SOME PROPERTIES OF STEEL.

by

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INTRODUCTION

The metallography of steel has suffered greatly in the past from the accumulation of inaccurate data. One cause of this has been failure to employ chemically pure alloys. In this connection, Desch cites examples of elaborate and highly accurate researches being carried out on materials described as "copper wire," or "zinc in sticks." Although more attention is now paid to the chemical composition of the alloys, it can hardly be said that the inaccuracies introduced by chemical impurities have been eliminated. The great importance of the purity of the materials is universally recognised, but the difficulty of removing impurities has proved so great that the majority of investigations on steel are still performed with alloys containing an appreciable amount of manganese, silicon, phosphorus, and other elements, in addition to the true constituents of the alloy - iron and carbon.

In recent years, a mass of data on the physical properties of steel has been accumulated. Considerable care has been taken in selecting suitable materials, and in avoiding experimental inaccuracy; nevertheless, in/

in few cases is it possible to correlate the work of one investigator with that of another. The discrepancies cannot be attributed solely to difference in chemical composition .

The main objects of the present research were: (a) to obtain accurate data on the physical properties of steel in various conditions of heat-treatment, and to find the reason for the wide divergences between the data published by different workers: and (b) to investigate the phenomena of quenching and tempering by new methods.

In general terms, it may be stated that the lack of agreement between the work of different investigators appears to have been caused in many cases by failure to appreciate the great importance of slight variations of heat-treatment. Thus, results obtained by quenching from a rising temperature, are not always comparable with those obtained by quenching from a falling temperature; in hypereutectoid steels of low manganese content, the results are in no way comparable. Again, an account of the physical properties of annealed steels, given without detailed information on the method of annealing, is not much more valuable than a description of the properties of "zinc in sticks".

The investigation described in this paper, has been/

been divided into four sections: (a) electrode potential: (b) graphitisation: (c) electrical resistance: and (d) change of resistance during tempering. The results obtained are discussed separately in each section, and are considered collectively in the conclusions at the end of the paper.

HEAT
TREATMENT

Heat-Treatment.

The specimens were heated in an electric resistance furnace, which consisted of a transparent silica tube wound with platinum wire. The tube was evacuated by means of a Toepler mercury-pump; no difficulty was found in keeping it air-tight even at 1100°C. The temperature of the specimens was measured by means of a platinum/platinum-iridium thermo-couple, which was standardised from time to time against known melting-points.

The operation of quenching was performed in the following manner:-

The specimens were heated in the resistance furnace, in a silica boat such as is used in carbon combustions. One end of the boat was ground off, so that the specimens could slide easily from it into the quenching-bath; to the other end was attached a wire by means of which the boat could be withdrawn from the furnace. When the specimens were ready for quenching, the boat was pulled out of the furnace, and the specimens slid into the quenching-bath, which was placed directly below the end of the furnace tube. Less than one second was occupied in transferring the specimens from the furnace to the quenching-bath.

Except where otherwise stated, (a) the heat-treatment was carried out in vacuo: (b) the quenched specimens were heated to the quenching-temperature, soaked for half an hour, and quenched in a large volume of iced brine: (c) the annealed specimens were heated for one hour at 850°C., and cooled slowly in the furnace.

SECTION 1

ELECTRODE POTENTIAL

Of the many physical properties of metals and alloys that have been investigated with the aim of determining the constitution of alloys, perhaps the most promising but disappointing is the electrolytic solution pressure. This is especially true of steel, which appears never to have been thoroughly investigated in spite of the fact that if the electrode potential of this substance could be determined accurately, the results should be of the greatest interest. Unfortunately, the investigation of the electrode potential of alloys is complicated by many factors; in the case of steel, it is doubtful whether the true value could be determined experimentally. Steel consists of iron mixed mechanically with various substances, and having various substances dissolved in it; the result of this heterogeneity is that, on placing a steel in contact with an electrolyte, not only are local electrolytic actions set up between the constituents, but the specimen also reacts chemically with the solution, so that the composition of the specimen and of the electrolyte is not constant, and the potential alters with time.

Hatfield(1) determined the electrode potential of numerous alloys in strong and normal acids. He took readings of potential at regular intervals after immersion, and gave the initial and final readings. Owing to chemical action, the composition of the acid and of the surface of the specimen must have altered rapidly, so that the values he obtained do not represent the electrode potential of the alloys. It is only the surface of the specimen that is in contact with the electrolyte, and the potential of an alloy depends therefore on the composition of the surface, and this may alter so greatly, by gradual solution of the most active metal, that the alloy may show ultimately the potential of the most noble metal, whereas it may initially have shown that of the most active.

When determining the electrode potential of a pure metal, the specimen is immersed in a solution of one of its salts, but in the case of alloys it is necessary to employ electrolytes that contain definite proportions of salts of the component metals. Reinders(2) examined the relation that exists between the composition of an alloy and that of the electrolyte with which it is in equilibrium, and showed that the composition of the electrolyte could be calculated. It appears that if the components of the alloy have widely differing electrode-potentials, satisfactory results can be obtained by using a salt of the most active metal. This method was adopted by Herschkowitz,(3) who studied non-ferrous alloys, and found that electrodes that are composed of two or more metals can be classified under three headings:-

Class I : The metals form a mechanical mixture. The potential is that of the most active metal. As an illustration of this fact, it may be mentioned that Laurie(4) soldered small pieces of zinc into holes in a copper plate, and found that the E.M.F. of the combination was the same as that given by zinc alone.

It has been shown that if a metal exists in several allotropic modifications, each of these has its own solution pressure. Thus, alpha iron and gamma iron should each have a definite electrode potential. However, if these two modifications are mutually insoluble, the potential of a mixture of the two must be that of the more active form. In quenched steels, both forms may be present; but if martensite is more active than austenite, then the electrode potential of a quenched steel must be that of the martensite, and cannot be directly affected by the presence of austenite unless the steel is completely austenitic. This point will be referred to again.

Class II : The metals form a solid solution. The concentration-potential curve has a logarithmic form, the solid solution being always less active than the most active component.

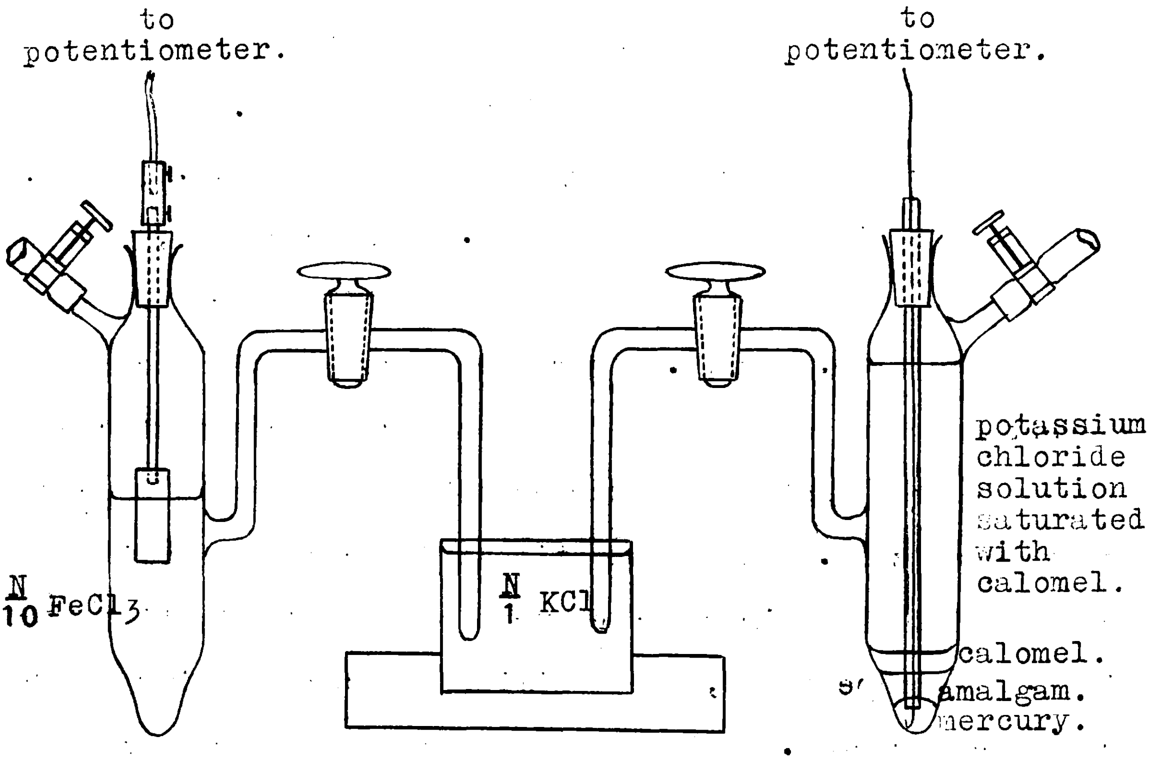
Class III : An intermetallic compound is formed. If the compound can exist as such in solution, it has its own solution pressure. On the concentration-potential curve, intermetallic compounds are indicated by discontinuities.

Pushin (5) investigated the electrode potential of various binary alloys, and determined the composition of numerous intermetallic compounds. He used a salt of the more active metal, when the components had widely different electrode-potentials, but when the potentials of the components were similar, he used special electrolytes such as a weak acid that formed a sparingly-soluble salt with the more active metal. According to Reinders, in such cases a complex electrolyte is theoretically necessary, but Pushin found that the application of the laws given for the composition of the electrolyte presented great difficulties. It does not appear to be possible to provide an electrolyte with which steel would be in true equilibrium; but during the investigation described below, it was found that satisfactory results could be obtained by using a solution of an iron salt.

In carrying out a determination of the electrode potential of an alloy, the method generally adopted is to use the specimen, dipping into the electrolyte, as one element of an electrolytic cell, a standard electrode such as the normal calomel-electrode being employed in conjunction with it. The potential-difference between the electrodes is measured by a potentiometer. Now, if the single potential of the normal calomel-electrode is known, and this value is subtracted from that found for the combination, the E.M.F. due to the contact of the alloy with its electrolyte is obtained. This statement must be modified if the potential difference at the contact of the solutions is not negligible.

The single potential of the mercury electrode in contact with normal potassium-chloride solution saturated with calomel has been determined by several methods. The value commonly accepted for ordinary temperatures is +0.56 volt, as found by Lippmann and Helmholtz. More recently, however, Billitzer(6) has obtained a value that differs from the above by as much as 0.74 volt. In consequence of this uncertainty, Nernst has advised the use of a standard hydrogen electrode, and suggested that all electrode potentials should be referred to this as zero. The International Congress at Berlin(7) recommended that directly measured values of potential should in all cases be given, the auxiliary electrode being a normal calomel-electrode. This method of stating the results has been followed in the present investigation - that is to say, the values quoted in this paper are the direct readings obtained from the combination: steel specimen/electrolyte/normal KCl/calomel cell. These values include the potential difference at the junction of the solutions.

ELECTRODE
POTENTIAL
APPARATUS



EXPERIMENTAL.

The composition of the steels employed in the present investigation is shown in Table I. The specimens were in the form of cylinders ¾" long and ¼" diameter, with a hole at one end into which fitted a tapered steel rod, which supported the specimen inside the cell. The rod was fitted with a piece of thick tight-fitting rubber tubing which acted as a stopper for the mouth of the half cell. A diagrammatic sketch of the apparatus is shown opposite.

Two differently treated steels of different carbon content were first investigated, and many attempts were made to get definite results, capable of repetition, for each. As has been stated above, electrode potential is a surface effect, and for this reason the surface of the specimen was always cleaned with polishing paper before immersion in the solution. The electrolyte originally used was a normal solution of ferric chloride.

The specimen was partly immersed in the electrolyte, the apparatus was connected up, and the electrode potential was read by means of a potentiometer. It was found that the potential altered steadily, falling in half an hour to a minimum from which it increased gradually without attaining a constant value even after four hours. It is the general practice among investigators of the electrode

electrode potential of alloys to take a final constant reading as the value for the alloy. There are numerous objections to this procedure, some of which have been mentioned in the preceding pages. It will be obvious, however, that any reading taken after a specimen has been for hours in contact with a solution with which it is not in equilibrium, will be of no value..

It was thought that the minimum value of the E.M.F. might be of significance, but this did not prove to be the case, as it could not be repeated for the same specimen.

A coating, apparently of ferric hydroxide, was found to form on the specimen, and this coating affected the results. Thus, if a specimen was partly immersed in the solution and allowed to remain till the reading was fairly constant, an immediate alteration of the E.M.F. was caused by pushing the specimen further in. This appeared to indicate that creeping of the electrolyte up the sides of the specimen would affect the results, but when the supporting rod and the upper part of the specimen were insulated by means of rubber-tubing and vaseline, and the specimen totally immersed in the electrolyte, the results were not improved,

Dilute ferric-chloride solution still gave the same deposit of ferric hydroxide, and the results were unsatisfactory. Sometimes the E.M.F. fell to a minimum and rose again - sometimes it commenced by rising and then fell.

Ferric nitrate was then tried but proved unstable. Dilute ferric sulphate gave the same indefinite results. In this case, bubbles of gas were observed to form quickly on the specimen, and it was thought that the failure to obtain definite values might be due to polarization. A rotating electrode was therefore employed, but this resulted in fluctuations of the spot of light, and the same general type of time-potential curve was obtained.

Ferrous solutions were tried, but the results were not less erratic. Moreover, the necessity of obtaining a ferrous solution free from ferric ions proved troublesome.

It was decided to persevere with dilute ferric chloride, and attention was paid to initial readings. The specimen was dipped into the electrolyte, and the E.M.F. noted immediately; this was repeated after cleaning the specimen. The readings varied considerably. Some improvement was caused by allowing the solution to come to rest before taking the reading, but the variation was still as large as 0.005 volt. It was found, however, that the condition of the surface had an even greater effect than had been believed, for, by polishing the specimen evenly with a fixed grade of paper, more consistent results were obtained. The sides were difficult to polish; so it was decided to insulate the sides, and polish the bottom evenly on 0$\cup$0 polishing-paper. Vaseline was used to effect the insulation. It was found that, though the results given by this method were fairly satisfactory for quenched steels, the accuracy was not sufficient to meet the case of annealed steels, which require an accuracy of one millivolt.

A close examination of the procedure showed that the main trouble lay in the protection of the sides from the solution by means of the vaseline coating. If this was too thin, the readings proved inconsistent; on the other hand, if it was too thick, bubbles of air were trapped at the foot of the vaseline, and these gave rise to erratic results. It was decided to avoid vaseline, and the sides of the specimen, as well as the bottom, were carefully polished on 000 paper, but this was a laborious process, and did not prove satisfactory.

Several other methods of protecting the sides were tried, but great difficulty was met with in preventing bubbles of air from being trapped at the bottom of the specimen. Finally, it was found that consistent results could be obtained by the method detailed below.

Method Employed for Determining the Electrode Potential of Steel.

The specimen was fitted to a rod similar to that described above, and the sides were carefully coated with shellac. When sufficiently dry, the specimen was inserted into the electrolyte (tenth-normal ferric-chloride) which, together with the calomel cell and the connecting link of potassium-chloride solution, was kept at 18°C. by means of of a simple thermostat. The apparatus was connected up as shown in the sketch facing page 17 .

Each step of the operation was timed, so that 15 seconds after inserting the specimen into the cell, the circuit was completed by depressing the potentiometer key, and the initial value of the potential was noted approximately. The specimen was then cleaned, polished on 000 paper, varnished, and again immersed in the electrolyte, the potentiometer being set at the value obtained in the first experiment. If any movement of the spot of light was noted, the potentiometer was re-set, and the operation repeated, until on depressing the key, no movement of the spot was observed. This value was considered to be the electrode potential of the specimen.

The direct readings for the series of carbon and chromium steels, quenched at different temperatures, are given in Table 2, and from these, Figures 1 to 7 have been plotted.

Consideration of Results.

