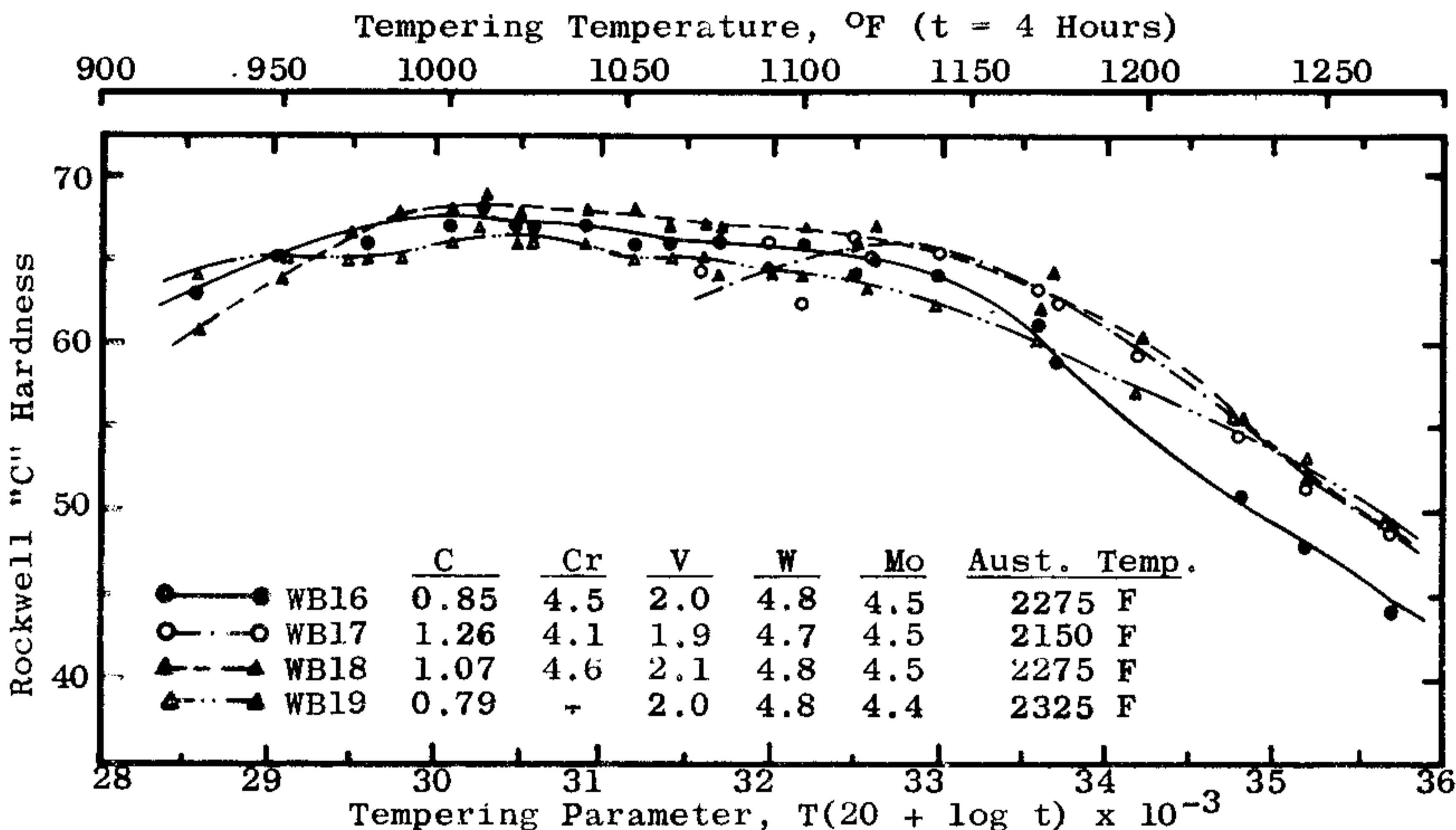


Figure 12. Master Tempering Curves for Steels WB16, WB17, WB18, and WB19

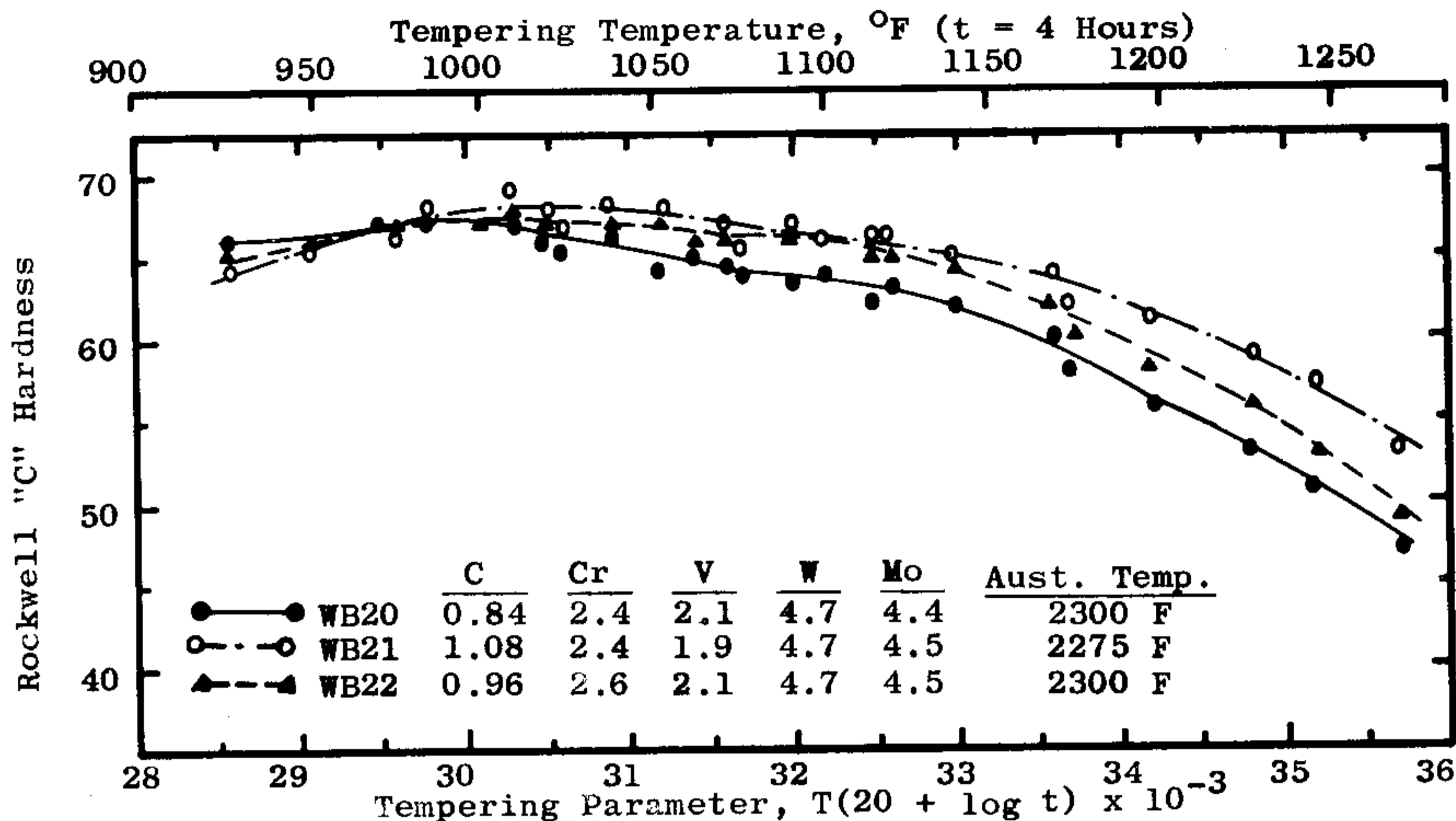


Figure 13. Master Tempering Curves for Steels WB20, WB21, and WB22

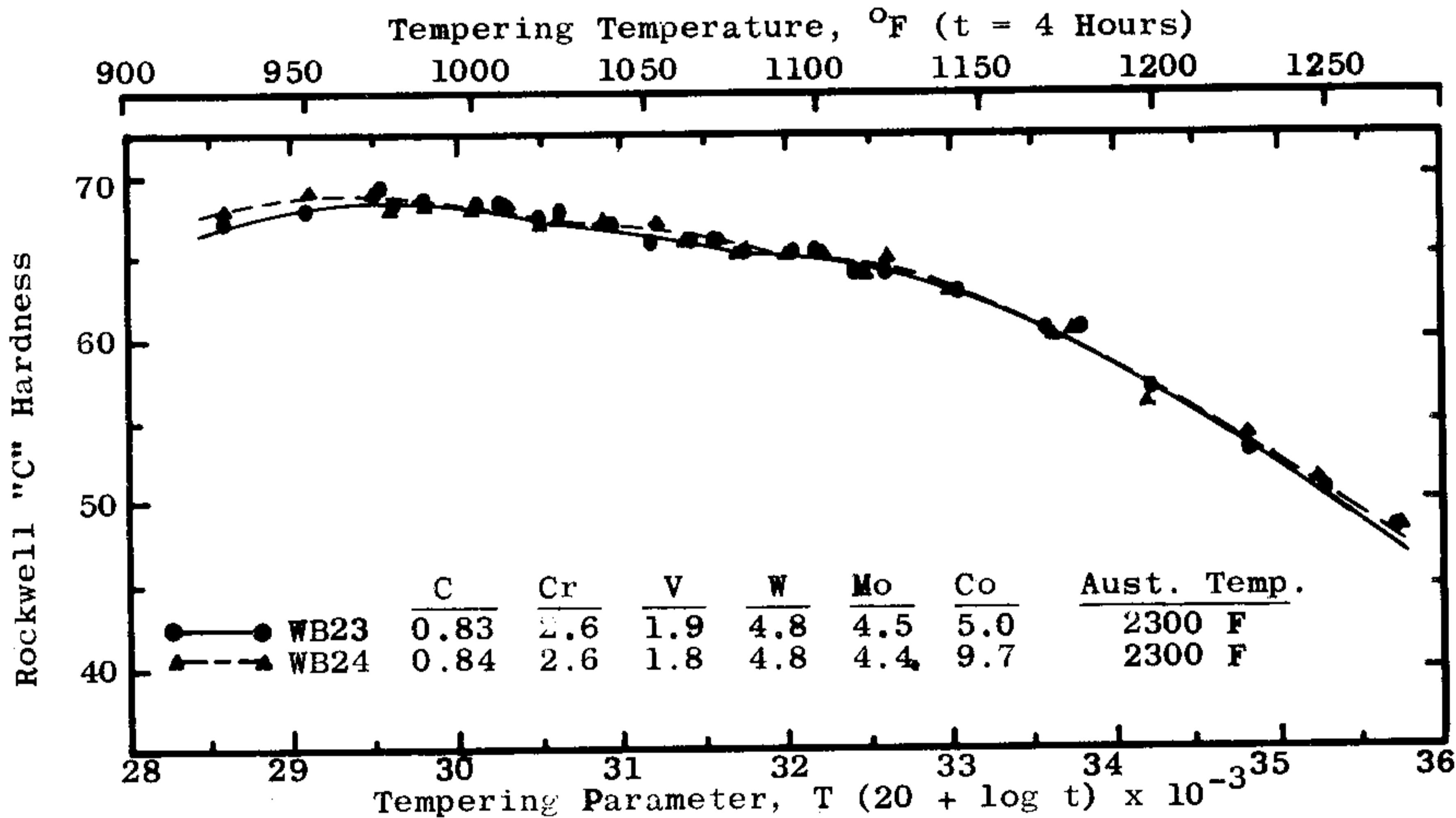


Figure 14. Master Tempering Curves for Steels WB23 and WB24