As far as the writer is aware, no systematic investigation of the electrode potential of steel has been published; it may be advisable therefore to describe the general meaning of the results before proceeding to a detailed discussion of the curves.

In the introduction, it was stated that the electrode potential of a mixture is that of the most active component. From this it follows that the electrode potential of a solid solution that is not homogeneous, will be determined by the most active area of the solid solution exposed to the electrolyte.

The values of electrode potential given in this paper are the direct readings of the combination: steel specimen/decinormal ferric-chloride/normal potassium-chloride/calomel electrode. From any of the curves for quenched steels (Figures 1 and 2), it will be seen that the electrode potential decreases as the carbon content increases. The maximum value of electrode potential obtained was 0.517 volt; this is below the value generally accepted for the single potential of the/

the calomel cell, 0·56 volt. It appears that in all cases the quenched specimen acts against the calomel electrode; under similar conditions, zinc acts in the same sense as the calomel electrode. This means that with addition of carbon, iron becomes less like zinc, that is less active*; consequently, the most active area of a quenched steel that is not homogeneous, will be the area that is least concentrated in carbon. As the electrode potential of an alloy is determined by the most active area, it follows that the electrode potential of a martensitic steel must be that of the least concentrated area of martensite. This reasoning was confirmed by immersing two quenched steel-specimens, one of higher carbon content than the other, in a beaker containing deci-normal ferric-chloride solution, and noting the potential-difference between the specimens. It was found that the specimen of lower carbon content was negative to the other; the experiment was repeated with several pairs of specimens, the same result being obtained in each case. Under similar conditions, zinc is negative to copper. This again shows that the lower the/

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* This is in agreement with the rule given by Herschkowitz that a solid solution is always less active than the most active component.

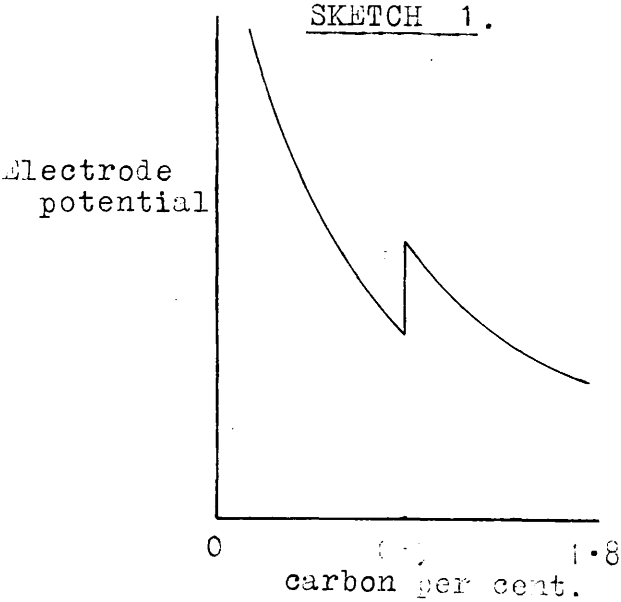
the concentration of carbon in the martensite, the more active the martensite is; and from this, that the electrode potential of a martensitic steel is that of the least concentrated area.

A further deduction from the curves is that austenite is less active than martensite, and that, in consequence, the electrode potential of a partly-austenitic steel is that of the least concentrated area of martensite, and is not directly influenced by the presence of austenite. To make this point clear, let us consider for a moment the electrode potential of the series of carbon steels quenched from 1100°C. There are three possibilities: (a) that the electrode potential of all the specimens is determined by austenite: (b) that up to a point, the electrode potential is that of martensite, and beyond that point, of austenite: (c) that in all cases the electrode potential is determined by martensite.

<u>Case (a)</u>. If we are to assume that the electrode potential of each specimen is due to austenite, then we must assume that all the quenched specimens contain that constituent. Although Enlund (8) has found evidence of the presence of austenite in steels of comparatively low carbon content, it is unlikely that the specimen containing 0.22 per cent. of carbon, quenched from 780°C., could contain that constituent.

More/

SKETCH 1.

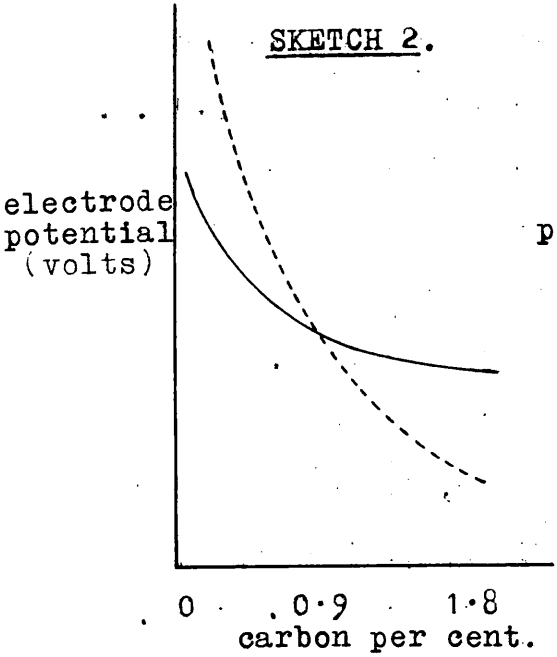


More conclusive evidence against this assumption is to be found in the investigation of the electrode potential of tempered steels carried out by Mr. J.M. Robertson. The potential-concentration curves of the quenched steels rose progressively as the tempering-temperature was increased, indicating a steady decrease in the concentration of carbon in solution. There was no discontinuity corresponding to the disappearance of austenite. From these results, it may be deduced that the electrode potential is determined by the concentration of carbon in solution in alpha iron.

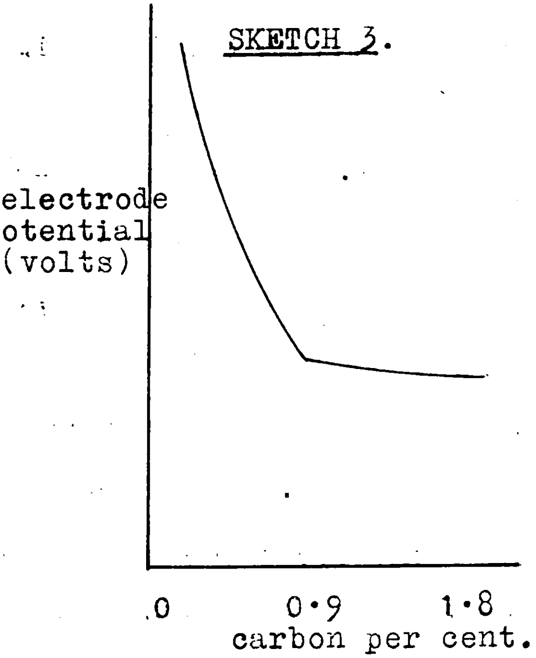
Case (b). First Alternative: Austenite may be more active than martensite, but may only be present in steels containing more than a certain percentage of carbon. If that were so, the potential-concentration curve of the series of quenched steels would be of the form shown in Sketch 1 opposite. There would be a break in the curve - a sudden increase of electrode potential - at the point where austenite first appeared. No evidence of this was found in any of the curves.

Second Alternative: The potential-concentration curve of a series of quenched steels may be a combination of two intersecting curves, one branch corresponding to martensite and the other to austenite. This case is illustrated in Sketches 2 and 3 opposite. The dotted curve/

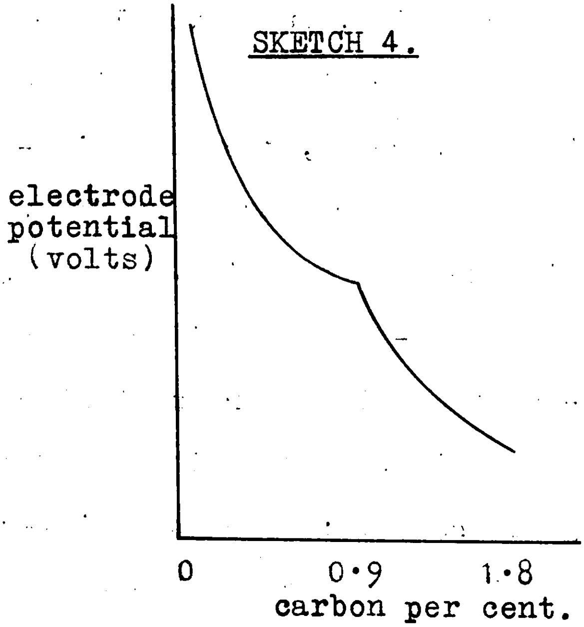
SKETCH 2.



SKETCH 3.



SKETCH 4.



curve in Sketch 2 may be considered to represent the electrode potential of martensite of varying carbon content, the curve drawn in full line being the corresponding curve for austenite (no difference in the reasoning is caused by assuming the position of the curves to be reversed). The electrode potential of a steel that contains both austenite and martensite is that of the more active constituent. In Sketch 2, up to the point of intersection, martensite (represented by the dotted curve) is more active; beyond the intersection, austenite is more active. The potential-concentration curve of the series of steels would therefore be of the form shown in Sketch 3, the first branch being a part of the martensitic curve in Sketch 2, and the second branch a part of the austenitic curve. There would thus be a discontinuity at the point of intersection, and at that point the curve would become abruptly flatter. Actually there is a discontinuity in the curves of quenched steels (Figures 1 and 2), but it is always in a direction opposite to that shown in Sketch 2. A typical experimental-curve is shown in Sketch 4. At the discontinuity, the curve becomes abruptly steeper. If the first branch of the experimental curve is due to one phase (martensite), then the second branch cannot be due to a different phase (austenite)/

(austenite) unless the first phase ceases to exist at the eutectoid, where the discontinuity occurs; otherwise the second branch would necessarily lie above the continuation of the first branch (Sketch 3), and not, as was found, below it (Sketch 4). There can be no question of martensite ceasing to exist at the eutectoid.

To sum up this discussion: (a) The electrode potential of quenched steels is not in all cases due to austenite: (b) If the experimental curve stepped upwards, or became abruptly flatter at the discontinuity, the second branch of the curve might be due to austenite; as the curve actually becomes abruptly steeper, the second branch cannot be due to austenite. There remains the third possibility - that in all quenched steels, the electrode potential is that of the martensite, and is not influenced directly by the presence of austenite. In what follows, this has been assumed to be correct.

The final conclusion is that in any quenched steel, the electrode potential is that of the least concentrated area of martensite, and is not affected directly by the presence of austenite in the specimen. The break at the eutectoid point will be discussed below.

Quenched Steels.

From Figures 1 and 2, it will be seen that raising the quenching-temperature causes an increase in the electrode potential of hypoeutectoid steels, and a decrease in the electrode potential of hypereutectoid steels; there is a discontinuity at approximately the eutectoid composition in all the potential-concentration curves of quenched steels. It may be mentioned that as only three of the specimens in the series of carbon steels contained between 0·7 and 1·2 per cent. of carbon, the position of the discontinuity in the curves may not be exactly at the eutectoid concentration as shown in Figure 1. Figures 1 to 7 were constructed by drawing smooth curves through the values of electrode potential obtained. It is clear however that in all the curves, the discontinuity must be close to the eutectoid composition.

Hypereutectoid Steels.

The decrease in the electrode potential of hypereutectoid steels with rising quenching-temperature is in accordance with the iron-carbon diagram. As the quenching-temperature is raised, the solubility of carbon in gamma iron increases, and more concentrated martensite is produced on quenching. The marked decrease of electrode potential caused by raising the quenching-temperature from/

from 1000°C. to 1100°C. indicates that the martensite formed by quenching from 1100°C. is much more concentrated than that obtained by quenching from 1000°C. - the difference being greater than would be deduced from the iron-carbon diagram. Two explanations of this anomaly may be suggested: (a) that the SE line of the diagram is more strongly curved than is generally believed, the solubility of carbon in gamma iron at 1100°C. being therefore much greater than at 1000°C.: or (b) that so long as particles of proeutectoid cementite remain undissolved at the quenching-temperature, they may act as nuclei, and cause partial precipitation of carbide even with rapid cooling.

Hypoeutectoid Steels.

When hypoeutectoid steels are heated through the temperature of the critical point Ac_1, areas of austenite of the eutectoid composition are formed; and as the temperature rises, the excess ferrite dissolves in the austenite. When the temperature reaches Ac_3, the last areas of ferrite are absorbed by the solid solution, and the steel consists entirely of austenite. It might reasonably be expected that at the temperature of the upper critical point the austenite would not immediately attain homogeneity; but the further the temperature is raised above this point, the more rapidly should

equilibrium/

equilibrium be attained. As the electrode potential of a quenched steel is that of the least concentrated area, the effect of raising the quenching-temperature should be to decrease the electrode potential by causing diffusion of carbon into the areas that were originally ferrite. The curves in Figures 1 and 2 show that instead of a decrease of electrode potential, an increase of potential was caused by raising the quenching-temperature. This means that the concentration of carbon in the least concentrated area of these steels is reduced by raising the quenching-temperature. From this it follows that the martensite formed by quenching becomes less homogeneous as the quenching-temperature is raised. There are only two causes that could produce that result; either (a) the austenite from which the martensite is formed by quenching, becomes less homogeneous as the temperature is raised, or (b) the martensite formed from homogeneous austenite by quenching is not homogeneous - the higher the quenching-temperature, the greater being the heterogeneity of the resulting martensite.

(a) Andrew and Hay (9) have suggested that segregation of carbide at the boundaries of the austenite crystals occurs during soaking at a temperature above Ac_3, and that the segregation increases with increase of/

of temperature.* As an increase in the concentration of carbon at the boundaries of the austenite grains would necessitate a decrease in the concentration of carbon in the interior of the grains, this suggestion would be in agreement with the results obtained in the present investigation. It would have to be assumed, however, that the segregation occurs to an appreciable extent at a temperature so low as 850°C.; and that the concentration of carbon at the grain boundaries is considerable - otherwise the concentration in the interior of the grains would not be materially affected. This explanation does not account for the discontinuity in the potential-concentration curves at the eutectoid composition.

(b) If the heterogeneity of the martensite in quenched steels is not due to heterogeneity of the austenite from which the martensite was formed, then homogeneous austenite must yield heterogeneous martensite on quenching. Hoyt (20) states that Hanemann has worked out a diagram showing the phase changes that take place in steel during quenching, and has proved martensite to be "not a single-phase material but a two-phase material/

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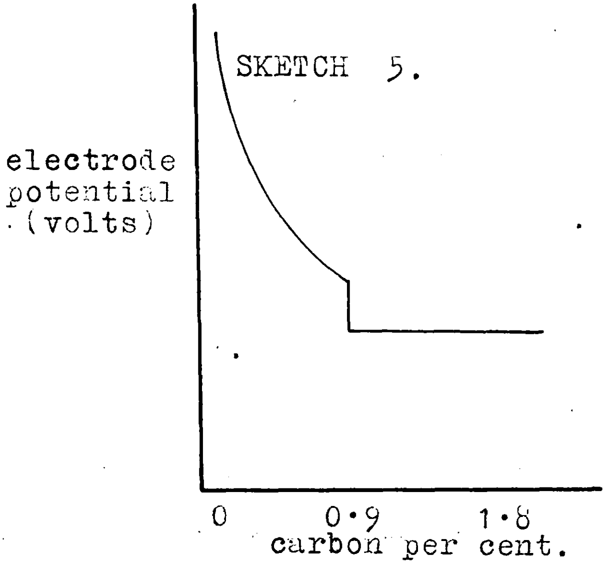
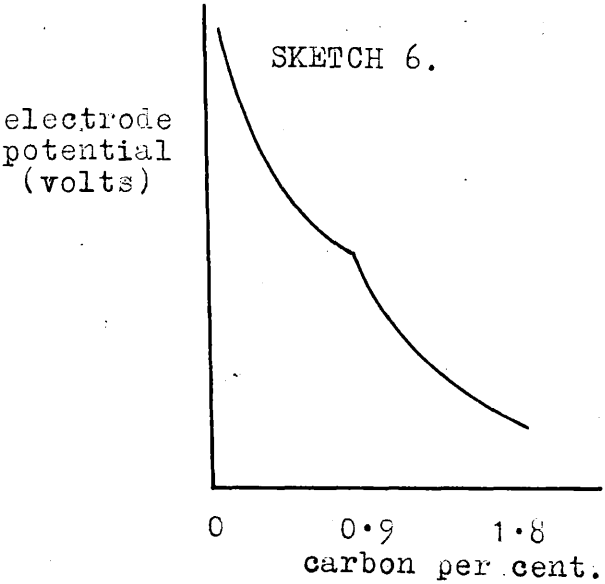
* Desch (42) considers that there may be a concentration of the solute at the boundaries of solid solution crystals.



SKETCH 5.



SKETCH 6.

material .... composed of a solid solution of iron with about 0.08 per cent. of carbon and a solid solution or compound containing about 0.9 per cent. of carbon". According to this theory, in any quenched steel containing less than 0.9 per cent. of carbon, there should be areas containing 0.08 per cent. of carbon; if this were so, the potential-concentration curve of a series of quenched hypoeutectoid-steels would be a straight line parallel to the composition axis. The curves in Figures 1 and 2 show that this is not correct. If the composition of the phase of lower carbon content varies as the percentage of carbon in the steels increases, and the second phase contains 0.9 per cent. of carbon, the potential-concentration curve of a complete series of quenched steels would be of the form shown in Sketch 5 opposite. There would be a discontinuity at the eutectoid, where the first phase ceases to exist; beyond the eutectoid composition, the curve would be parallel to the composition axis, for, according to Hoyt, the martensite in hypereutectoid steels contains 0.9 per cent. of carbon, the remaining carbon being present in the austenite. In order to explain the form of the experimental potential-concentration curves (Sketch 6, opposite), it would be necessary to assume that the second solid-solution also is/

is of variable composition.

Thus the results obtained by the writer may be explained by assuming that martensite is a two-phase material, but both of the phases must be assumed to be of variable composition. The less concentrated phase must increase in concentration as the percentage of carbon in the steel increases, its concentration must be decreased by raising the quenching-temperature, and it must cease to exist at the eutectoid composition; the second phase must increase in concentration beyond the eutectoid point. From this, it is clear that although the electrode-potential determinations do not disprove Hanemann's contention that martensite is a two-phase material, they indicate at least that his conception of the nature of the phase is incorrect.

One other explanation of the discontinuity at the eutectoid composition, and of the effect of varying the quenching-temperature must be considered. It has been shown (21) that when a metal is strained by cold-working, it is rendered more active - that is, its electrode potential is increased. From this, it might reasonably be expected that the electrode potential of a quenched steel would be affected by the strain to which the specimen was subjected during the quenching; the effect of/

of raising the quenching temperature would probably be to increase the strain, and therefore to raise the electrode potential. This suggestion would account for the variation of electrode potential caused by raising the quenching-temperature of hypoeutectoid steels (Figures 1 and 2). The break at the eutectoid point could be explained by assuming that the strain caused by quenching increases with the carbon content of the martensite and reaches a maximum at approximately the eutectoid concentration, decreasing thereafter due to the retention of austenite.

From Figure 3, it will be seen that when a series of quenched carbon-steels was allowed to remain at room-temperature, the electrode potential of all the specimens decreased with time. The fall of electrode potential could not be due to tempering of martensite, for precipitation of carbon from solution would cause a rise of electrode potential instead of a decrease. Nor could the fall of electrode potential be due to tempering of austenite. If the austenitic areas decomposed into martensite of greater carbon content than the martensitic areas that were originally present, the electrode potential of the specimen would be unaltered, for the electrode potential is determined by the least concentrated area of martensite; and if the austenitic areas/

areas decomposed into martensite of lower carbon content than that of the original martensitic areas, the electrode potential would be raised. Thus, if the reaction that occurs at room-temperature consisted of tempering either of martensite or of austenite, the electrode potential of the specimen might be raised, but could not decrease as it was found to do. The evidence appears to indicate that the fall of electrode potential at room-temperature is not due to tempering; it may be due to the release of the strain caused by quenching. Tammann (22) strained a piece of silver wire by cold-working, and found that the rise of electrode potential caused by the strain decreased in three days to one-third of its initial value. These experiments appear to support the suggestion that the electrode potential of quenched steel is considerably affected by the strain produced by quenching, and that the increase of electrode potential caused by raising the quenching-temperature may be due to the additional strain caused by the more drastic quenching. Against this, it must be stated that if the foregoing suggestion is accepted, then it must be assumed that the increase of electrode potential caused by strain is large. From Figure 1, it will be seen that the difference between the electrode potential of a carbon steel containing 0.71 per cent. of/

of carbon, quenched at 780°C., and that of the same steel quenched at 1100°C. is 6.0 millivolts; if the additional strain caused merely by varying the quenching-temperature is responsible for a rise of 6.0 millivolts, then the total increase of electrode potential due to the strain of quenching must be much more than 6.0 millivolts. Spring (21) investigated the increase of electrode potential caused by straining various substances, and found that with one exception the increase was less than 1.0 millivolt - the exception was bismuth, which gave a reversed effect. Turher and Jevons (23) found that the difference of potential between strained and unstrained steel was less than 0.1 millivolt.

Liquid-Air Treatment.

Two series of carbon steels, quenched at 1100°C. and 780°C. respectively, and a series of chromium steels quenched at 1100°C., were immersed in liquid-air for one hour, the electrode potential of each specimen being determined at 18°C. before and after immersion; the results obtained by this treatment are shown in Figures 4 and 5. From these Figures, it will be seen that no appreciable change of electrode potential was caused by cooling to the temperature of liquid-air. Microscopic examinations of the specimens showed that in/

in the steels of high carbon content, some austenite had been converted into martensite by immersion in liquid-air. The martensite thus formed was not less concentrated than the martensite originally present in the specimens; otherwise there would have been a rise of electrode potential. The curves do not show whether the martensite formed from the austenite was of the same concentration as the martensite originally present, or of greater concentration; for the electrode potential is that of the least concentrated areas of martensite, and could only be affected by the formation of areas of lower concentration than those initially present. The curves indicate however that martensite was not affected by immersion in liquid-air; for precipitation of carbon from solution in martensite would be accompanied by a rise of electrode potential.

Quenching from a Falling Temperature.

A series of carbon steels was heated to 1100°C. for 15 minutes, cooled slowly to 780°C. (8° per minute), soaked at that temperature for 5 minutes, and quenched in water; a second series was heated to 1100°C. for 15 minutes, cooled slowly to 850°C., soaked for 5 minutes and quenched. The potential-concentration curve of each series was determined. In Figures 6 and 7, these/

these curves are shown together with the curves obtained by quenching from the same temperatures (780°C. and 850°C. respectively) without previously heating to 1100°C. As will be seen from the curves, quenching after cooling to the quenching-temperature, had the same effect on the electrode potential of hypoeutectoid steels, as quenching after heating to the quenching-temperature. On the other hand, the potential-curves of the hypereutectoid steels that had been cooled from 1100°C. before quenching, lie below the curves obtained by quenching without first heating to 1100°C.; this indicates either that solution of cementite and diffusion through the austenite were not completed after soaking for one half hour at the quenching-temperature, or that precipitation of cementite, on cooling slowly from 1100°C. to the quenching-temperature and soaking for 5 minutes, was not completed. Possibly both of these causes contributed to the result. The methods of quenching discussed above, have been used indiscriminately by different investigators; the curves in Figures 6 and 7 show that the two methods do not give comparable results.

SECTION 2

GRAPHITISATION

In discussing the valuable results that have been attained by metallurgical research, Dr. Rosenhain justly states that the crowning achievement of physical metallurgy was the construction of the iron-carbon diagram. The many controversies through which the diagram has passed, have each contributed something to the form that is now generally accepted; and to-day there remain only a few points on which metallurgists are not in agreement. One of these concerns beta iron, which is still commonly regarded by American metallurgists as a separate phase, whereas the great majority of European workers consider it to be non-magnetic alpha iron. A second point on which agreement has not yet been reached is the relation of cementite to graphite; and, curiously enough, on this point also American metallurgists differ from the generally accepted European view. The latter view is summarised in the following paragraph, taken from Dr Hatfield's book on cast iron:-

As the molten hypereutectic alloy of iron and carbon "cools down to the freezing range, carbide is thrown out of solution, and simultaneously dissociates into iron and carbon with the production of 'kish.' The iron then freezes, the eutectic splitting up into austenite (solid solution) and carbide, which latter constituent, controlled by the prevailing conditions of rate of cooling and composition, either persists or dissociates. Further cooling causes the gradual precipitation of further free carbide, which behaves in a similar manner to the carbide separated at the eutectic point. When we arrive at the temperature of the pearlite change-point we still have the solid solution corresponding in composition to hardenite, which now resolves itself into carbide and free iron. The carbide will, again, either dissociate, with the production of perfectly soft pig-iron free from combined carbon, or will persist, and be recognised as pearlite in the final iron."

According to this theory, graphite (the stable phase) is always formed by the decomposition of cementite (the metastable phase); and for many years, this view has met with general approval. Recently, however, American investigators have carried out much research on the graphitisation of iron-cementite alloys, and have concluded that cementite does not break down directly into graphite. These workers (Storey, Archer, Mercia, and others whose work will be described later) still differ among themselves, but in general they consider that graphite may be deposited directly from solution.

The methods adopted in investigating the process of graphitisation have generally consisted of micro-examination and chemical analysis. Neither of these is satisfactory; for micro-examination gives results that are only qualitative, and though chemical analysis yields quantitative data, it involves the destruction, more or less complete, of the specimens, so that the same specimen cannot be used for several treatments. The writer found that specific volume determinations clearly demonstrated the presence of even a trace of graphitisation in steel; and in the hope of throwing some additional light on the relations of graphite and cementite, a series of carbon steels was subjected to various heat-treatments, the process of graphitisation being examined by means of specific-volume determinations.

As much of the previous work on this subject appeared to be of a somewhat haphazard nature, an attempt was made to carry out the investigation in a strictly systematic manner; but a large number of experiments had been completed before the great importance of one of the factors of the heat-treatment - the rate of cooling - was fully appreciated. It then became obvious that a thorough investigation would necessitate a lengthy research, and could not be satisfactorily carried out with the available apparatus. However, the results that had already been obtained were sufficient to suggest a hypothesis, and further experiments were performed to prove or disprove this.

METHOD EMPLOYED FOR DETERMINING SPECIFIC VOLUME.

The method employed for determining the specific volume of the specimens was based on that of Andrew and Honeyman,(11) but some modifications were introduced.

The specific gravity of a quantity of paraffin oil at 15°C. was accurately measured. The specific volume of an annealed carbon steel at 15°C. was then determined, by weighing the specimen in air and in the oil at that temperature. This specimen was used as a standard throughout the subsequent experiments. Like all the other specific-volume specimens, it was in the form of a thin section weighing 18 grammes.

The specific volume of each heat-treated specimen was measured by weighing the specimen in air and in oil, the standard being weighed in oil before and after the specimen. If the two values for the standard differed by more than ±0.2 milligramme, due to alteration in the temperature of the oil, the weighings were repeated. The specific volume of the heat-treated specimen at 15°C. was calculated from the weight in air and in oil, a correction being made for the temperature of the oil by means of the standard. In this method, the assumption is made that all the steels had the same temperature-coefficient of expansion; but as the maximum variation in the temperature of the oil throughout the experiments was less than ± 2°C., the error involved was negligible.

The weights employed were calibrated from time to time during the investigation. Repeated determinations of the specific volume of the same specimen showed that the experimental error was less than ±0.00001.

EXPERIMENTAL.

In order to avoid confusion, the research has been divided into twenty experiments. A descriptive number has been allocated to each heat-treatment, and the specific-volume results shown in Table 3 and plotted in Figures 8 to 14, correspond to these numbers.

In each experiment, a series of carbon steels was heat-treated - the composition of the specimens is shown in Table 1. In most cases, a separate series of steels was used for each heat-treatment, but in some instances one series was subjected to several consecutive treatments. This has been made clear in the description below, by giving a distinguishing letter to each fresh series of steels used, and retaining this letter for any particular series throughout the experiments which were performed with it.

EXPERIMENT 1. The specific volume of a series of carbon steels (Series A) was measured.

EXPERIMENT 2. Series A was heated to 850°C. for 1 hour, and cooled slowly in the furnace. The specific volume of each specimen was again determined. This treatment had practically no effect on the specific volume of the steels (See Table 3).

EXPERIMENT 3. Series A was reheated to 1100°C. for 15 minutes, cooled very slowly to 900°C., soaked at that temperature for 1 hour, and cooled slowly. This treatment produced a slight increase in specific volume in the case of the steel containing 1.19 per cent. of carbon, a considerable increase in the steels of higher carbon content, but no change in those containing less than 0.9 per cent. of carbon. Micro-examination showed the presence of graphite (minute specks of temper carbon) in the three high-carbon steels.

EXPERIMENT 4. Series A was reheated to 1100°C. for 15 minutes, cooled slowly to 850°C., soaked for 3 hours, and slowly cooled. The steels containing up to 0.9 per cent. of carbon showed no change; those containing above this percentage showed further graphitisation.

EXPERIMENT 5. Series A was reheated to 850°C., soaked for 3 hours, and cooled slowly. There was again no change in the steels containing up to 0.9 per cent. of carbon, but a considerable increase in the specific volume of those containing more than that percentage.

EXPERIMENT 6. Series A was reheated to 850°C., soaked for 3 hours, and cooled slowly. There was no change in the steels containing up to 0.9 per cent. of carbon. In order to avoid repetition, it may be stated now, that throughout all the heat-treatments to which the various series were subjected, no graphitisation was ever observed in the hypoeutectoid steels. A further increase in the specific volume of the steels containing more than 0.9 per cent of carbon. was found. It was evident from Experiments 5 and 8 that graphitisation, once initiated, would continue at a constant soaking temperature of 850°C.

EXPERIMENT 7. Series A was reheated to 850°C., soaked for 1 hour <u>in air</u>, and cooled slowly. In the hypereutectoid steels, this treatment caused a slightly greater increase in specific volume than had been produced in the previous experiment by soaking for 3 hours at the same temperature <u>in vacuo.</u>

EXPERIMENT 8. The specific volume of a second series of carbon steels (Series B) in the "as rolled" condition, was measured.

EXPERIMENT 9. Series B was heated to 850°C., soaked for 3 hours, and cooled slowly. There was a very slight decrease in the specific volume of the hypereutectoid steels.

EXPERIMENT 10. Series B was reheated to 850°C. for 3 hours, and cooled slowly. There was no change in the specific volume of the steels.

The results so far obtained, indicated that graphitisation was not readily produced by soaking at 850°C.; but that a short period of heating at a high temperature, followed by slow cooling and soaking at 850°C., did produce graphitisation; and that once the reaction started, it would continue at 850°C.

EXPERIMENT 11. The treatment to which Series A had been subjected was complicated by the fact that the specimens had been heated twice to 1100°C. A new series (Series C) was therefore measured "as rolled."

EXPERIMENT 12. Series C was heated to 1100°C. for 15 minutes, cooled slowly to 850°C., soaked for 3 hours, and slowly cooled. There was an increase in the specific volume of the hypereutectoid steels, but it was much smaller than that which had been produced in Experiment 3. In the latter case, similar specimens had been cooled from 1100°C. to 900°C., and soaked for

1 hour. From the results, it appeared that hyper-eutectoid steels graphitised more than twice as much at 900°C. in 1 hour, as in 3 hours at 850°C. It would certainly be expected that the temperature would influence the rate of the reaction, but the difference appeared to be too great to be due to this alone. Both series had been soaked at 1100°C. for the same length of time, and had been cooled slowly to the soaking-temperature. The rate of cooling had been roughly regulated by means of rheostats, but had not been carefully measured at the time. However, in carrying out the experiments, notes had been made of the actual time at which each stage of the heat-treatment commenced and finished, and it was therefore possible to calculate roughly the rate of cooling from 1100°C. to the soaking-temperature for each series. It was found that in Experiment 3, the rate had been less than 5°C. per minute, and that in the present case it had been more rapid - 8°C. per minute. It appeared therefore that the anomalous results might be due to the difference in the rate of cooling. As this would mean that the rate of cooling was of great importance, it was decided that in all future experiments, the rate of cooling should be as nearly constant as possible. The rate of 4°C. per minute was chosen.