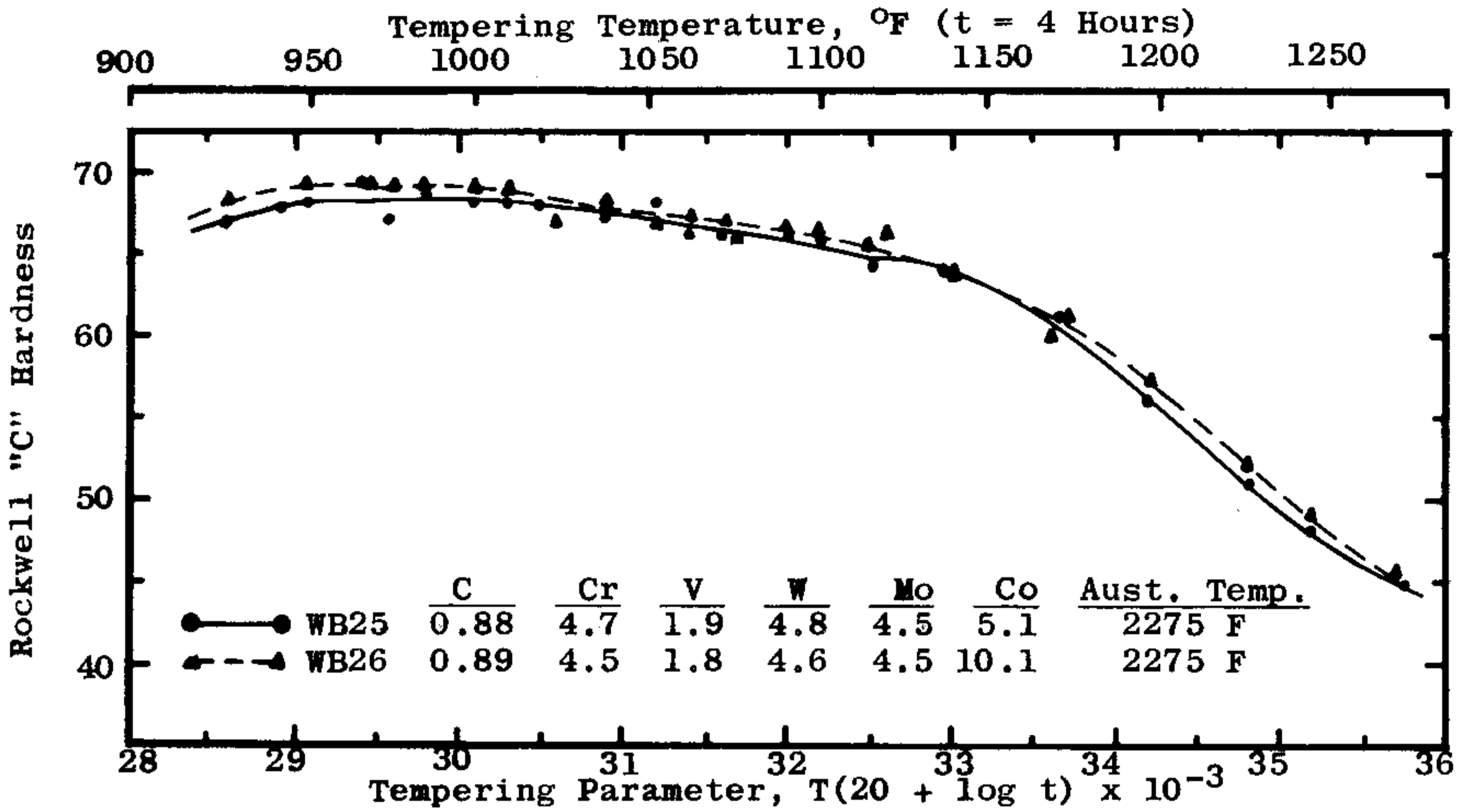


Figure 15. Master Tempering Curves for Steels WB25 and WB26

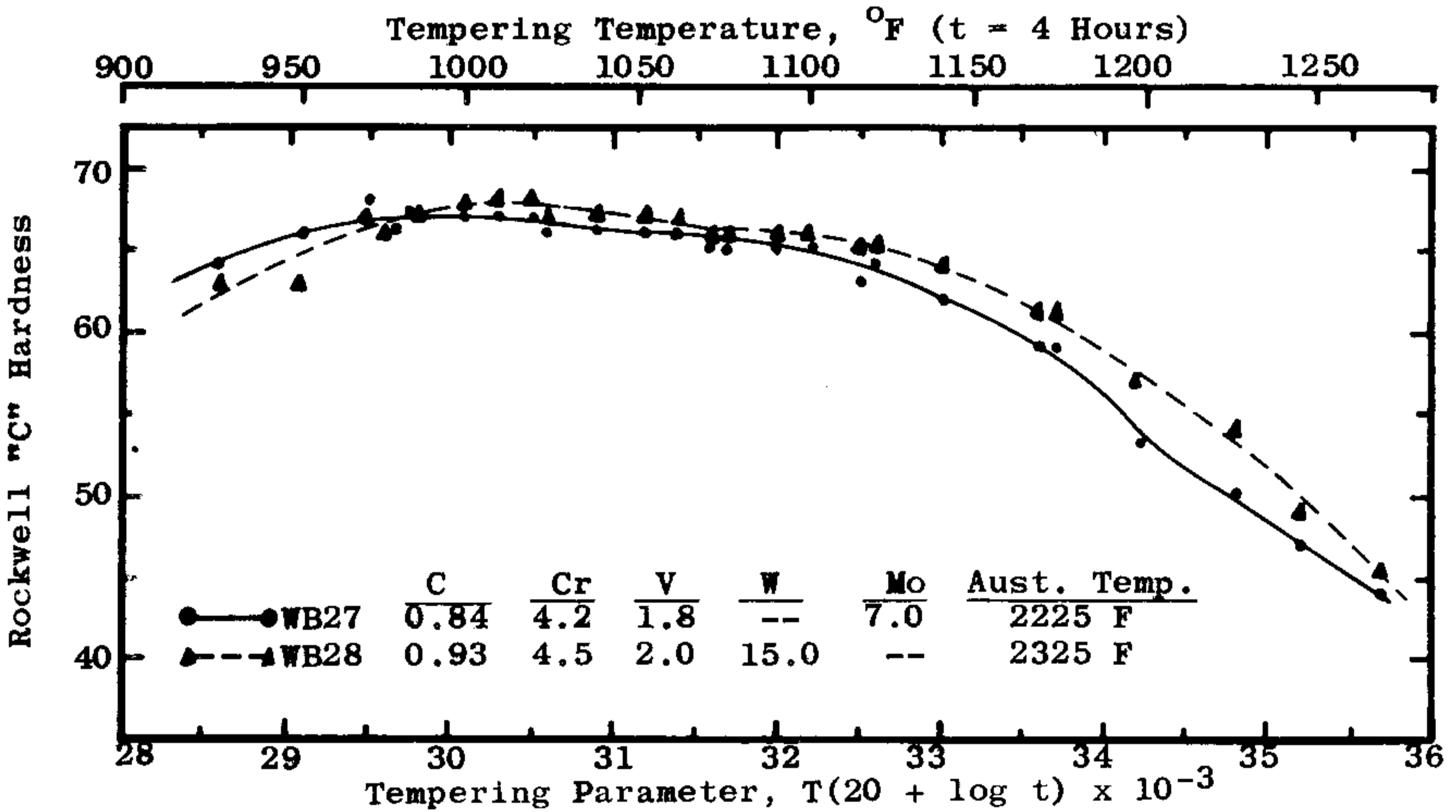


Figure 16. Master Tempering Curves for Steels WB27 and WB28

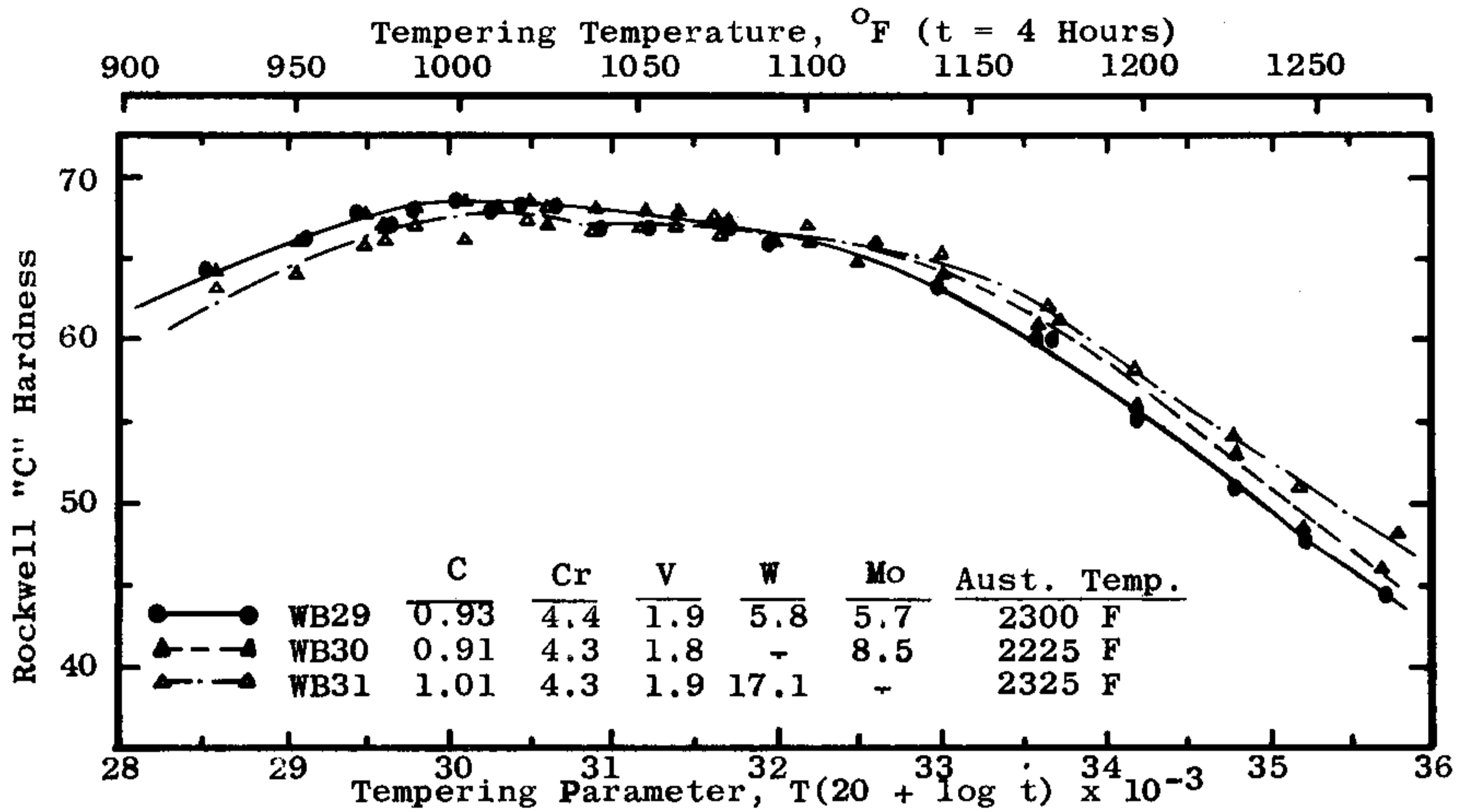