EXPERIMENT 13. The specific volume of a fresh series of steels (Series D) as rolled, was measured.

EXPERIMENT 14. Series D was heated to 1100°C. for 15 minutes, cooled, at 4°C. per minute, to 850°C., and soaked for 3 hours. This treatment produced a much greater increase of specific volume than had been obtained with quicker cooling in Experiment 12, and actually a greater increase than had been obtained in Experiment 3 with the higher soaking-temperature.

It was now clear that to initiate graphitisation in hypereutectoid steels, an initial period of heating to a high temperature was effective - possibly necessary - and that the rate of cooling was an important factor.

EXPERIMENT 15. In order to find whether the duration of heating at the high initial temperature had an important influence on the results, a new series (Series E) was heated to 1100°C. for 1 hour, cooled to 850°C., and soaked for 3 hours. In two of the hypereutectoid steels, there was a slightly greater increase of specific volume than in Experiment 14; but it was apparent that the duration of heating at 1100°C. was not so important as the rate of cooling. In this and subsequent series, the specific volume of the specimens in the "as rolled" condition was not measured, as the previous experiments had shown that the steels were quite uniform.

EXPERIMENT 16. In order to find whether the results were affected by the temperature at which the soaking was carried out, Series F was heated to 1100°C. for 15 minutes, cooled to 950°C., and soaked for 3 hours. There was a slightly greater increase in specific volume than in Experiment 14, in which the soaking temperature was 850°C.

EXPERIMENT 17. Series G was heated to 950°C., and soaked for 6 hours. There was only a trace of graphitisation - especially noticeable in the steel containing 1.19 per cent of carbon.

EXPERIMENT 18. Series H was heated to 1100°C. for 15 minutes, cooled to 750°C., and soaked for 3 hours. The amount of graphitisation was similar to that obtained in Experiments 14 and 16.

EXPERIMENT 19. Series K was heated to 750°C. for 3 hours. There was no trace of graphitisation.

These experiments indicated that on soaking at a constant temperature, up to 950°C., graphitisation did not readily commence, but that once it was initiated, the reaction could proceed at a constant temperature. On the other hand, the results did not give any definite information on the relative rate of the reaction at different soaking-temperatures. This was probably due to the fact that in Experiments 14, 16, and 18, the conditions were not strictly comparable; for, although the soaking time was in each case 3 hours, account must be taken of the time occupied in cooling from 1100°C. Thus, in Experiment 18, the effective heating-time was 50 minutes longer than in Experiment 16. When allowance is made for this, it appears that graphitisation proceeds more rapidly at 950°C. than at the lower temperatures. This is confirmed by Experiment 17, in which a slight increase of specific volume was produced by soaking for 6 hours at 950°C. without previously heating to 1100°C., whereas similar treatment at 850°C. had had no effect.

EXPERIMENT 20. Series L was heated to 950°C. for 3 hours, cooled to 850°C., and soaked at that temperature for 3 hours. All the hypereutectoid steels showed an increase of specific volume. The steel containing 1.19 per cent. of carbon was graphitised to a greater extent than those containing 1.61 and 1.73 per cent. - a result that had not been produced by any previous treatment.

Before passing on to the next section, it may be as well to refer again to the importance of the rate of cooling, and to admit that with an electric resistance furnace and hand-controlled rheostats, it is not possible to regulate the rate of cooling with any great accuracy. For this reason, the various heat-treatments are not truly comparable. They are, however, sufficiently accurate to substantiate the hypothesis below.

Consideration of Results.

During the description of the experiments, it was pointed out that the results showed that on soaking at a constant temperature (up to 950°C.), graphitisation did not readily commence; and that, in order to start the reaction, it was desirable to heat to a high temperature, cool slowly, and soak at some temperature above A_1. Evidently, graphite nuclei are formed at the high temperature, or during cooling, and once these are present graphitisation can proceed at a constant soaking temperature.

So far, the results might no doubt be considered to be consistent with the direct decomposition theory; but the great importance of the rate of cooling requires some other explanation. It is enlightening to consider Experiments 9-10, 17, and 20 together. From these, it is seen that when a series of steels was heated to 850°C. for 6 hours, no graphitisation occurred: when a series was heated to 950°C. for 6 hours, only a trace of graphite was formed: but when the specimens were heated to 950°C. for 3 hours, slowly cooled to 850°C., and soaked for 3 hours, the high carbon steels were considerably graphitised. If graphite is formed by direct decomposition of cementite, then the second treatment (six hours at 950°C.) should have produced more graphite than either of the others. Actually, the formation of graphite was greatly increased, by introducing a short period of cooling. During cooling, carbon is precipitated from solution in austenite; and apparently it is during precipitation from solution that graphite is most liable to be formed. This fact might be accounted for by supposing that cementite is deposited from solution in austenite, and immediately breaks down into graphite. A similar explanation has been advanced by Portevin to explain the peculiar form of the graphite flakes in cast iron. Reasoning such as this is founded on the assumption that in solution in molten iron or austenite, carbon exists as carbide of iron. If this assumption is accepted, it would seem reasonable to believe that the carbide would come out of solution as such, and break down, under suitable conditions, to the stable form, graphite. Nevertheless, it is not easy to understand why cementite should break down rapidly during precipitation, and yet be comparatively stable once precipitated.

The explanation just given is still less acceptable, if carbon exists in solution in gamma iron as atomic carbon, and not as cementite; for the mechanism of deposition would then involve the formation of cementite and its immediate decomposition. Since the introduction of X-ray analysis, the trend of modern opinion has been towards the conception of carbon as atomic carbon in solution (Rosenhain, Westgren, Jeffrey, Griffiths, Owen, Bain, Hull, and Bragg). The results obtained in the present experiments do not throw any light on this point, but they indicate that, whatever may be the form of carbon in solution, it can separate, under appropriate conditions, as graphite or cementite; and that, once deposited, cementite does not break down directly into graphite.

In the present research, the process of graphitisation has been investigated by subjecting high carbon steels to heat-treatment. Exactly the same end may be attained with white cast-iron, which is essentially a steel of very high carbon content. There are two methods of producing malleable cast-iron from white iron, the Reaumur and Blackheart. The aim of the former process is to eliminate carbon, and that of the latter to convert brittle cementite into the finely divided form of graphite known as temper carbon. It will be seen that the Blackheart process is closely related to the present investigation.

Malleableising, by the Blackheart process, is practically confined to America, and a great deal of research has been performed in that country in order to establish the best conditions for graphitisation. As a result of this work, American investigators have come to the conclusion that cementite cannot break down directly into graphite.

Thus, H.A.Schwartz (12) showed that when white iron is heated above A_1 and soaked at a constant temperature, graphitisation, once initiated, proceeds slowly at a constant rate. He does not think this consistent with the view that cementite decomposes directly into graphite.

In collaboration with Payne, Gorton and Austin, (13) the same author examined a large number of white irons that had been heated till graphitisation just commenced. It was found that the graphite never commenced to form in the centre of a grain of cementite, very seldom in contact with cementite, and usually within a patch of austenite, remote from any particles of cementite. As a result of their work, these authors advanced a theory that the formation of graphite on reheating a white iron was due to the existence of two distinct solid solutions of carbon in gamma iron - austenite, in which the carbon is present as carbide; and boydenite, in which carbon is less soluble, and is present as atomic carbon. According to that view, the mechanism of graphitisation is as follows:-

> On soaking above A_1, cementite dissolves in austenite till the solid solution is saturated. Some austenite then changes to boydenite, and as this constituent cannot retain so much carbon in solution as austenite can, it rejects the surplus carbon, which is deposited as graphite. The solid solution then reverts to austenite, which dissolves more cementite to maintain its concentration. The former process is then repeated; and the solution of cementite in austenite alternates with the formation of boydenite and deposition of graphite, till all the free cementite has been transformed. The

alloy then consists of particles of temper carbon embedded in saturated austenite. Up to this point, the reaction is comparatively rapid and proceeds at a constant rate. The austenite then changes to supersaturated boydenite, which deposits its excess carbon as graphite. This stage of the reaction proceeds slowly till the equilibrium state for the temperature is reached. The alloy then consists of saturated boydenite and temper-carbon. If the temperature is now allowed to fall, boydenite continues to deposit temper-carbon. To achieve this however the rate of cooling must be very slow - otherwise boydenite reverts to austenite, and cementite is deposited.

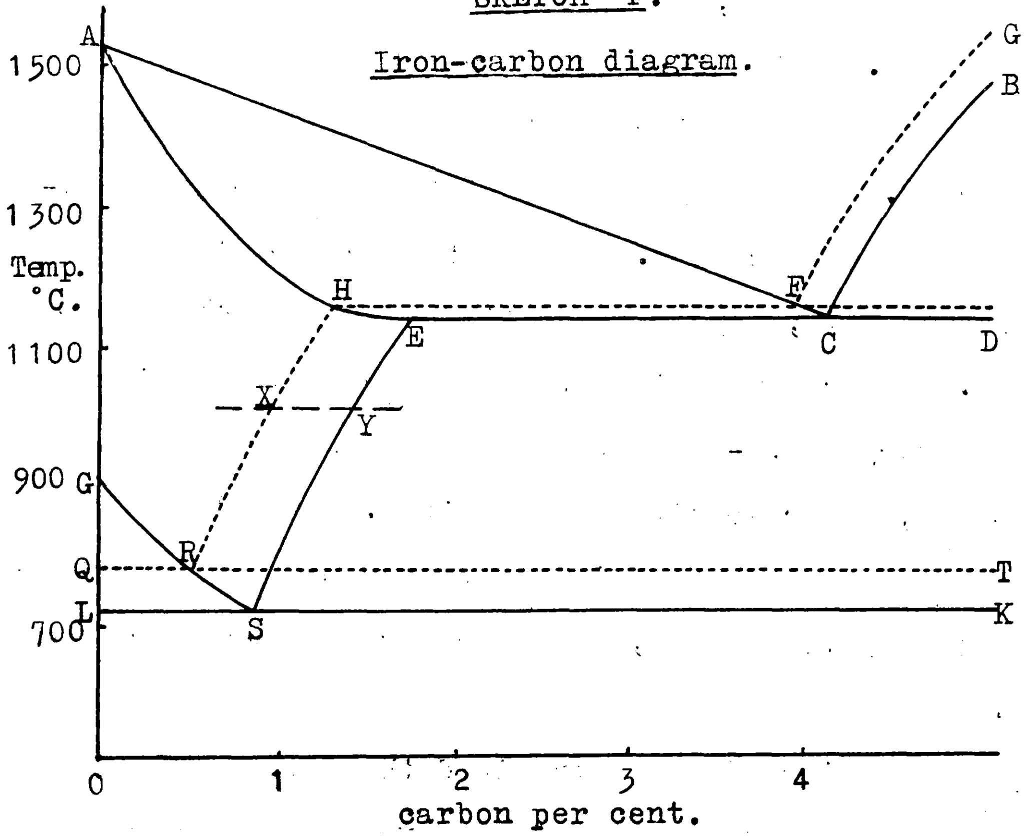
The hypothesis is ingeniously worked out by its authors, but it has many weaknesses, and introduces unnecessary complications.

Hayes and Diederichs (14) examined the data published by Schwartz, and suggested that the observed phenomena could be simply explained by assuming that the solubility of graphite in austenite is less than that of cementite. There are analagous cases in general chemistry where a liquid in contact with two solid modifications, one stable and the other metastable, gives the same solution, or at any rate solutions of the same type, differing only in the amount of solid dissolved.

Phillips and Davenport (15) consider that the formation

SKETCH F.

Iron-carbon diagram.



of graphite on heating a white iron, is due to slight and unavoidable fluctuations of temperature. As the temperature rises, cementite dissolves in austenite; as the temperature falls, the less soluble graphite is deposited. Although this idea is more acceptable than the boydenite theory, it is equally unnecessary. As will be shown later, the fact that graphite is less soluble than cementite, is a sufficient explanation of the phenomena.

The work of the American investigators appears to show that on heating white iron, cementite does not break down directly. The general opinion is that, in some way or other, graphite is deposited from solution. The results of the present investigation are in complete agreement with this conclusion.

The Mechanism of Graphitisation.

In the Sketch opposite this page is shown the iron-carbon diagram as commonly accepted. The full lines represent the metastable system, and the dotted lines the stable system. The Sketch differs from the usual diagram, however, in that the curve representing the separation of graphite from gamma iron, has been produced to meet GS at R, and this is considered to be a eutectoid of ferrite and graphite. The explanation of/

of the graphitisation phenomena lies in this diagram.

If a high carbon steel (or a white cast-iron) is heated to, say, 1000°C., and maintained at that temperature, cementite will dissolve in austenite till the concentration of dissolved carbon reaches the point on SE corresponding with that temperature. In the instance given, cementite will dissolve till the austenite contains 1.5 per cent. of carbon, as represented by point Y (Sketch F.). In the stable condition, however, austenite can only hold 1.0 per cent. of carbon in solution (point X). The excess carbon will therefore be deposited, not as cementite, which is soluble up to 1.5 per cent., but as graphite which is only soluble to 1.0 per cent. As the deposition of graphite reduces the carbon content of the austenite to less than 1.5 per cent., cementite will be enabled to dissolve, to maintain that concentration. These two reactions, solution of cementite and deposition of graphite, will proceed simultaneously till all the free cementite has been transformed to graphite. Thereafter deposition of graphite will continue till the carbon content of the austenite has been reduced to 1.0 per cent. The alloy will then consist of austenite of this composition, together with graphite, and further soaking will have no effect.

If the temperature now falls very slowly, graphite will be deposited along HR till, at about 775°C., the remaining/

remaining austenite contains approximately 0.56 per cent. of carbon. At this point, the austenite will break down into the eutectoid mixture of ferrite and graphite, and the alloy will consist of grains of graphite embedded in the eutectoid. The deposition of graphite on cooling will proceed only if the rate of cooling is excessively slow. The critical rate of cooling depends largely on the amount of silicon present. No data is available for pure iron-carbon alloys, but judging from the figures given by German and American workers for alloys containing silicon, the rate for those with no silicon must be about 1°C. or 2°C. <u>per hour</u>. If this rate is exceeded, graphite will not be deposited completely, but cementite will separate at a lower temperature, as given by the ordinary metastable diagram.

American writers appear to be greatly concerned because a eutectoid structure of ferrite and graphite has never been observed microscopically. Phillips and Davenport attempt to explain this by suggesting that the eutectoid is at 0.0 per cent. of carbon. Schwartz, Payne, Gorton, and Austin propose that just above the eutectoid temperature, apparently about 770°C., HR bends off abruptly to near 0.0 per cent. of carbon, so that the eutectoid consists essentially of ferrite. It is evident, however, that if there is no ferrite-graphite eutectoid/

eutectoid, then A_1 stable must be at or above 900°C.; and if there is a eutectoid, it must occur at the point where HR cuts GS (Sketch F.) - that is, a ferrite-graphite eutectoid containing approximately 0.0 per cent. of carbon could only occur within a few degrees of 900°C. No suggestion has ever been made that A_1 stable is at or near 900°C. All the evidence indicates 775°C. as the temperature of the eutectoid. It would be impossible for HR to bend at 775°C. to 0.0 per cent. of carbon, for at that temperature it must cut GS at approximately 0.56 per cent. of carbon, and this must be the composition of the eutectoid.