Figure 17. Master Tempering Curves for Steels WB29, WB30, and WB31

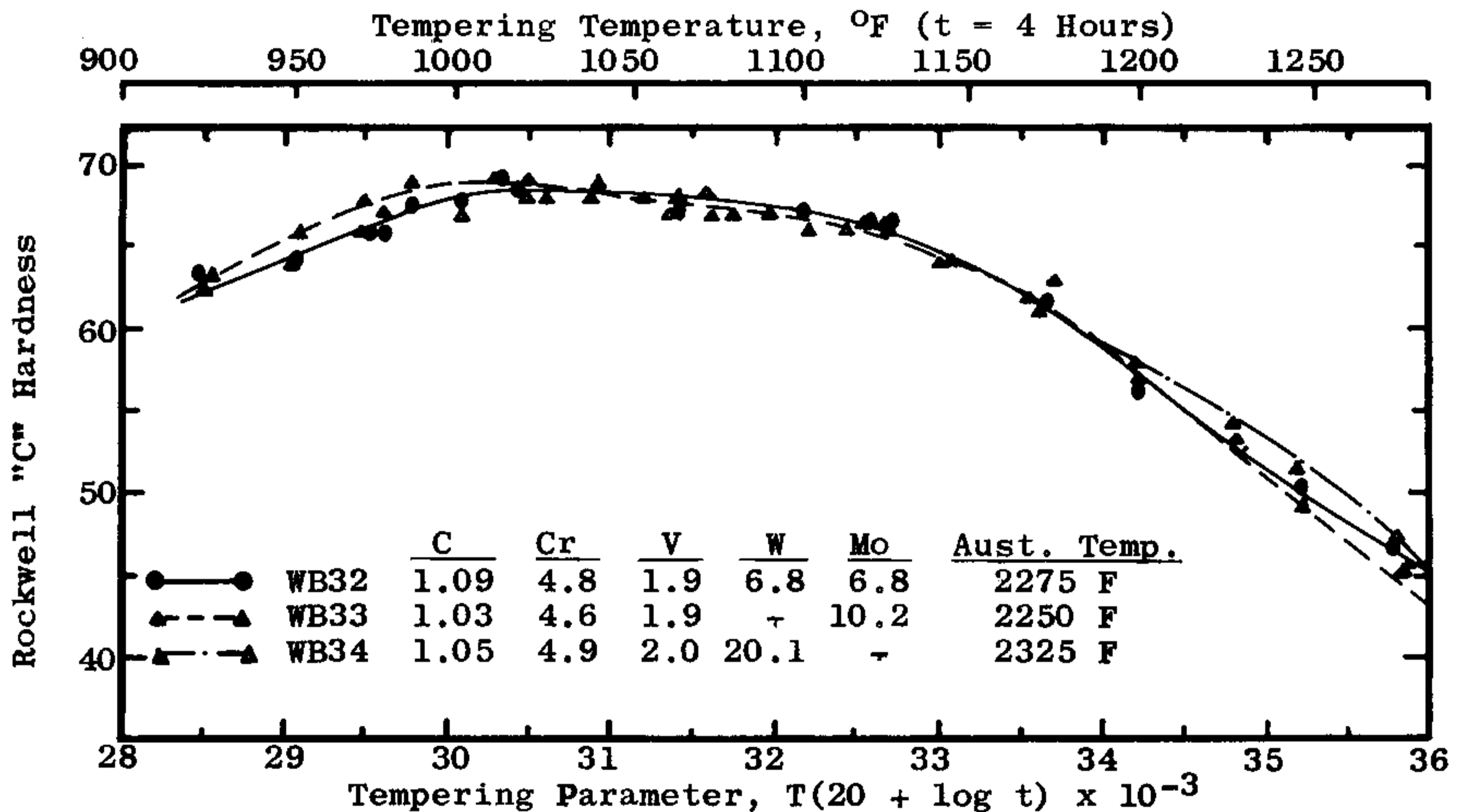


Figure 18. Master Tempering Curves for Steels WB32, WB33, and WB34

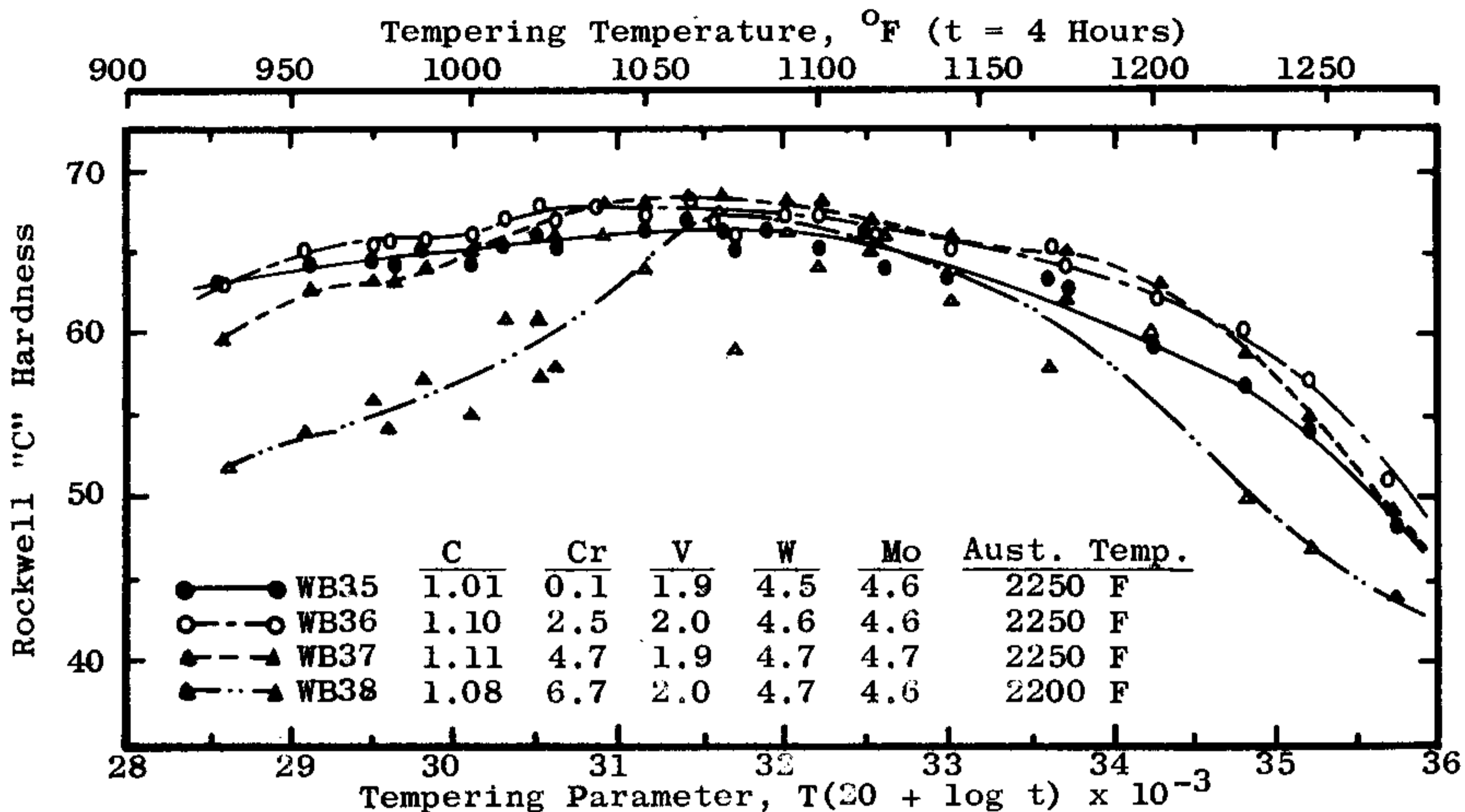


Figure 19. Master Tempering Curves for Steels WB35, WB36, WB37, and WB38

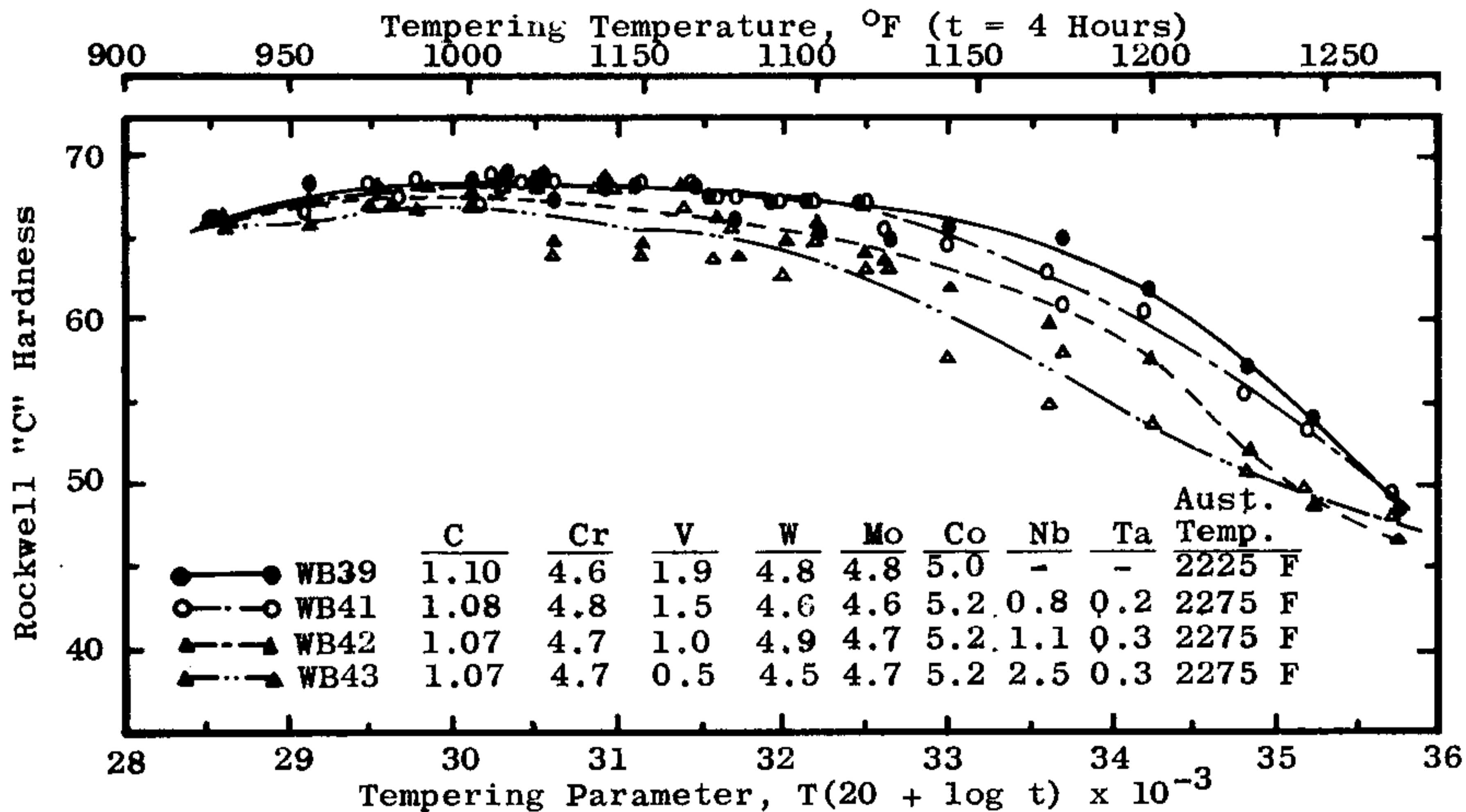


Figure 20. Master Tempering Curves for Steels WB39, WB41, WB42, and WB43

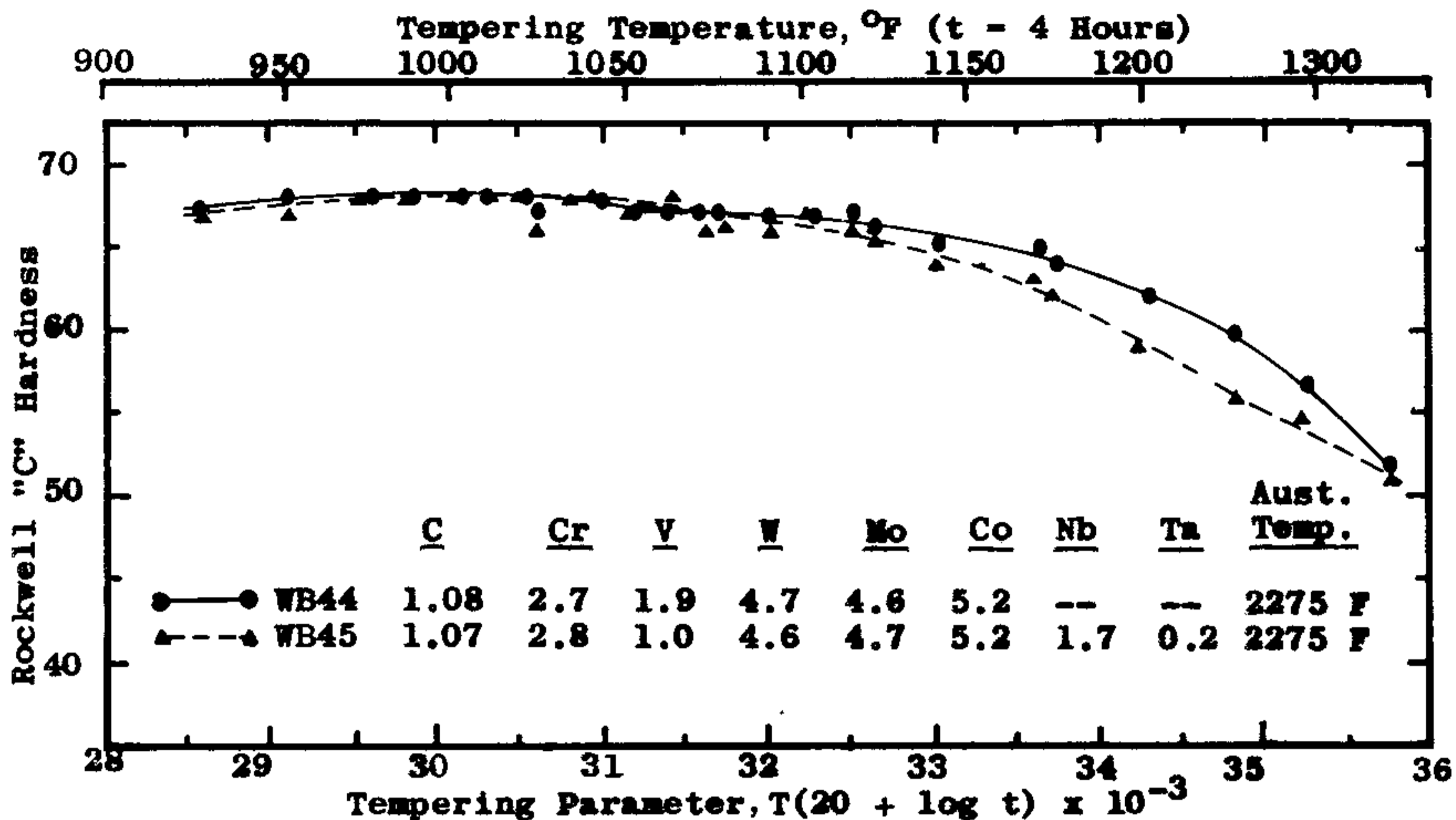


Figure 21. Master Tempering Curves for Steels WB44 and WB45.

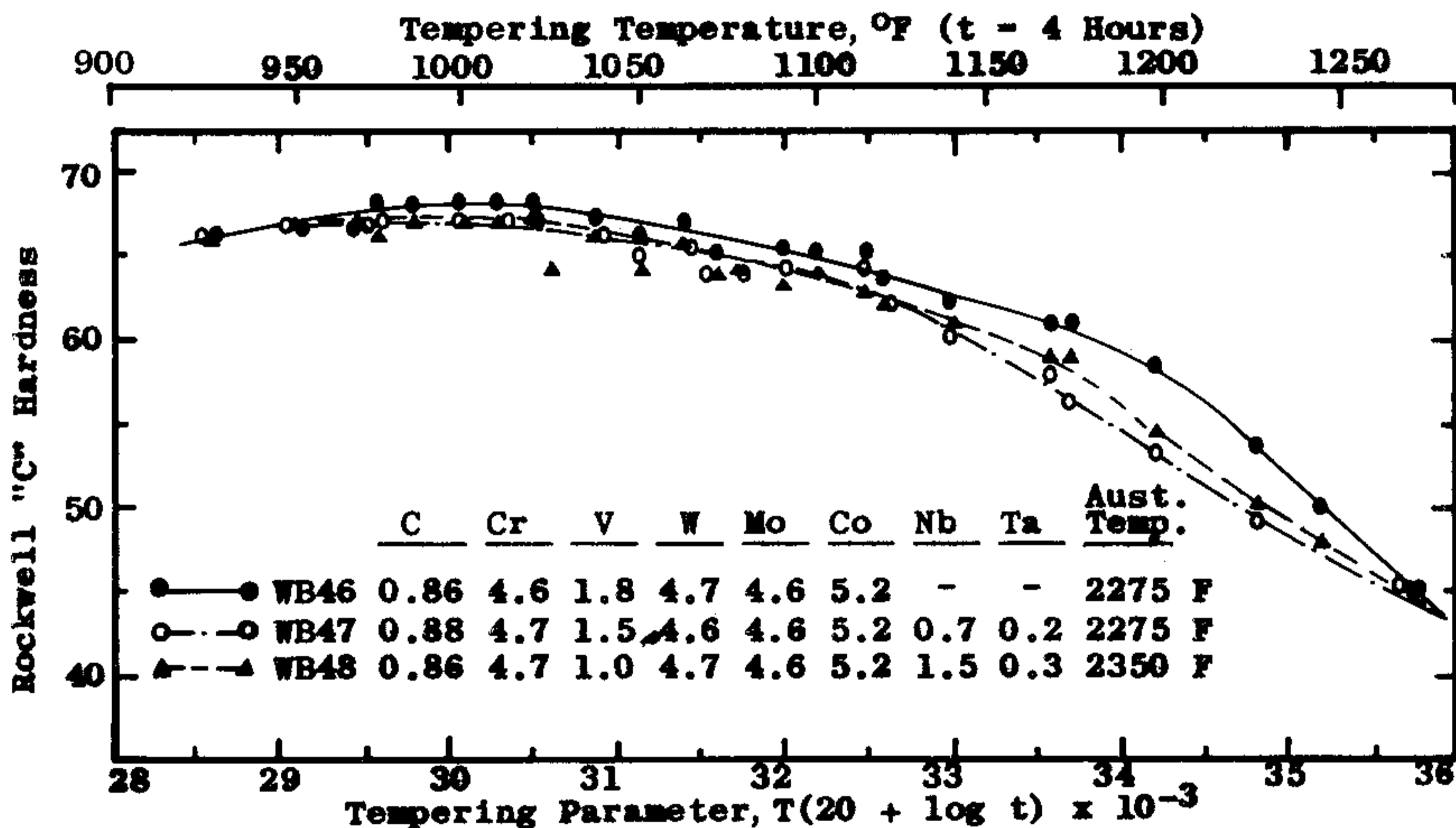


Figure 22. Master Tempering Curves for Steels WB46, WB47, and WB48.