The absence of a typical eutectoid structure is not surprising, for it must be remembered that only slow cooling will cause the alloy to behave according to the stable system. Even with 1.0 per cent. of silicon, the rate must not exceed 4°C. per hour. With this rate of cooling, the constituents of the eutectoid have every opportunity of segregating into a coarse mixture such as is actually found.

It may be advisable to give here some of the evidence indicating that the temperature of the stable eutectoid transformation is about 775°C. Hayes and Diederichs found that when completely malleableised iron (ferrite + graphite) is heated for the first time, carbon enters into solution to approximately 0.6 per cent., the solubility increasing gradually with rising temperature. This indicates that the eutectoid contains about 0.6 per cent. of carbon. These authors also point out that work done by Schwartz shows that the solubility of graphite in/

in iron increases from less than 0.1 per cent., to more than 0.5 per cent. as the temperature rises from 750°C. to 775°C., which places the temperature of the eutectoid between these limits. Schwartz found by quenching alloys of ferrite and graphite from gradually increasing temperatures, that there was a marked increase of electrical resistance between 760°C. and 780°C. - again placing the temperature of the eutectoid in the neighbourhood of 775°C. Hayes and Diederichs find confirmation of this temperature in the iron-carbon diagram founded on the work of Carpenter and Keeling.(16) They point out that no satisfactory explanation has been given of the many points on the diagram between 775°C. and 800°C. Up to 0.4 per cent. of carbon, these points lie at a lower temperature than the range mentioned, and undoubtedly refer to the magnetic transformation; above 0.4 per cent. of carbon, however, the temperature of the change rises abruptly by about 25°C. and remains thereafter in the neighbourhood of 800°C. It seems possible that these points may be indications of the stable eutectoid. Hayes, Flanders, and Moor(17) took thermal curves of iron-carbon alloys containing 0.95 per cent. of silicon, with slow rates of heating and cooling, and found A_1 stable to be at 771°C. and A_1 metastable at 759°C. With the available data, it is not possible to determine the exact temperature of the eutectoid, for pure alloys, but 775°C. appears to be a satisfactory approximation.

It/

It is of interest to note how the above hypothesis agrees with the results obtained in the process of malleableising. It is obviously impossible to heat a white iron within the area between the curves RH and SE in order to initiate graphitisation; but the presence of 1.0 per cent. of silicon in commercial irons greatly accelerates the commencement of the reaction.

At the soaking-temperature, as we have seen, temper carbon is formed by a process of solution and deposition. As was indicated by the present experiments, and is more clearly shown by Schwartz, raising the soaking-temperature increases the rate at which graphitisation proceeds; but, as will be seen from Sketch F., the reaction proceeds further at lower temperatures; for the amount of carbon retained in solution in the austenite is then smaller. To achieve complete graphitisation, this dissolved carbon must also be deposited as graphite; and this can only occur during cooling. Consequently, cooling must be slow.

The commercial white irons used in the Blackheart process contain about 1.0 per cent. of silicon. It has been found that with such an alloy, cooling at the rate of about 10°C. per hour will ensure deposition of graphite along HR, but the rate must be reduced to 4°C. per hour before 775°C., so that the eutectoid of ferrite and graphite may be formed. The temperature of the stable eutectoid/

eutectoid is not accurately known. It depends to some extent on the composition of the iron. For this reason, the slow rate of 4°C. per hour must be maintained during a considerable interval.

For commercial white irons (low sulphur), the following scheme of cooling should give complete graphitisation:-

10°C. per hour to 780°C.: 3-4°C. per hour from 780°C. to 720°C.: thereafter as convenient.

In practice, this scheme would often prove troublesome, because the temperature of the boxes in the centre of the furnace may be as much as 200°C. higher than that of the boxes near the outside. If, then, the temperature were allowed to fall at 10°C. per hour till the centre boxes had reached 780°C., those at the outside would have passed through the eutectoid point at that rate, and the castings in these boxes would contain pearlite. For this reason, it is generally advisable to cool at a steady slow rate from the soaking-temperature to below 700°C. Actually, Schwartz recommends a cooling rate of 5.5°C. per hour. This does not give complete graphitisation, as the rate is too high at the eutectoid temperature; still, the final iron generally contains less than 0.15 per cent. of combined carbon.

The selection of the most suitable soaking-temperature for graphitisation is a matter of some difficulty. The higher/

higher the soaking-temperature, the more rapidly is the free cementite converted into graphite; but the castings must be cooled very slowly from the soaking-temperature, and this means that much of the time saved by using a high temperature is lost during the more prolonged cooling. Moreover, high temperature soaking produces coarse temper carbon, and this is considered to be detrimental to the material. Actually the effect of the coarseness of the carbon is not so great as is believed. Thus, Phillips and Davenport found little difference in the mechanical properties of cast-irons malleableised at 1100°C. and at 830°C. respectively. The specimens that had been treated at the higher temperature actually gave a better elongation, though their tensile strength was slightly inferior.

It is universally stated in text books that in the Blackheart process, the soaking temperature should be between 750°C. and 850°C. These temperatures are too low. In practice, the temperature usually lies between 850°C. and 1000°C.

The Mechanism of Solidification.

Although no experiments on solidification were performed in the present research, it appears to be reasonable to believe that as graphite can be deposited directly from austenite, it may also be deposited directly/

directly from the liquid.

The usual view on the solidification of cast iron, is that cementite is deposited from the liquid, and may break down immediately or subsequently to graphite. Thus, just after solidification the alloy consists of the eutectic mixture of austenite and cementite, together with varying amounts of primary austenite or cementite. The graphite in a grey cast iron, is produced by decomposition "in situ" of the cementite. The occurrence of the graphite in the characteristic curved flakes, which bear no relation to the form of the cementite from which they are supposed to arise, is a weak point in this theory. To explain this point, Howe suggests that cementite, deposited from the liquid at 1130°C., redissolves momentarily in the melt and breaks down into graphite flakes. Portevin states that the curved flakes of graphite are formed by decomposition of cementite during solidification, and that decomposition in the solid produces temper carbon. Despite the peculiar form of the graphite, however, many investigators are convinced that these flakes are formed in the solid. Portevin's explanation is credible, but the other suggestions are not easily believed. It has never been shown that cementite, in a solid cast-iron, can form flakes of graphite. Phillips and Davenport heated/

heated a white iron slightly below its melting-point for periods up to 60 hours, and obtained temper-carbon, but no signs of graphite flakes.

It is a reasonable assumption that the mechanism of the formation of graphite on solidification does not differ from the graphitisation of a white iron during slow cooling from the soaking-temperature. That is to say, if the cooling is slow enough, graphite is deposited from solution; if the cooling is more rapid, cementite is formed. This is in accordance with the generally accepted iron-carbon diagram, which shows the stable eutectic between austenite and graphite, and the metastable eutectic between austenite and cementite. It may be contended that the graphite-austenite eutectic has never been observed microscopically. This apparent anomaly may however be due to the necessary slowness of cooling, and to the low specific gravity of graphite. During the formation of the eutectic, the graphite constituent is tending to rise to the surface, the typical eutectic structure being thus deranged. To ensure the separation of the graphite-austenite eutectic, very slow cooling is necessary; but even though the cooling be sufficiently rapid to undercool the melt past this eutectic point, and the separation of the cementite-austenite eutectic commence, graphite may still separate at the lower temperature if the period of solidification be prolonged,

prolonged, for at this point graphite is labile.

As it is not possible to observe microscopically the actual solidification, examination of quenched material has been substituted. By this method, some easily misinterpreted data have been obtained. Thus, Wust allowed a sample of pig-iron to freeze, and quenched it as soon as it appeared to be solid. He then quenched a similar sample when it was only two-thirds solid. He observed that the latter contained less graphite and more cementite than the former. Similar experiments have been performed by other workers. Honda (18) states that: "The question whether the liquid iron-carbon alloys contain carbon as graphite or as cementite, has an intimate connection with the two theories of graphitisation above referred to, and can be solved by a quenching experiment with the melt, in which the molecular state of the dissolved carbon in the liquid can be kept at room temperature." After showing that by increasing the rapidity of quenching from the melt, the amount of graphitisation is decreased, he continues: "These results show that in the liquid state, carbon exists in the combined state, otherwise quick cooling would cause a greater degree of graphitisation. If this conclusion is correct, it is very natural to conceive that the first product of solidification is cementite but not/

not graphite."

Few metallurgists believe that the successive states obtained by progressively increasing the rate of cooling are necessarily intermediate between the state that existed at the quenching-temperature, and that to which it changes on slow cooling. Thus, troostite and martensite, which are obtained by quenching austenite, are not intermediate stages in the formation of pearlite from austenite during slow cooling. Similarly, the fact that cementite is obtained by quenching a molten cast-iron, does not prove that cementite is an intermediate stage in the formation of graphite.

Some investigators, who have quenched iron-carbon alloys immediately after solidification and at lower temperatures, have been led by their results to suppose that cementite separates from the melt, and that graphite is formed subsequently in the solid state. Among these are K. Taware and G. Asahara (19) who cast pure iron-carbon alloys into moulds that had been heated to different temperatures. The castings were soaked for varying periods at the temperature of the mould, and cooled either quickly or slowly. They describe the castings as being partly graphitised, completely graphitised, or graphitised in certain parts. In some cases, they give analyses, but the figures naturally give no indication of the form of the graphite. Their results are/

are very inconclusive. Thus, when they poured a cast-iron into a mould maintained at 1100°C., they had no means of finding out how long the alloy took to solidify. When the solidifying alloy was quenched after having been thirty minutes at 1100°C., it was found to contain cementite in certain parts; whereas when it was allowed to remain ninety minutes at that temperature before quenching, it was completely graphitised. The appearance of graphite in the first case was probably due to the fact that solidification was incomplete; consequently, certain parts of the alloy were quenched from the liquid state. In support of this view, it may be mentioned that when the castings were allowed to stand for ninety minutes at 1125°C., they were still liquid. The results obtained by these investigators appear to show that when the alloys were cast into a mould maintained at 1100°C., solidification was slow, and produced complete graphitisation. When the mould was pre-heated only to 1000°C., solidification was more rapid, and graphitisation was incomplete. When the temperature of the mould was 900°C., solidification was sufficiently rapid to prevent formation of graphite flakes.

In the opinion of the writer, the data published on the solidification of iron-carbon alloys are in agreement with the results obtained during the present investigation into the graphitisation of high-carbon steels. They/

They indicate that if cooling is sufficiently slow, graphite is deposited directly from solution - either in molten iron or in gamma iron. There is no definite evidence in favour of the theory that in solid iron-carbon alloys, cementite ever decomposes directly into graphite. Although this may not apply to alloys containing a high percentage of silicon, American research on malleable cast-iron appears to show that 1.0 per cent. of silicon does not materially affect the mechanism of graphitisation.

SECTION 3

ELECTRICAL RESISTANCE

The electrical resistance of alloys has been investigated by physicists and metallurgists for many years, and the importance of the subject is universally admitted. The foundations of this branch of physical metallurgy were laid by Matthiessen,(24) from whose experimental results, as was shown by Le Chatellier,(25) it was evident that alloys could be classified in groups.

In the first of these groups are all alloys whose components are insoluble in the solid state. The electrical conductivity is, here, a linear function of the volume-concentration of the two components. As was pointed out by Guertler,(26) this is not always the case. For example, if one of the components be a poor conductor and be concentrated at the grain boundaries, the resistance must be greater than if it were dispersed evenly through the alloy.

In the second group are these alloys whose components form a continuous series of solid solutions. The conductivity curve, in this case, is U-shaped.

The effect of foreign elements on the resistance of iron has been studied by numerous investigators. Barrett Brown and Hadfield(27) investigated a large number of alloys, and concluded that the effect on the resistance of iron caused by adding to it 1 per cent. of a foreign element was inversely proportional to the atomic weight of the element, and that if several foreign elements were present, each had its own effect independent of the others.

Benedicks(28) formulated the rule that equivalent quantities of foreign elements dissolved in iron caused the same increase of resistance, and gave an equation by means of which the resistance of a quenched steel could be calculated from its chemical composition. It will be evident that the effect of foreign elements in steel will depend on whether they are in solution or not. If the impurities are dissolved in the iron, as in quenched steel, the conductivity-concentration curve will be of the form U. It is generally considered that in quenched steels the foreign elements exist in solution - enforced or otherwise - and empirical formulae, such as Benedicks', for the calculation of the resistance, can only hold for small amounts of the added elements.

In annealed carbon-steels, according to the accepted view, the carbon exists as free cementite. The resistance-concentration curve should, therefore, resemble that of alloys of Group I above, the concentrations concerned being those of iron and cementite. An examination of the resistance of annealed steels should therefore show whether carbon is insoluble in alpha iron.

Method of Measurement Employed.

The resistivity determinations were carried out by passing a known current through a long narrow cylindrical specimen, and measuring the voltage-drop between two knife-edges, which were a known distance apart and in contact with the specimen. The specific resistance, at the temperature of the test, was obtained by calculation from the diameter of the specimen and the distance between the knife-edges.

As the specific resistance is inversely proportional to the square of the radius, it is necessary that the radius should be accurately known; and if the specimen is only a few millimetres in diameter, it is difficult to measure it with accuracy. In spite of this fact the specimens used in the present investigation were thin - the actual dimensions being 9 cms. long by 3 mms. diameter. It was believed that in practice the advantages of thin specimens outweighed their disadvantages. For example, if the diameter is large, a satisfactory potential-drop at the knife-edges can only be secured by passing a current of several amperes through the specimen and a large battery of accumulators is necessary. This difficulty could be overcome, by using long specimens, but in practice it is essential that the specimens should be comparatively short, as otherwise they bend badly on quenching. An even more important point is that thick specimens crack on quenching. Edwards(29) used bars one centimetre in diameter when investigating the resistance of chromium steels. The majority of the specimens cracked, and he was forced to take the minimum diameter of the bar, in the plane of the crack, as the diameter of the specimen. None of the specimens used in the present investigation cracked on quenching.

A method will be described later, by which the mean cross-sectional area of the bars was measured with satisfactory accuracy.

Measurement of the Resistance of a Standard Specimen.

The first step in the investigation was the measurement of the resistance, at 17°C., of an annealed specimen, which was to be used as a standard during the experiments. For this purpose, the specimen was held down firmly by powerful elastic-bands on two knife-edges, which were rigidly fastened 5 cms. apart on a wooden block. A Beckmann thermometer was placed in contact with the specimen, which was enclosed in a box so that its temperature could be accurately measured. A shunt of 0.1 ohm resistance was connected in series with the standard, and a current of 1.5 amperes, taken from a battery of accumulators, was allowed to flow through the circuit. The potential-difference between the knife-edges was measured with a potentiometer, and the value of the current was obtained with a second potentiometer, by reading the fall of potential between the terminals of the shunt. The resistance of the specimen at the temperature of the experiment was obtained by calculation.

Readings of resistance were taken during several days, the temperature at each experiment being read to one hundredth of a degree. By means of a switch, the specimen could be cut out of the circuit, and a piece of wire substituted for it. The current was sent through this alternative circuit except when readings were being taken - thus eliminating any danger of heating the specimen. If the current was shut off completely, by breaking the circuit, between readings, it altered rapidly for some time on switching on again.

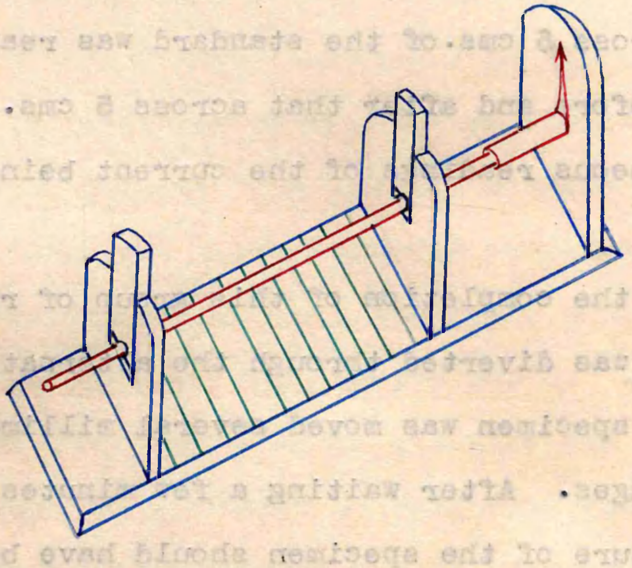
The results of the standardisation are shown in Table 4. It was found that a rise in temperature of 1°C. produced an increase in resistance of 2.3 microhms. The corrected readings in the third column of the table show the resistance of the specimen at 17°C., and the mean of all these values has been taken as the true resistance of the standard specimen at that temperature.