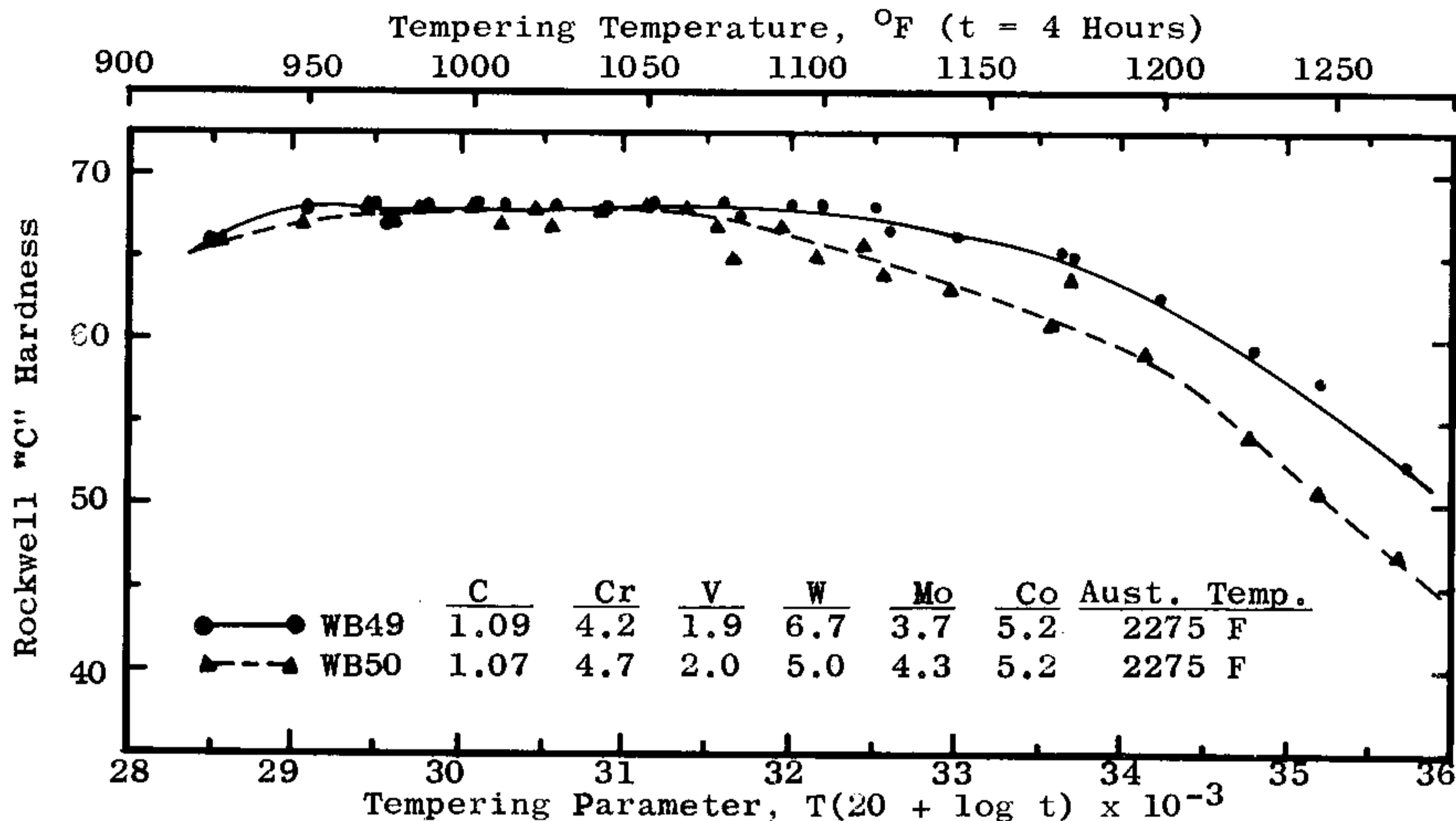


Figure 23. Master Tempering Curves for Steels WB49 and WB50

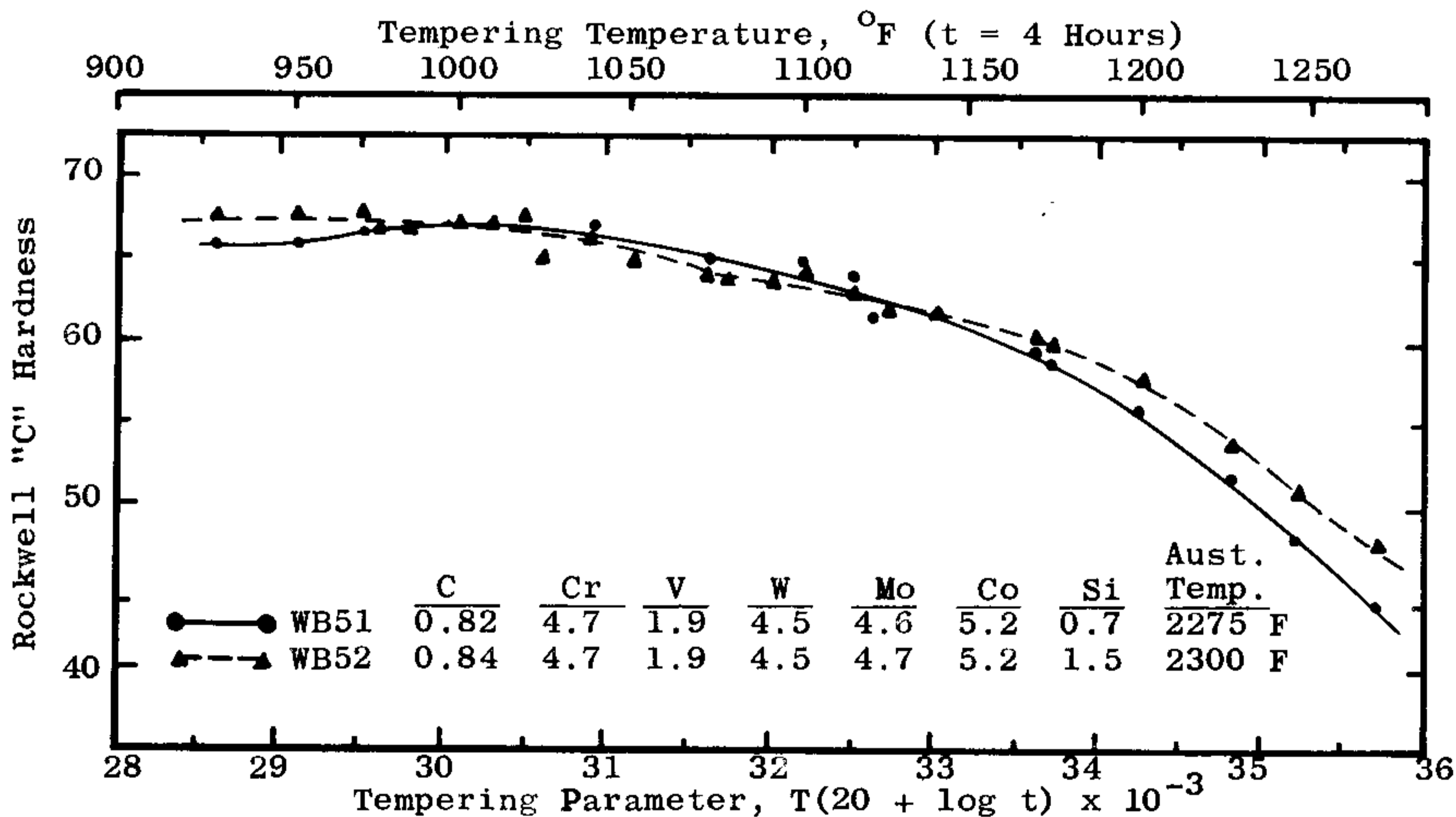


Figure 24. Master Tempering Curves for Steels WB51 and WB52

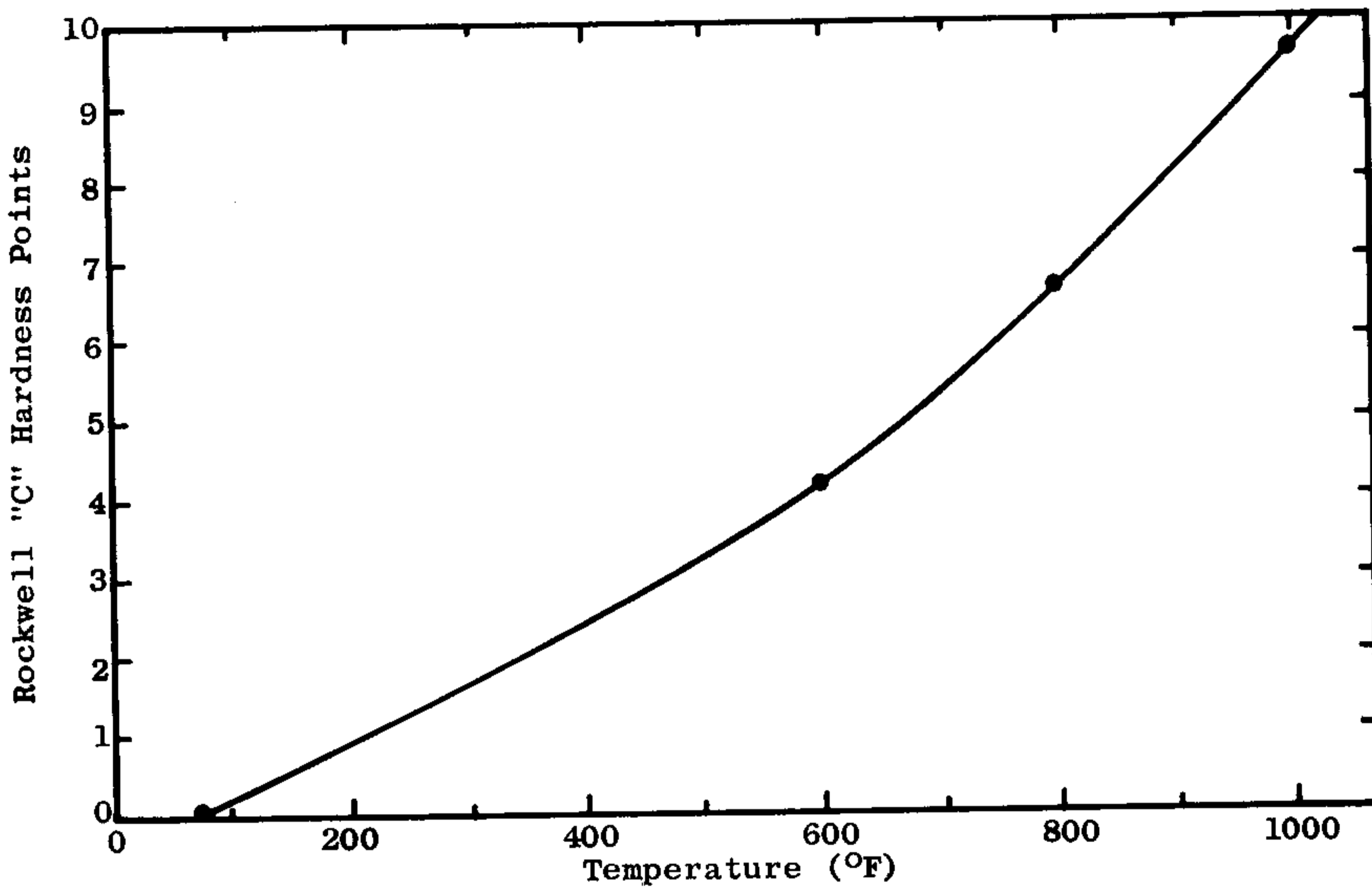


Figure 25. Average Hardness Drop From Room Temperature to 1000 F (For Steels Heat-Treated to R<sub>C</sub> 65 to 69 at Room Temperature)

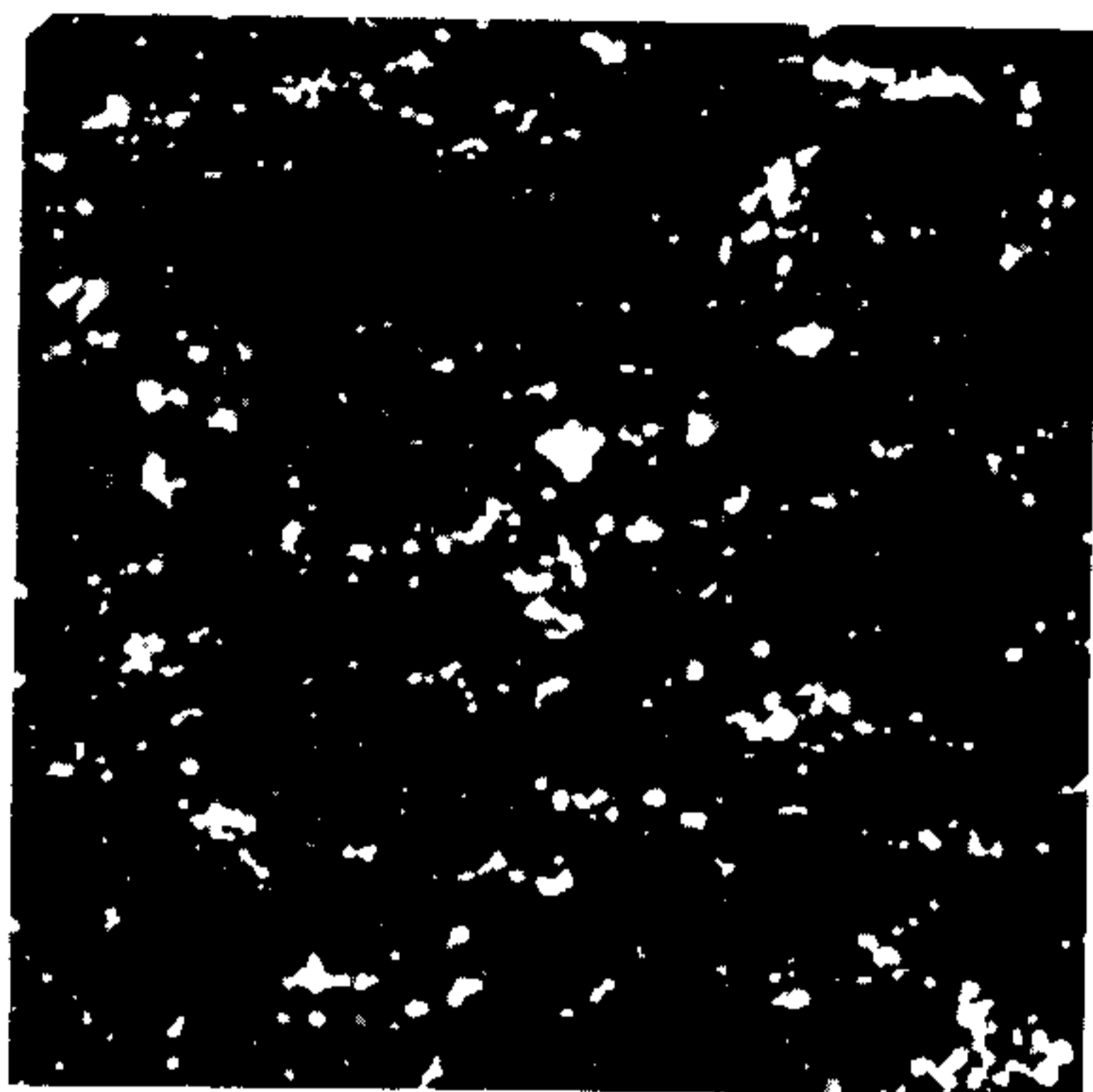


Figure 26. Transverse Micro-structure of Steel WB18 showing carbide size and distribution (austenitize 2225 F, oil quench, temper 2 + 2 hours at 1050 F)  
Etchant: 5% Picral + 1 drop HCl (750X)

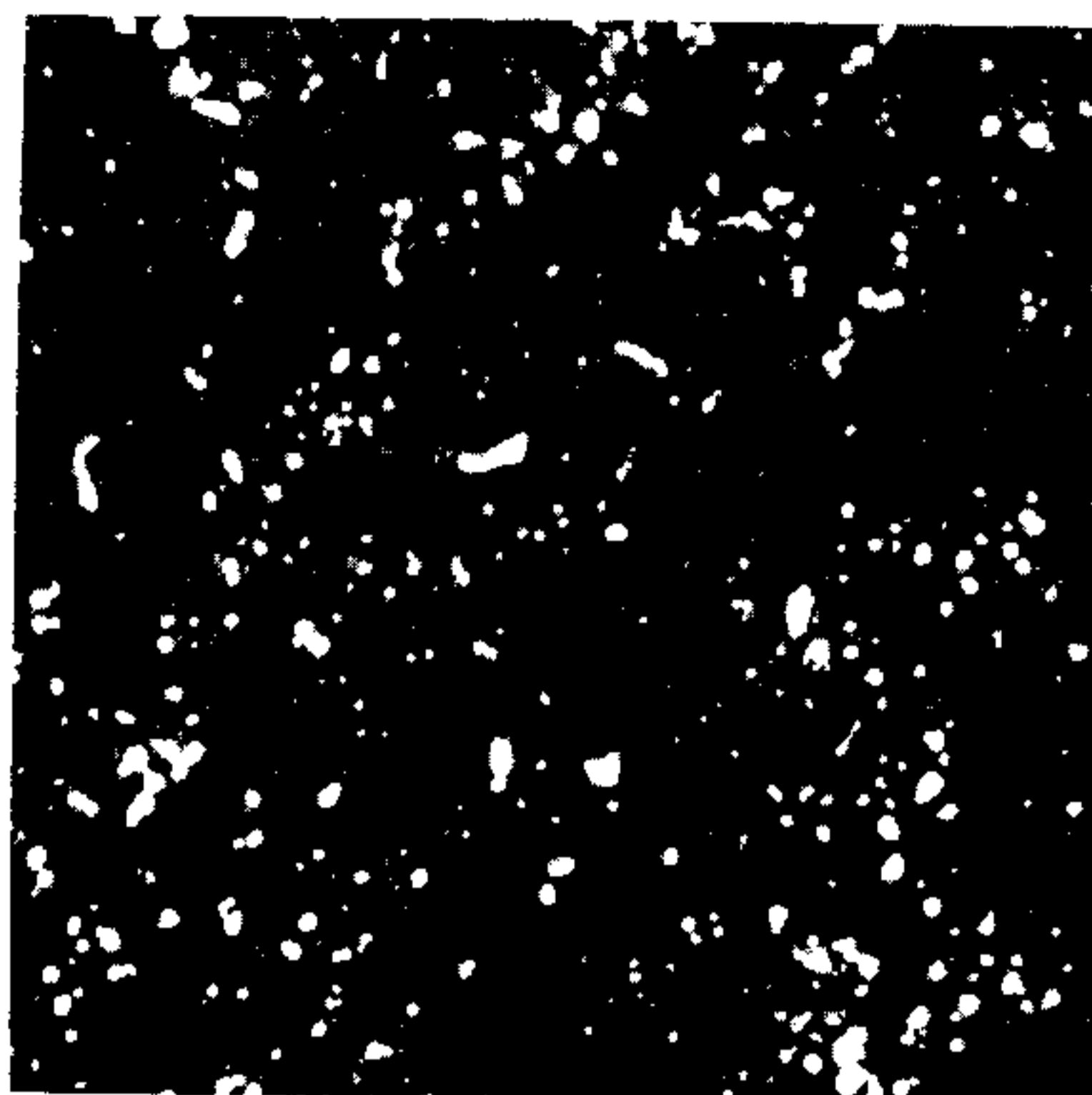


Figure 27. Transverse Micro-structure of Steel WB21 showing carbide size and distribution (austenitize 2225 F, oil quench, temper 2 + 2 hours at 1050 F)  
Etchant: 5% Picral + 1 drop HCl (750X)

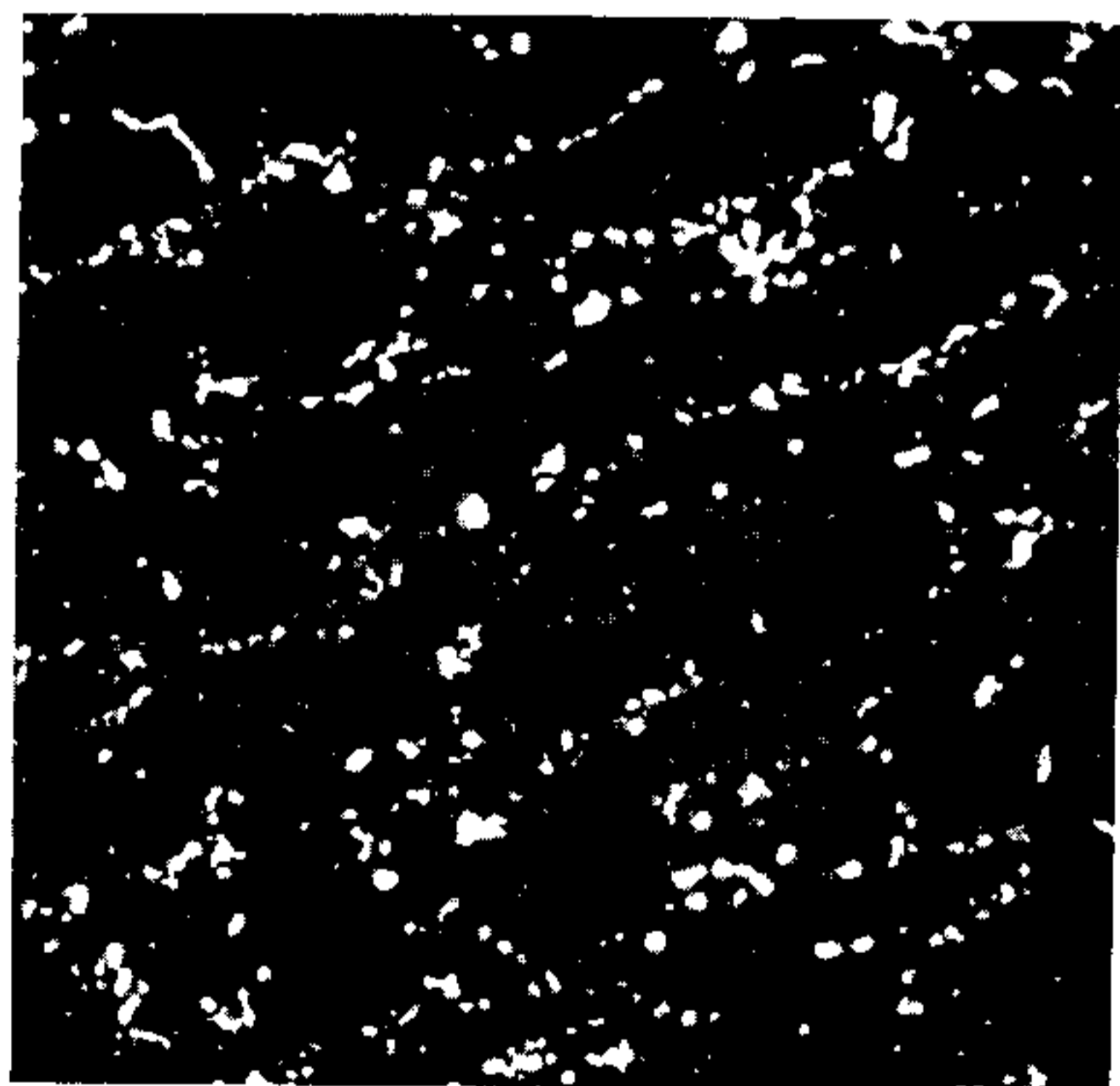


Figure 28. Transverse Micro-structure of Steel WB25 showing carbide size and distribution (austenitize 2225 F, oil quench, temper 2 + 2 hours at 1050 F)  
Etchant: 5% Picral + 1 drop HCl (750X)