Measurement of the Resistance of the Heat-treated Specimens.

The heat-treated specimen, the resistance of which was required, was held firmly on a second pair of knife-edges, similar to that described above, by means of elastic bands, and was joined in series with the standard as is shown in Figure 15. The current, which had been flowing through the alternative circuit until it was fairly constant, was switched through the specimens. The potential drop across 5 cms. of the standard was read on the potentiometer before and after that across 5 cms. of the unknown, simultaneous readings of the current being taken in each case.

On the completion of this group of readings, the current was diverted through the alternative circuit, and the specimen was moved several millimetres on the knife-edges. After waiting a few minutes in case the temperature of the specimen should have been affected by handling, a second series of readings was taken. Several groups of readings were taken for each specimen, the number depending on the uniformity of the steel.

From each group of readings, the resistance of the heat-treated specimen was calculated, the temperature being corrected to 17°C. by calculation from the resistance of the standard at the temperature of the experiment and its known resistance at 17°C. In making this correction,



it was assumed that the temperature-coefficient of resistance of the heat-treated specimen was the same as that of the standard. Since the temperature never varied from 17°C. by more than 2°C., this assumption did not introduce an appreciable error. The mean of all the values obtained was taken as the resistance of the specimen.

In order to calculate the specific resistance of the steel, it was necessary to find the mean cross-sectional area of each specimen. This was done by measuring the diameter at equal intervals along and around the specimen. A simple device, illustrated in the sketch opposite, was employed.

The specimen was held by elastic bands between two grooved supports on a wooden base. The length of the base between the supports was divided by a scale into twelve equal parts. Readings of the diameter were taken at points that lay directly above these divisions. By means of the pointer that was fitted to one end of the specimen the latter was rotated through 60°, and the diameter again measured at twelve equal parts. A final set of readings was taken after rotating the specimen through 120°. The quarter-square of each reading was found, and the mean of the quarter-squares was taken as the square of the radius of the specimen.

It should be noted that the specific resistance is is inversely proportional to the square of the radius, and this means not only that the diameter should be measured accurately, but that the cross-sectional area should be calculated at each measured diameter, and the average of these values found. The cross-sectional area is, of course, directly proportional to the square of the radius.

Consideration of Results.

Quenched Steels.

From Figure 16, it will be seen that the specific resistance of hypereutectoid carbon-steels is increased by raising the quenching-temperature, as would be expected from the iron-carbon diagram. Campbell (30) and Soldau (31) found that the specific resistance of hypoeutectoid steels was increased by raising the quenching-temperature; on the other hand, data published by Le Chatelier, (32) Benedicks, (33) and McCance, (34) do not show this increase. The results obtained by the author indicate that the specific resistance of hypoeutectoid steels is decreased by raising the quenching-temperature.

The values of specific resistance of carbon steels quenched at 1100°C., plotted against carbon content, lie approximately on a straight line, in spite of the fact that the hypereutectoid steels contain austenite. From this it might be deduced that the specific resistance of austenite at 17°C. is the same as that of martensite of the same carbon content. The specific resistance of the steel containing/

containing 1.73 per cent. of carbon quenched at 1000°C. differs only slightly from that of the same steel quenched at 1100°C., whereas more austenite was retained by quenching from 1100°C. than from 1000°C. According to the generally accepted iron-carbon diagram, the solubility of carbon in gamma iron at 1000°C. is not more than 1.5 per cent. In consequence, if austenite has the same specific resistance as martensite, the specific resistance of a steel containing 1.61 per cent. of carbon should be increased approximately 3 microhms by raising the quenching-temperature from 1000°C. to 1100°C. - this value was obtained from the slope of the resistance-concentration line of carbon steels quenched at 1100°C. (Figure 16). Actually the specific resistance of the specimen was increased only 0.44 microhms by this treatment. As more austenite was retained by quenching from the higher temperature, the results indicate that austenite has a lower specific resistance than martensite.

From Figures 17, 18, and 19, in which the resistance-composition curves of the three series of alloy steels are plotted, it will be seen that, in general, raising the quenching-temperature increases the specific resistance of hypereutectoid steels/

steels by causing increased solution of cementite, but decreases the specific resistance of hypoeutectoid steels. The chromium steel containing 1.73 per cent. of carbon, quenched at 1100°C., consisted mainly of austenite, and had a lower specific resistance than the same steel quenched at 1000°C., which was mainly martensitic; this again indicates that martensite has a higher specific resistance than austenite. Similarly, a nickel-steel containing 1.06 per cent. of carbon had a higher specific resistance when quenched at 900°C. than when quenched at higher temperatures.

In order to compare the results obtained by the writer with those of other investigators, the values of specific resistance must be corrected for manganese, silicon and phosphorus. Thus:-

1.0 per cent. of manganese raises the resistance of iron 5.5 microhms......(Barrett, Lang and Matsushita);

1.0 per cent. of silicon raises the resistance of iron 13.5 microhms.....(Burgess, Gumlich & Le Chatelier);

1.0 per cent. of phosphorus raises the resistance of iron 11.0 microhms.....(D'Amico).

The values of the specific resistance of carbon steels quenched at 1100°C., corrected according to these figures, lie on a line that may be expressed by the equation:

$$R = 7.5 + (26.7 \times C),$$

where R = resistance in microhms,
and C = carbon content.

This equation is in excellent agreement with Benedicks' formula. According to the equation, the specific resistance of pure iron is 7.5 microhms - this being the value found by extrapolation of the resistance-concentration curve of quenched steels to 0.0 per cent. of carbon; Thompson(35) found the resistance of pure iron to be 7.3 microhms. On the other hand, Guillet,(36) Gumlich,(37) and Yensen,(38) found the resistance of pure iron to be between 9.5 and 9.9 microhms; this is the value that would be obtained by extrapolation of the corrected resistance-concentration curve of annealed steels to 0.0 per cent. of carbon.

In Figure 21, the resistance-concentration curves of the carbon steels and alloy steels quenched at 1000°C., are plotted together. From the curves, it will be seen that the addition of 1.67 per cent. of chromium has approximately the same effect on the resistance of steel as 3.64 per cent. of nickel; this is true also for steels quenched at 900°C. and 1100°C. The curve for nickel-chromium steels lies above the other curves; and it can be shown that the resistance of/

of a nickel-chromium steel of given carbon content is greater than that of a corresponding carbon steel by an amount that equals the sum of the increases of resistance caused by each of the special elements. Thus, allowing for the fact that the nickel-chromium steel does not contain exactly the same percentage of nickel or of chromium as the two ternary steels, the calculated resistance of a nickel-chromium steel containing 0.51 per cent. of carbon, quenched at 1000°C., is:-

$$\left(20 + \frac{1.72}{1.67} \times 9.0 + \frac{3.52}{3.64} \times 8.5\right) \text{ microhms,}$$

where 20 microhms is the resistance of the carbon steel containing 0.51 per cent. of carbon;

9.0 microhms is the value, obtained from the curves, of the increase of resistance caused by the addition of 1.67 per cent. of chromium to the carbon steel containing 0.51 per cent. of carbon;

and 8.5 microhms is the corresponding increase due to 3.64 per cent. of nickel.

The calculated resistance of the nickel-chromium steel is thus 37.5 microhms; the value experimentally found was 37.6 microhms.

The resistance of a nickel-chromium steel containing 1.0 per cent. of carbon, calculated in the above manner, is:-

$$\left(35 + \frac{1.80}{1.67} \times 7.0 + \frac{3.50}{3.64} \times 6.0\right) \text{ microhms,}$$

that/

that is, 49.0 microhms; the value experimentally found was 48.0 microhms.

Similarly, the calculated resistance of a nickel-chromium steel containing 0.51 per cent. of carbon, quenched at 1100°C., is 36.3 microhms, as against 37.6 microhms actually found; at 1.0 per cent. of carbon, the calculated resistance is 49.0 microhms as against 48.6 microhms experimentally found.

According to Benedicks' formula, the increase in the specific resistance of iron due to each percent. of chromium added should be 6.2 michroms; but the results obtained by the writer show that the effect of chromium cannot be so simply expressed. The effect of special elements on the resistance of steel depends on whether the steel is in the quenched state or in the annealed state, on the quenching-temperature, and on the carbon content. From Figure 21, it will be seen that the addition of 1.0 per cent. of chromium increases the resistance of a carbon steel quenched at 1000°C. by about 4.6 microhms when the carbon content is less than 1.0 per cent.; beyond this carbon content, the increase of resistance due to 1.0 per cent. of chromium falls to 3.3 microhms at 1.61 per cent. of carbon. Similarly, in the case of steels quenched at 1100°C., the increase due to 1.0 per cent. of chromium is between 4.4 and 4.6 microhms/

microhms, when the carbon content is less than 1.0 per cent. - beyond this percentage the results are influenced by the retention of much austenite in the chromium steels. The increase of resistance of steels quenched at 900°C. due to 1.0 per cent. of chromium varies from 5.3 microhms for steels of low carbon-content, to 4.3 microhms at 1.0 per cent. of carbon.

In nickel steels of low carbon-content, the average increase of specific resistance due to 1.0 per cent. of nickel is 2.3 microhms; in steels of high carbon-content, the increase is 1.9 microhms. According to Benedicks' formula, 1.0 per cent. of nickel should increase the specific resistance by 5.4 microhms.

Annealed Steels.

From Figure 16, it will be seen that the specific resistance-concentration curve of annealed carbon-steels shows a discontinuity at the eutectoid composition. The resistance of hypoeutectoid steels increases more rapidly with increase of carbon content than does the resistance of hypereutectoid steels. This indicates that in annealed steels, the amount of carbon retained in solution in the ferrite increases with the carbon content up to the eutectoid composition. Gumlich (37) also/

also found that the resistance-concentration curve of annealed steels consisted of two branches with a discontinuity at the eutectoid composition. On the other hand, Matthiessen (41) and Le Chatelier (32) found that the specific resistance of annealed steels bore a linear relation to the carbon content. It is probable that the lack of agreement is due to the use of different methods of "annealing," rather than to inaccurate methods of determining the resistance; for the accuracy of the measurement of the resistance of heterogeneous substances such as annealed steels is determined not only by the accuracy of the method, but also by the size and disposition of the constituents. Thus, if an alloy consisting of a mixture of a good conductor and a poor conductor be made into a long narrow specimen, such as is generally used in resistivity-determinations, the maximum value of the conductivity of the specimen will be that given by the rule for alloys of Group I above; but the minimum value of the conductivity may be much nearer that of the poor conductor if the particles of the latter be so large, or so concentrated in parts of the specimen, that they tend to occupy the whole of the cross-sectional area in these parts. The less uniform the composition of the specimen, the more likelihood is there of obtaining a high value for the resistance; the coarser the constituents/

constituents, the less likelihood is there of obtaining the maximum conductivity.

In Figure 29, the resistance-concentration curves of the annealed carbon-steels are plotted together with those of the alloy steels. It will be seen that the curve for the nickel steels is parallel to that of the carbon steels; the increase of resistance due to 1.0 per cent. of nickel in annealed steels is between 1.9 and 2.3 microhms, just as was found for quenched steels. This shows that in nickel steels, annealed or quenched, the nickel is always present in solution in the iron. The curve for annealed chromium-steels lies above that of the annealed carbon-steels, but the maximum difference of resistance between the two curves is approximately 4 microhms (at 0.35 per cent. of carbon) whereas in quenched steels, the difference of resistance between the chromium steels and the carbon steels is about 7.5 microhms; this shows that all the chromium is not in solution in the annealed steels. The resistance-concentration curve of annealed chromium-steels approaches that of the carbon steels as the carbon content increases; this appears to indicate that the amount of chromium in solution in the ferrite decreases as the carbon content increases.

Tempered Steels.

The resistance-concentration curves of tempered steels are shown in Figures 16, 17, 18, and 19. The specimens were those used in measuring continuously the change of resistance during tempering at 240°C.; each specimen was tempered for more than six hours. As will be shown in the next section, the resistance of the austenitic chromium and nickel-chromium steels was still falling slowly when tempering was stopped; so these steels were not fully tempered. All the other specimens were in the "fully tempered" condition.

From Figure 16, it will be seen that the resistance of quenched carbon-steels is reduced to within approximately 3 microhms of the resistance of the annealed carbon-steels by tempering at 240°C.; on the other hand, as will be seen from Figure 18, the resistance of the quenched chromium-steels is affected to a much smaller extent by tempering at that temperature. The resistance-concentration curve of tempered carbon-steels is roughly parallel to that of the tempered chromium-steels (Figure 18); the average difference between the resistance of the tempered chromium-steels and that of the tempered carbon-steels is the same as the difference between the resistance of the quenched chromium-steels and that of the quenched carbon-steels. This indicates (a)/

(a) that the amount of carbon retained in solution in chromium steels after tempering at 240°C. is much greater than the amount retained in carbon steels after the same treatment: or (b) that when quenched chromium-steels are tempered at 240°C., iron-carbide is precipitated, leaving the chromium in solution in the iron. The second explanation is in agreement with the views of Andrew and Hyman.(39)

SECTION 4

MEASUREMENT
of
RESISTANCE
during
TEMPERING.

In order to investigate the reactions that occur during the tempering of steel, specimens were immersed in oil at 240°C., and the process of tempering was watched by measuring the resistance of the specimen at the closest possible intervals of time, during several hours.

A number of preliminary experiments showed that special apparatus was required. It was essential: (a) that the temperature of the oil should be constant; (b) that the specimen should be heated rapidly to the tempering-temperature; and (c) that readings should be taken immediately tempering commenced, for at the start of the reaction, the tempering of martensite is rapid. To satisfy these conditions, a large bulk of heated oil was necessary, and this demanded efficient stirring.

The apparatus ultimately devised is shown in Figures 22 and 23.

The Oil-Heating and Circulating System.

A copper tank, containing about 4 gallons of heavy cylinder-oil was heated electrically by means of heating-elements. The oil was pumped, by means of a sud pump, from the heating-tank into one end of a small brass box, in which the specimens were suspended. From the other end of the brass box, the oil flowed under gravity back into the heating-tank. Efficient circulation of the heated oil was ensured by fitting a baffle plate at the inlet end of the tempering-box. The tank and the brass box were packed in asbestos wool, and the system of pipes was heavily lagged, so as to minimise heat losses.

The tempering-box was provided with two lids. The first of these was kept in position before an experiment until the temperature was constant. It was provided with a hole for a thermometer. The knife-edges and the current connections were attached to the second lid.

Electrical Measuring Apparatus.

The resistance of the specimen was measured by passing a current of 1.5 amperes through it, and finding the voltage drop across a length of 5 cms. The current was taken from a battery of "Nife" accumulators, and passed through a standard specimen, the heat-treated specimen, an ammeter (for rough measurement), two rheostats (for coarse and fine control), and a standard resistance of 0.1 ohm.

Two Tinsley Vernier potentiometers were used to measure the voltage drop across 5 cms. of the standard, and that across 5 cms. of the heat-treated specimen simultaneously. Accurate measurement of the current was

was obtained by reading the drop in volts between the ends of the standard resistance on one of the potentiometers.

Determination of the Resistance of the Standard Specimen.

The first step in carrying out the investigation was to determine the resistance of a standard annealed specimen at the tempering-temperature, which was fixed at 240°C. This temperature was chosen because it was sufficiently high to permit tempering of austenite as well as martensite, and not so high as to cause rapid deterioration of the oil.