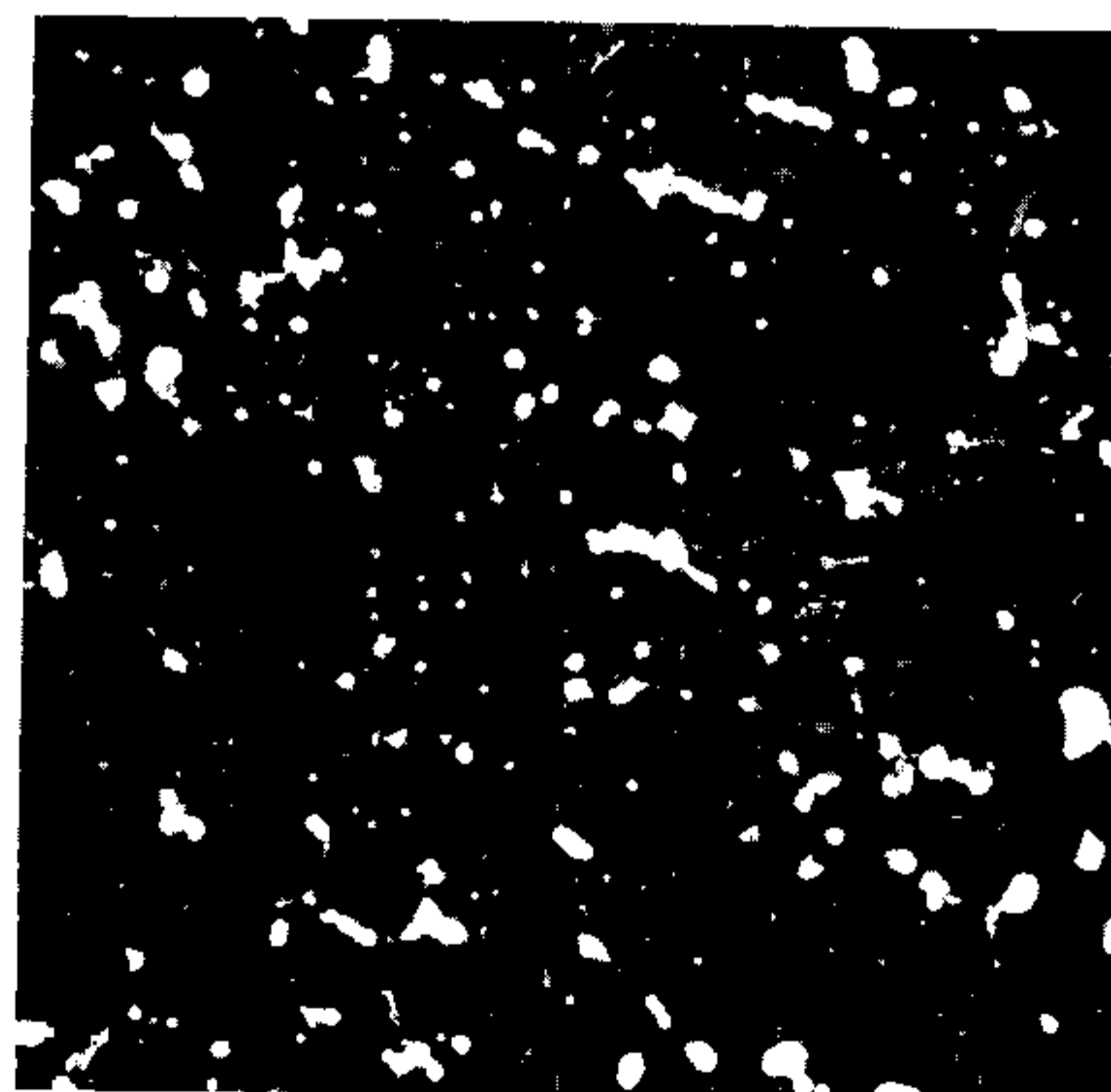


Figure 29. Transverse Micro-structure of Steel WB39 showing carbide size and distribution (austenitize 2225 F, oil quench, temper 2 + 2 hours at 1050 F)  
Etchant: 5% Picral + 1 drop HCl (750X)

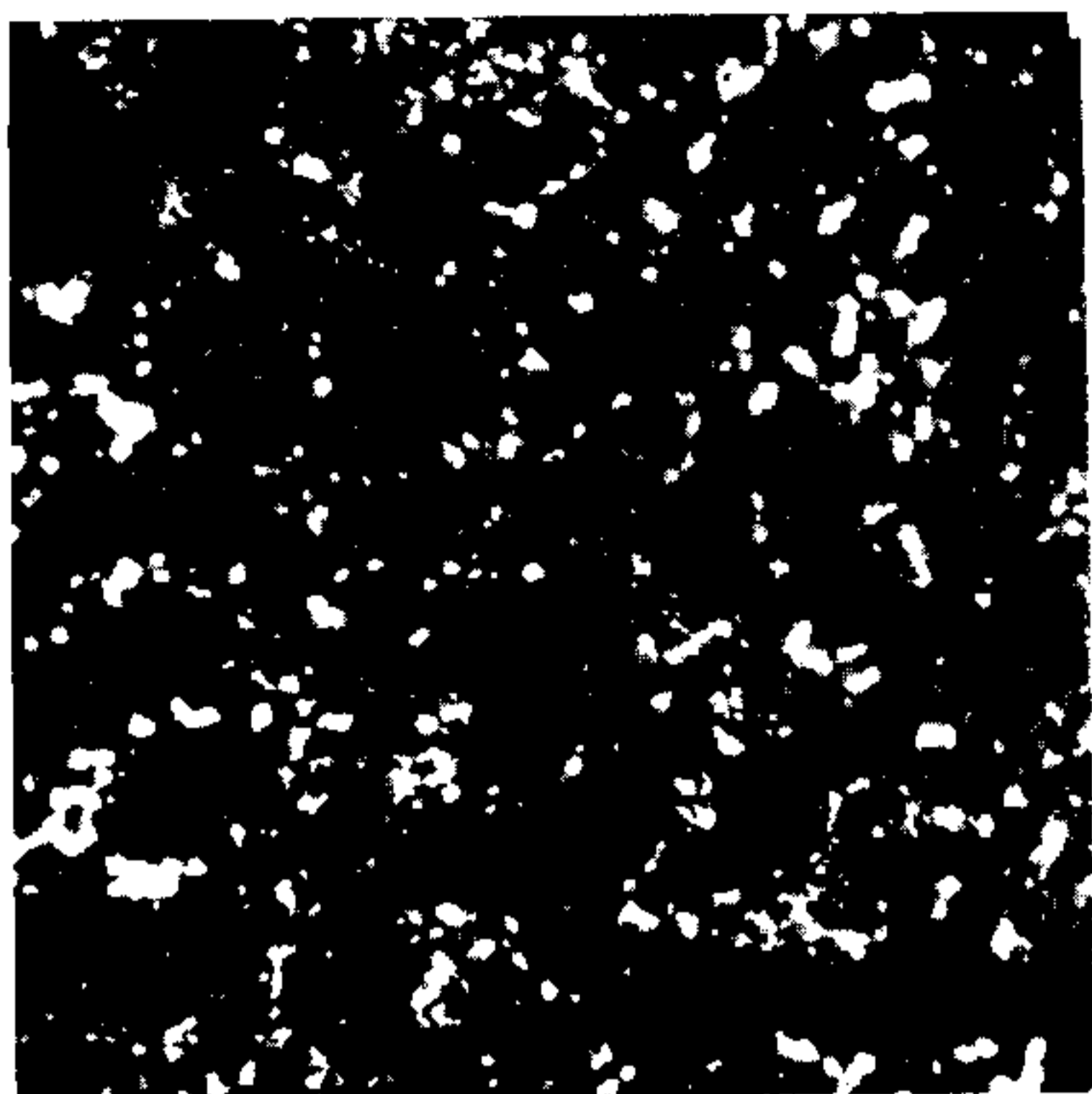


Figure 30. Transverse Micro-structure of Steel WB41 showing carbide size and distribution (austenitize 2225 F, oil quench, temper 2 + 2 hours at 1050 F)

Etchant: 5% Picral + 1 drop HCl  
(750X)

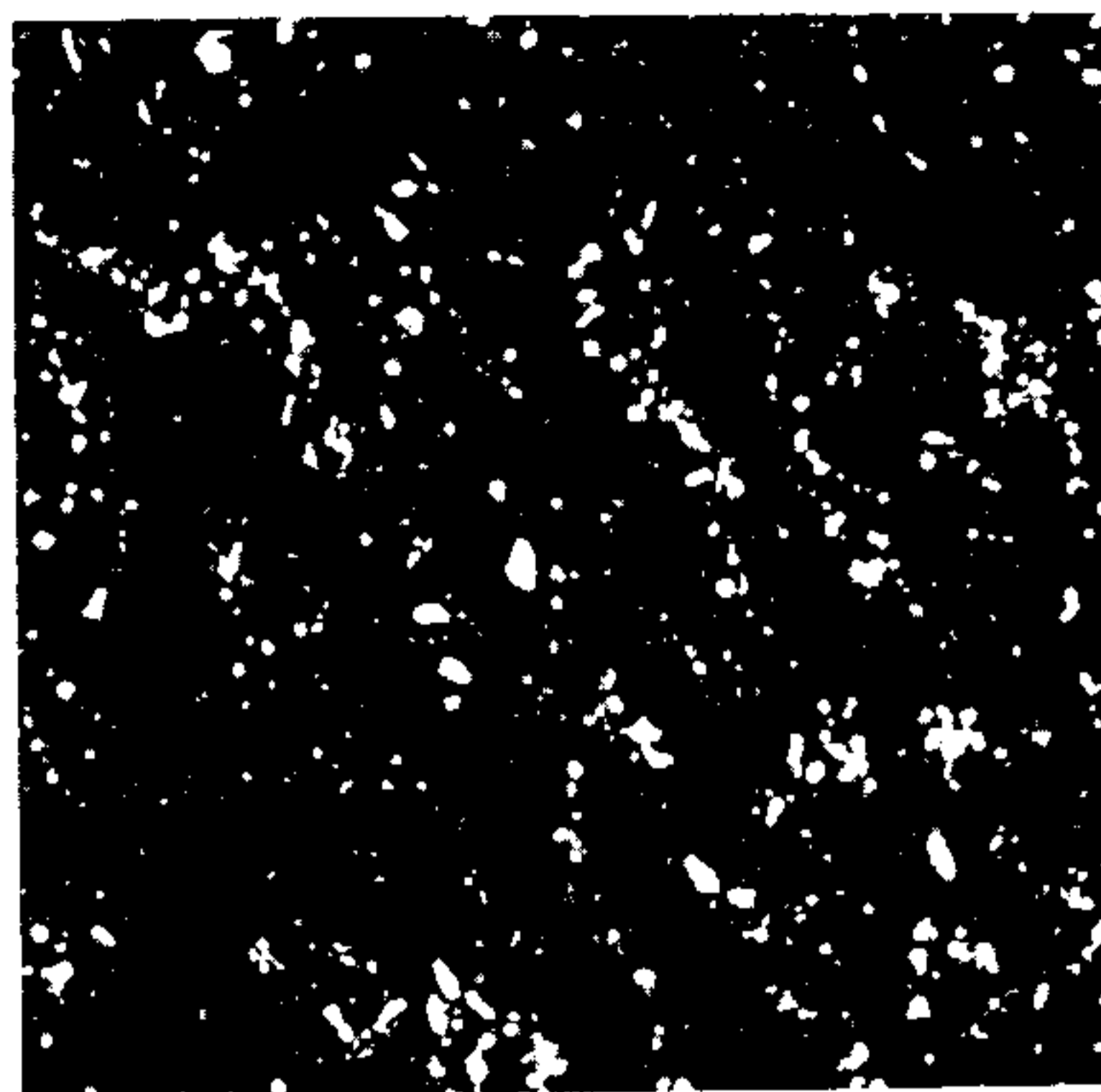


Figure 31. Transverse Micro-structure of Steel WB44 showing carbide size and distribution (austenitize 2225 F, oil quench, temper 2 + 2 hours at 1050 F)

Etchant: 5% Picral + 1 drop HCl  
(750X)

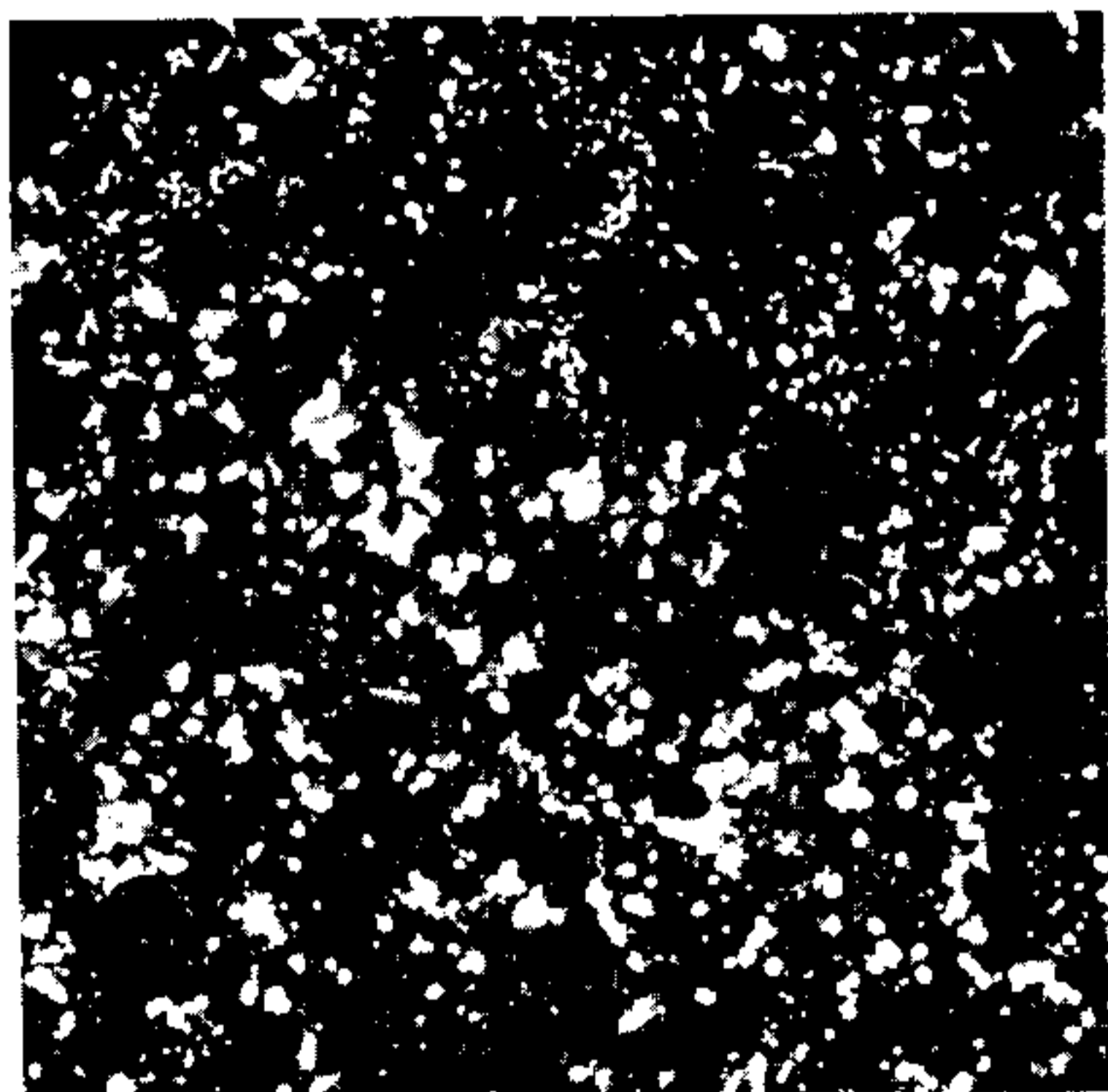


Figure 32. Transverse Micro-structure of Steel WB45 showing carbide size and distribution (austenitize 2225 F, oil quench, temper 2 + 2 hours at 1050 F)

Etchant: 5% Picral + 1 drop HCl  
(750X)

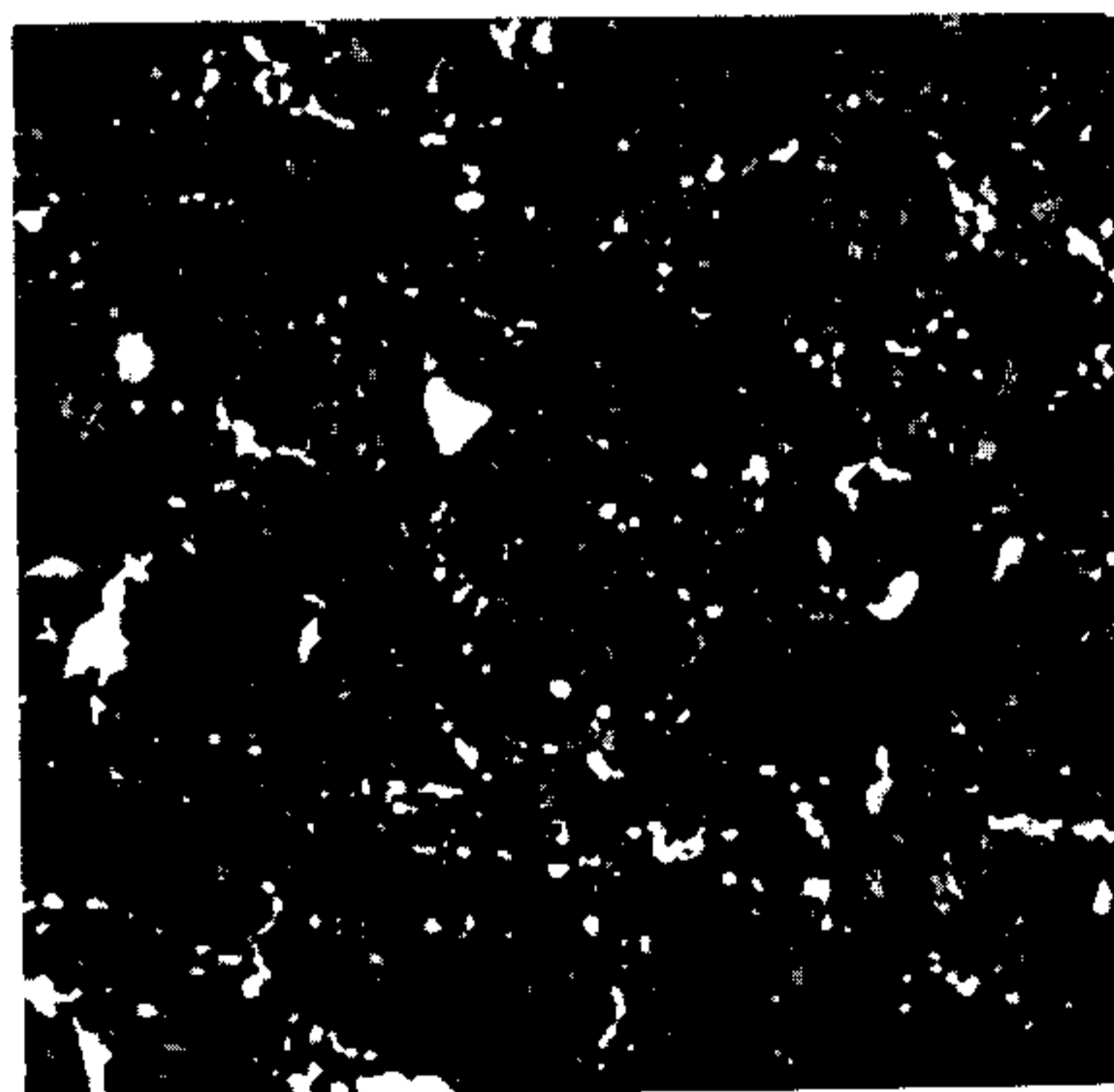


Figure 33. Transverse Micro-structure of Steel WB49 showing carbide size and distribution (austenitize 2225 F, oil quench, temper 2 + 2 hours at 1050 F)

Etchant: 5% Picral + 1 drop HCl  
(750X)