The specimen was suspended between two pairs of knife-edges, as is indicated in the sectioned drawing in Figure 23, the knife-edges being set approximately 5 cms. apart by means of a distance piece. Each end of the specimen was wrapped in lead foil, and gripped between clamps to which the current connections were fixed. The lid was then placed in position on the tempering-box, through which the heated oil was flowing, and the resistance of the standard was obtained by taking a large number of simultaneous readings of the current flowing round the circuit, and of the voltage drop between the knife-edges. The temperature of the oil, which was read by means of a mercury thermometer, was varied from several degrees below 240°C. to several degrees above, and the resistance of the standard at 240°C. was found by plotting all the values obtained against the corresponding temperatures, and reading off the value at the desired temperature.

Measurement of Resistance During Tempering.

The quenched specimen that was to be tempered was fixed between two pairs of knife-edges on the lid of the tempering-box, and was joined in series with the similarly fixed standard. The current was switched on, adjusted to 1.5 amperes, and allowed to flow till quite steady. During the subsequent tempering, no further readings of the current were taken, as this factor was eliminated by taking simultaneous readings of the voltage-drop across the standard and the quenched specimen by means of the two potentiometers.

As in the case of the standard, the length of the quenched specimen between the knife-edges was approximately 5 cms.; however, as the knife-edges were not rigidly fastened to the lid of the box, but were fixed to brass extension arms so that the specimens should be several inches below the surface of the oil, it was not possible to fix them exactly 5 cms. apart - the distance piece was not found to give sufficient accuracy. In order to find the exact length of the specimen between the knife-edges, the specific resistance of the specimen at 17°C. was measured by the method described earlier for the determination of electrical resistivity at 17°C. Then, after fixing the specimen in position on the lid, the resistance, at 17°C., of the length of the specimen between the knife-edges was measured, and the exact distance was calculated thus:-

If r = the specific resistance of the specimen at 17°C.

R = the resistance, at 17°C. of the length of the specimen between the knife-edges,

& a = the cross-sectional area of the specimen,

then the distance between the knife-edges = $\frac{R.a}{r}$

When the quenched specimen had been fixed in position on the lid, and the current had become steady at 1.5 amperes, the lid was placed on the tempering-box so that the specimens (standard and quenched) were immersed in the heated oil, and simultaneous readings of the voltage drop across each of the specimens were taken one half minute after immersion, at intervals of a half minute thereafter up to ten minutes, at intervals of one minute till twenty minutes, and as frequently thereafter as seemed desirable.

The specific resistance of the tempering specimen at 240°C. was calculated at each reading by means of the formula:-

$$R_t = \frac{E_t \times r \times a}{e_t \times l}$$

where R_t = specific resistance of tempering specimen at time t,

E_t = volt. drop across 5 cms. of " " " " "

e_t = " " " " " " standard " " " "

r = resistance of the standard at 240°C.

a = cross-sectional area of the specimen,

and l = length of the specimen.

In making this calculation, it is assumed that the temperature coefficient of the standard is the same as that of the quenched specimen. This assumption is not correct; but, as the temperature of the oil did not, at any time, differ from 240°C. by more than ±2°C., the error introduced is negligible.

Each quenched specimen was tempered for more than 6 hours. In order that this length of time should be available each day for tempering, it was necessary that the temperature of the oil should be constant at 240°C. by 10 a.m. As it was not possible to leave the apparatus running continuously day and night, the oil had to be allowed to cool during the night, and in spite of the efficient lagging, the apparatus required more than 5 hours to heat again to 240°C. and become constant at that temperature. It was therefore necessary that heating should commence before 5 a.m. Moreover, the oil was so viscous at low temperature, that the circulating-system could not be started till the temperature of the oil was over 150°C. It was therefore required that the heating current should be switched on about 5 a.m., and that the motor for driving the sud-pump should be started as soon as the oil was sufficiently fluid - that is, about two hours later. These requirements were met by the following arrangements:-

At 4.30 a.m. an alarm-clock released a spring rat-trap which switched on the heating-current. At the same time, the first alarm pulled the catch of a second clock into the alarm position. This clock was set to ring at 6.30 a.m.; and at that time it switched on the sud-pump motor, thus starting the circulation of the heated oil. By 9.30 a.m., the temperature of the oil flowing through the tempering-box, was steady at 240°C., and all other preparations for the experiment had been made as described above.

Determination of the Starting Point of the Reaction.

From the results of the tempering experiments, it was found that the fall of resistance during the first few minutes was usually large, and it was evident that extrapolation of the time-resistance curves to meet the resistance axis, in order to obtain the starting-point of the reaction, would not be justifiable.

The tempered specimens were therefore re-quenched, and their temperature coefficients of resistance between 0°C. and 80°C. were measured. The specimens were immersed in a water-bath, the temperature of which was regulated by addition of boiling water. Readings of resistance were taken every few degrees. The values were plotted, and the temperature coefficient was calculated from the line through the points. From this data, the specific resistance of the quenched specimens at 240°C. was calculated, and this was taken as the starting-point of the reaction.

Consideration of Results.

The curves obtained by measuring the change of resistance during tempering at 240°C. are shown in Figures 24 to 35.

As will be seen from any of the curves for martensitic steels, the greater part of the tempering of martensite, as measured by the fall of resistance, is completed within a few minutes. Steels that contain an appreciable amount of austenite, such as chromium and nickel-chromium steels of high carbon-content, temper much more slowly.

It is a reasonable deduction from the curves that martensite decomposes rapidly into troostite, whereas austenite tempers slowly and progressively. If at any time during the tempering of an austenitic steel, the austenite suddenly decomposed into martensite, the martensite thus formed would rapidly temper, causing a rapid fall of resistance; no evidence of this was noted in any of the curves taken at 240°C. At the tempering-temperature, austenite in effect slowly breaks down into alpha iron and cementite, giving rise to a slow steady fall of resistance such as is seen in the/

the curve of the chromium steel containing 1.73 per cent. of carbon quenched at 1100°C. (Figure 30).

As martensite tempers much more rapidly than austenite, there can never be an accumulation of martensite during the tempering of austenite at 240°C.; yet it is commonly known that if an austenitic or partly austenitic steel is heated to 240°C. and cooled, martensite is found in the specimen. This has been shown clearly by Andrew and Honeyman (11) who quenched a carbon steel containing 1.61 per cent. of carbon from 1100°C., and found that the specific volume of the quenched specimen was considerably increased by tempering for one hour at 250°C. Enlund (8) found that the specific volume of a quenched carbon steel containing 1.57 per cent. of carbon was increased by tempering for half an hour at 240°C. The increase of specific volume was due to the formation of martensite from austenite. The tempering-resistance curve of the steel containing 1.61 per cent. of carbon (Figure 25) shows that the martensite was not formed at 240°C.; at that temperature, as was stated above, austenite virtually tempers directly to troostite. The martensite found in the specimen after cooling must have been formed during the cooling from the tempering-temperature. This suggestion is confirmed by the following experiment:-

A quenched chromium-steel containing 1.73 per cent. of carbon was tempered at 240°C.; after a slight initial fall/

fall of resistance due to the tempering of the small amount of martensite in the quenched specimen, the resistance fell slowly and steadily, as will be seen from Figure 30. A second quenched specimen of the same composition was tempered for 320 minutes at 240°C., and was then withdrawn from the tempering-bath and cooled to room-temperature. When the specimen was reheated to 240°C., it was found that the resistance had risen slightly owing to the formation during cooling of some martensite, which rapidly tempered in the characteristic manner of martensite: thereafter the remaining austenite tempered slowly as before. The curves obtained by interrupting the tempering in this manner are shown in Figure 34. A similar experiment was performed with a nickel-chromium steel containing 1.28 per cent. of carbon, quenched at 1100°C.; in this case, the specimen was twice removed from the tempering-bath and cooled to room-temperature. On each occasion, when the specimen was reheated to 240°C., it was found that the resistance had risen slightly due to the formation of martensite; and the martensite thus formed tempered rapidly in its characteristic manner. This is shown clearly in Figure 35.

As was stated above, when the partly-tempered austenitic specimen was replaced in the tempering-bath after cooling to room-temperature, its specific resistance was/

was slightly higher than it had been before the specimen was cooled. This shows that the martensite formed during cooling had a higher specific resistance than the austenite from which it was formed - in other words, martensite has a higher specific resistance than austenite, at 240°C. On page 82 it was shown that at room-temperature also, martensite appears to have a higher specific-resistance than austenite. On the other hand, McCance found, by cooling austenitic specimens in liquid-air, that the resistance of austenite was greater than that of martensite. In order to confirm this point, a number of partly-austenitic steels were immersed in liquid-air for one hour, the specific resistance of each specimen at 17°C. being measured before and after immersion. The results obtained by this treatment are shown in the following Table:-

TABLE A

Specimen (quenched 1100°C.)	Resistance in Microhms Before immersion in liquid-air.	After immersion.	2 months later.
Chromium Steel B 1·73% of carbon.	41·91	41·53	41·04
Nickel-Chromium G1 1·28% carbon.	37·43	36·56	36·27
Nickel-Chromium G2 1·50% carbon.	46·17	45·18	42·80

From Table A, it will be seen that after immersion in liquid-air, the specific resistance of the specimens at 17°C. was less than it had been before immersion. In order to find whether the same relation would hold at 240°C., the temperature-coefficient of resistance of each specimen was measured after the liquid-air treatment; the results are shown in Table B.

TABLE B

Specimen (Quenched 1100°C.)	Resistance in Microhms Before Immersion in Liquid-Air	After Immersion
Chromium B	0·001664	0·001216
Nickel-Chromium G1	0·001588	0·001319
Nickel-Chromium G2	0·001341	0·001225

Although the readings in the second (middle) column of Table B do not refer to the actual specimens that were used in the liquid-air treatment, they refer to specimens of the same composition quenched from the same temperature (1100°C.); the readings in both columns are therefore strictly comparable. It will be seen that the temperature-coefficient of resistance was reduced by the liquid-air treatment; from this it is clear that the specific resistance of the specimens, both at 17°C. and at 240°C., was less after immersion in liquid-air than/

than it had been before immersion. Microscopic examination showed that some austenite had been converted into martensite by immersion in liquid-air; and the formation of austenite was accompanied by a fall of resistance. It follows that (a) martensite has a lower specific resistance than austenite: or (b) that immersion in liquid-air caused separation of carbon from solution either in austenite or in martensite.

(a) The first explanation is contrary to the other evidence, discussed on Page 104. It is possible however that the martensite formed at a very low temperature differs constitutionally from that formed above room-temperature. It was thought that if this were so, the martensite formed by the liquid-air treatment might change to the normal modification at room-temperature, thus causing a rise of resistance. The specific resistance of the specimens was therefore re-measured two months later, and the results shown in Column (4) of Table A were obtained. It will be seen that instead of a rise of resistance there was actually a slight decrease.

(b) Electrode-potential measurements showed that immersion in liquid-air did not cause separation of carbon from martensite (Page 37); so if the fall of resistance caused by cooling partly-austenitic steels in liquid-air is due to separation of carbon from solution/

solution, the separation must be from solution in austenite.

Although the results obtained in these experiments do not give definite information on the nature of the reaction that occurs during liquid-air treatment, they indicate that the reaction does not consist merely of decomposition of austenite into martensite such as is formed during quenching or during the cooling of a partly-tempered austenitic steel from 240°C. to room-temperature.

The tempering-resistance curves shown in Figures 24 to 35 extend only to 260 minutes; actually, as will be seen from Tables 7A to 7L, each steel was tempered for more than 300 minutes. Whenever possible, each curve was continued till there was no further appreciable change of resistance; so the final values of resistance obtained approximate closely to the stable condition of the steel at 240°C. The resistance of the alloy steels of high carbon content continued to fall even after tempering for 10 hours at 240°C.; as the apparatus could not be left running during the night, the tempering-curves of these steels are incomplete.

In Figures 16, 17, 18, and 19, the resistance-concentration curves of the fully tempered steels are plotted/

plotted along with the curves of the annealed steels. From these curves, it is apparent that the quenched steels do not attain the annealed state on tempering at 240°C. Mr J.M.Robertson has shown, by means of electrode-potential measurements and specific-volume determinations, that after tempering for 3 hours at any temperature up to 550°C., a certain amount of carbon remains in solution; and that the amount of dissolved carbon is not reduced by increasing the duration of tempering, but is reduced by raising the tempering-temperature. Thus, when quenched steels are tempered at any temperature below 550°C., they attain a state that is stable for that temperature, and approaches closer to the annealed state as the tempering-temperature is raised.

W. Fraenkel and E. Heymann (40) measured the resistance of steel specimens at intervals during tempering, the readings of resistance being taken at the tempering-temperature. The authors found that the specimens required one half minute to reach the temperature of the oil in which they were heated; but they quote only one instance in which they succeeded in measuring the resistance of the tempering specimen within four minutes of the commencement of tempering. As the initial velocity of tempering was too great to permit of extrapolation, they/

they devised the following method of filling in the first part of the tempering-curve:-

The quenched specimens were tempered intermittently, a few moments at a time, the resistance being measured at room-temperature before and after each brief period of tempering. The readings of tempering-time and resistance were plotted, and from the shape of the curve thus obtained, the first part of the tempering-curve was deduced. In this method of determining the first points on the tempering-curves, it is assumed that no change occurs during cooling from the tempering-temperature, and that the temperature-coefficient of resistance is not altered by tempering - these assumptions detract seriously from the value of the work. The authors deduced from their results that the tempering of martensite consists of a reaction between iron and carbon to form a carbide FeC_{8-10}. There are not sufficient data in the paper to substantiate this hypothesis; moreover the hypothesis is based on a questionable conception of the process of tempering. In the opinion of the writer, the tempering of martensite consists essentially of precipitation from a viscous solvent. If this be so, then it should be possible to find a simple mathematical expression to represent all the tempering-curves of martensitic steels.

Mathematical Interpretation of the Tempering-Curves.

The mathematical analysis of the tempering-curves by/

by the method described in this section has been confined to martensitic steels, for the only steel in the present series that consisted entirely of austenite was the nickel-chromium steel containing 1.50 per cent. of carbon, and this steel did not temper at 240°C. The tempering-curves of the other quenched steels that contained austenite represent the sum of two reactions - the tempering of austenite and of martensite; these curves were analysed by a method that will be described later.

Martensitic Steels.

By trial and error, the experimental curves were found to conform to an equation of the type:-

$$R_0 - R_T = \frac{X}{1 + \frac{b}{T}} \quad \ldots\ldots\ldots \quad (1),$$

where R_0 = the resistance of the quenched specimen,
= the resistance of the specimen at zero time on the tempering-curve:

R_T = the resistance of the specimen at time T:

and X and b are constants.

When T is very large, that is when tempering is almost completed, equation (1) becomes:

$$R_0 - R_T = X \quad \ldots\ldots\ldots\ldots \quad (2).$$

Now, the specific resistance of the quenched

carbon/

carbon-steels is a linear function of the carbon content; so the resistance of any quenched carbon-steel is given by an equation of the form:

$$R_q = Mc + N \quad . \; . \; . \; . \; . \; . \; . \; . \; . \; . \; (3),$$

where R_q = the resistance of the quenched steels,

c = the carbon content,

and M and N are constants.

It is probable that the specific resistance of the steels fully tempered at 240°C. should likewise be a linear function of the carbon content; however, as was stated above, the specimens were not all tempered for the same length of time, and several of the curves were incomplete. When the values of the specific resistance of the fully tempered steels were plotted, it was not found to be possible to draw a fair straight line through the points; but the points were found to lie approximately on a curve that was represented by the equation:

$$\bar{R} = Ac^2 + Bc + K \quad . \; . \; . \; . \; . \; . \; . \; . \; . \; (4),$$

where $\bar{R}$ = the resistance of the quenched specimen,

c = the carbon content,

and A, B, and K, are constants.

The two curves, one representing the specific resistance of the quenched steels, and the other the resistance/

resistance of the fully-tempered steels, meet at a point the co-ordinates of which are: 0.16 per cent. of carbon, and 14.0 microhms, indicating that the resistance of a steel containing 0.16 per cent. of carbon would not be altered by tempering at 240°C. As there were no specimens in the series containing less than 0.16 per cent. of carbon, it cannot be stated whether the curves would cross at 0.16 per cent. of carbon, or would coincide between 0.16 and 0.0 per cent. of carbon; but it is a reasonable and convenient deduction that 0.16 per cent. of carbon remains in solution after tempering fully at 240°C. Making this assumption, the origin of the resistance-concentration curves may be shifted from 0.0 per cent. to 0.16 per cent. of carbon, thus making the constants K and N (the intercepts on the resistance axis) equal.

As was shown above, when T is very large, equation (1) reduces to:-

$$R_0 - R_T = X,$$

But when T becomes very large, R_T approaches $\bar{R}$; therefore:

$$R_q - R_T = R_0 - R_T = X.$$

$$\text{but} \quad R_q = Mc + N \quad \ldots\ldots\ldots \quad (3),$$

$$\text{and} \quad \bar{R} = Ac^2 + Bc + K. \quad \ldots\ldots \quad (4),$$

$$\text{therefore} \quad X = R_q - \bar{R} = -Ac^2 + (M - B)c \ldots (5),$$

the constants N and K being equal

For carbon steels, the values of the constants are:- $M = 27{\cdot}5$; $B = 9{\cdot}56$; $A = -\ 2{\cdot}23$; $N = K = 14$ microhms; and c = the carbon content minus 0·16 per cent. Substituting these values in equation (5),

$$X = 2{\cdot}23c^2 + 17{\cdot}94c.$$

Equation (1) now becomes:-

$$R_O - R_T = \frac{2{\cdot}23c^2 + 17{\cdot}94c}{1 + \frac{b}{T}} \quad \ldots\ldots\ldots \quad (6)$$

The constant b is the only unknown factor in the equation. By applying the equation to the experimental tempering-curves of the carbon steels, the value of the constant b was found to be 1·5 for all carbon steels. By means of equations (6) and (3), the tempering-resistance curve of any martensitic carbon-steel can be plotted when the carbon content only is known.

The equations for nickel steels are similar; thus:-

$$R_q = 29{\cdot}5c + 23{\cdot}5$$

$$\bar{R} = -\ 5{\cdot}08c^2 + 12{\cdot}1c + 23{\cdot}5$$

$$X = R_q - \bar{R} = 5{\cdot}08c^2 + 17{\cdot}4c.$$

The curve representing the resistance of quenched nickel-steels and that representing the resistance of the same steels fully tempered at 240°C., meet at a point, the co-ordinates of which are: 0·2 per cent. of carbon/

carbon, and 23.5 microhms; so in this case, c is the percentage of carbon in the steel minus 0.2 per cent. For nickel steels, the constant b has the same value as for carbon steels, 1.5. The general equation for nickel steels is therefore:-

$$R_0 - R_T = \frac{5.08c^2 + 17.4c}{1 + \frac{1.5}{T}}$$

Again, by means of this equation, the tempering-resistance curve of any of the nickel steels can be drawn, when the carbon content is known.

Tempering-curves of the carbon steels and nickel steels of the compositions used in the present research were calculated by means of the foregoing equations; they were found to agree excellently with the experimental tempering-curves of the martensitic specimens, as will be seen from Figures 25, 26, 26A, 27, and 28, in which the experimental tempering-curves are shown in full line, and the calculated curves dotted.

The calculated curve for the carbon steel containing 1.61 per cent. of carbon quenched at 1100°C. (Figure 26A) does not agree with the experimental curve because of the presence of a considerable amount of austenite in the quenched steel; as was stated above, the equations are only applicable to martensitic steels. With the available data, it is not possible to deduce an/

an equation for austenitic steels by the same method as was used for martensitic steels; but the divergence of the calculated curve of the austenitic carbon steel from the experimental curve shows that austenite tempers much more slowly than martensite.

The nickel steels containing less than 0.67 per cent. of carbon are in agreement with the calculated curves, but the steels containing 0.87 and 1.06 per cent. of carbon evidently contain austenite.

In the present investigation, only one temperature (240°C.) was used. The rate of tempering must obviously depend on the tempering-temperature. Fraenkel and Heymann investigated the process of tempering at a number of temperatures by means of resistance measurements. They assumed that at any temperature a quenched steel finally attains the annealed condition, and expressed each reading of resistance as a percentage of the difference between the resistance of the steel as quenched and as annealed. It is not possible from their data to deduce accurately the real limiting values of resistance to which the steels would tend at the different tempering-temperatures; still, by means of their results a general idea of the effect of temperature on the rate of tempering may be obtained. Curves showing the tempering-rates of a/

a steel containing 1·19 per cent. of carbon at various temperatures, are shown in Figures 36 and 37. These curves were calculated from the results of Fraenkel and Heymann. The general equation for the tempering of martensitic steels was applied to the curves in Figure 36; but, probably owing to the indefinite nature of the data, the results were unsatisfactory. However it was clear that the factor b becomes large as the tempering-temperature is lowered. At low temperatures, according to Fraenkel and Heymann's results, the factor b is not constant, but varies with time.

Although the tempering-resistance curve of any martensitic steel can be drawn with fair accuracy by means of the equation derived above, the equation in itself merely expresses the fact that at the commencement of tempering, the reaction is rapid, and that the rate diminishes as the concentration of carbon in solution decreases. The equation indicates that there is no reason to believe that the process of tempering consists of a reaction between iron and carbon to form a carbide FeC_{10}, as Fraenkel and Heymann were led to believe; but it does not throw any light on the exact nature of the process.

In the opinion of the writer, the tempering of steel/

steel comprises a number of simple reactions that proceed concurrently. Thus, each tempering-resistance curve is the resultant of several component curves, each of which represents a simple reaction. It is not possible to investigate each of these reactions separately, but the tempering-resistance curves may be resolved into their components by the method that will now be described.

Second Method of Analysis. Martensitic Steels.

The process of tempering in martensitic steels may be divided into three reactions:-

(1) Separation of carbon from solution in alpha-iron.

As the resistance of a series of quenched steels is a linear function of the carbon content - that is, as the resistance is proportional to the amount of carbon in solution - the fall of resistance due to separation of carbon from solution must be proportional to the amount that separates.

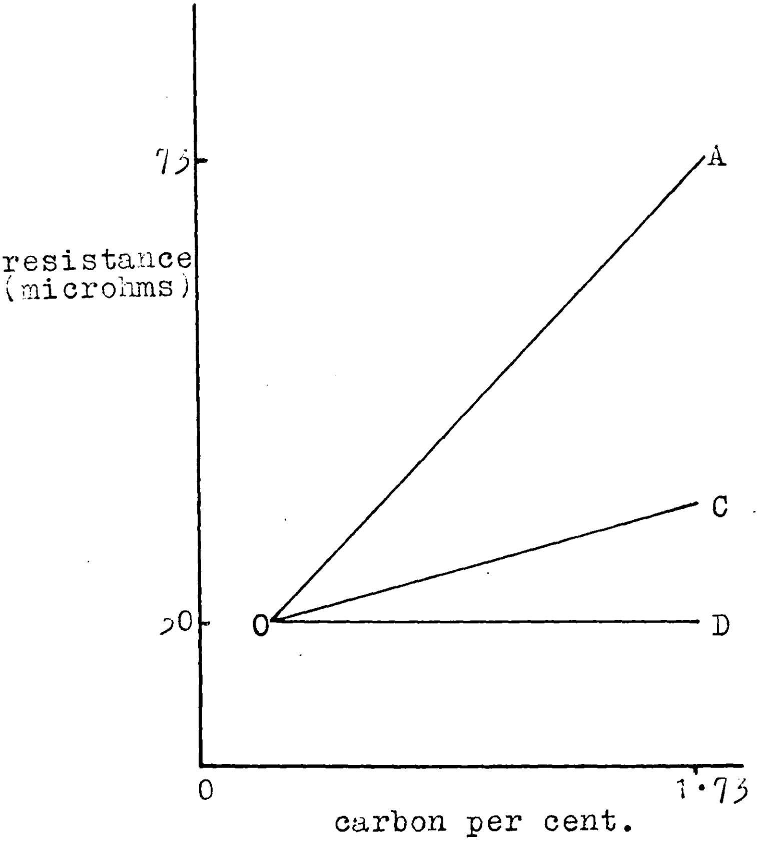
(2) Accumulation of cementite.

The carbon that is precipitated from solution during tempering is not removed from the system, but remains present as free carbide, the accumulation of which causes a decrease of conductivity proportional/

proportional to the volume of carbide formed. For the present purpose, it may be assumed that the accumulation of carbide gives rise to an increase of resistance proportional to the amount of carbon precipitated.

(3) Coagulation of the precipitated particles.

These three reactions combine to give the experimental tempering-curves. Reactions (1) and (2) give rise to an initial rapid fall of resistance, which is completed within a number of minutes; thereafter the gradual coagulation of the precipitated particles proceeds alone, and causes a slow and comparatively small decrease of resistance. As the third reaction is much slower than the others, its influence on the first rapid fall of resistance - due to reactions (1) and (2) - is negligible. The third reaction can therefore be considered to start when the others are completed, thus simplifying the analysis of the curves. The point at which the rapid fall of resistance is completed, is not clearly marked on the curves - any point between 20 and 30 minutes might be selected with equal justification. In analysing the curves, the junction between the different reactions was assumed to occur at 26 minutes; no difference would be caused by taking any time within a few minutes of this. It is/

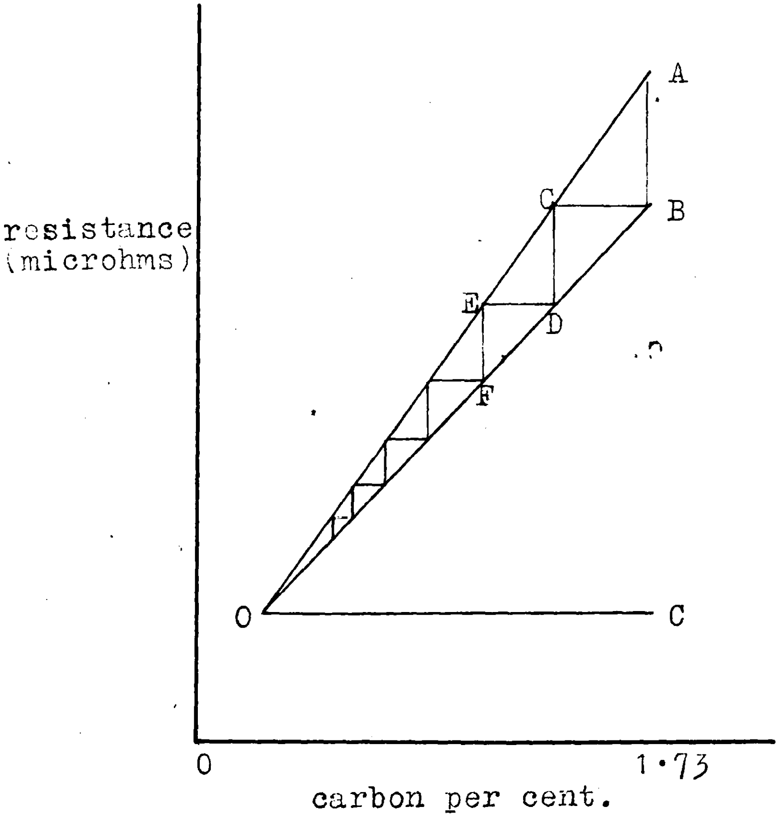

75
resistance (microhms)
50
A
C
O
D
0
1·75
carbon per cent.

is suggested therefore that when a martensitic steel is tempered at 240°C., the precipitation of cementite may be considered to be completed within 26 minutes. The subsequent change in the steel, which constitutes reaction (3), is small and proceeds slowly; as far as can be seen from the curves, it proceeds in the same manner and at the same slow rate in all tempering steels. As the influence of the third reaction on the tempering of steel is very small, compared with the change that occurs during the first rapid fall of resistance, the analysis of the tempering-curves of martensitic steels has been confined mainly to reactions (1) and (2). The analysis was performed as follows:-

The values of the resistance at 240°C. of the quenched carbon-steels were plotted against the carbon content, and a mean straight line (AO in Figure 38 and in Sketch opposite) was drawn through the points. Similarly, the straight line CO was drawn through the points representing the resistance at 240°C. of the same steels tempered for 26 minutes at 240°C. These lines meet at the point O whose coördinates are 30 microhms and 0·24 per cent. of carbon, indicating that after tempering for 26 minutes at 240°C., a carbon steel retains this amount of carbon in solution. Therefore, if the cementite on separating were removed from/

from the system, the resistance of the carbon steels, after tempering for 26 minutes, would be represented by the horizontal line OD (Figure 38 and Sketch); so the specific resistance of a quenched steel containing 1·73 per cent. of carbon would fall from 73 microhms (point A) to 30 microhms (point D) in that time. But the carbon is not removed from the system, and its effect in raising the resistance is represented by the line OC. Thus, the specific resistance of the tempered steel containing 1·73 per cent. of carbon is raised from 30 microhms (point D) to 47 microhms (point C) by the accumulated cementite. The actual effect caused by the tempering of this steel is therefore a fall of resistance from A to D (43 microhms) and a rise of resistance from D to C (17 microhms), which together give the experimentally observed fall from A to C (26 microhms). The experimental tempering-curves can therefore be resolved into their components by multiplying the observed decreases of resistance in successive intervals of time by $\frac{43}{26}$ to obtain the fall of resistance due to separation of carbon, and by $\frac{17}{26}$ to obtain the rise of resistance due to the accumulation of the free cementite formed.

When the time-resistance curves were resolved into their two components in the manner described above, it was found that the component curves representing the fall of resistance due to the separation of carbon/


A
C
B
E
D
F
O
C
resistance
(microhms)
0
1·73
carbon per cent.

carbon could be superimposed on each other so that each formed part of one general curve; but this curve extended only to 1.19 per cent. of carbon, for the steels containing more than that percentage contained austenite. Before proceeding to the analysis of the partly-austenitic steels, it was necessary that the general curve for martensite should be extended from 1.19 per cent. to 1.73 per cent. of carbon.

In the Sketch opposite this page (and in Figure 38), the line OA represents the specific resistance of the quenched steels; OC represents the specific resistance of the same steels after tempering for 26 minutes at 240°C., assuming the precipitated carbon to be removed from the system. If the rate of separation of carbon from solution in martensite is proportional to the concentration of carbon in solution,* the resistance of the steels at any time x between 0.0 minutes and 26 minutes should be given by a straight line such as OB opposite. From the lines OA and OB, the general curve for the fall of resistance due to separation/

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* The rate of separation of carbon is not proportional to the total amount of carbon in the steel, but more accurately to the amount of carbon in solution minus 0.24 per cent. - this being the amount that remains in solution after tempering for 26 minutes at 240°C.

separation of carbon from martensite can be obtained by a simple geometrical method. Thus, by stepping down the lines as shown in the Sketch, the decreases of resistance in successive intervals of time may be found. The vertical lines between OA and OB represent the successive decreases of resistance in successive equal time-intervals; for example, AB represents the fall of resistance caused by tempering a steel containing 1·73 per cent. of carbon for time x - that is, in the first x seconds after the commencement of tempering: CD represents the fall of resistance during the next x seconds: EF, the fall during the following x seconds ......etc. Similarly, the horizontal lines represent the successive decreases in the amount of carbon in solution.

The theoretical general curve for the tempering of martensite containing up to 1·73 per cent. of carbon was obtained in this manner, by plotting the fall of resistance against time. This curve extended to 1·73 per cent. of carbon; but the units of the time axis were unknown, because the time interval between OA and OB (x seconds) was not known. However, a curve extending to 1·19 per cent. of carbon had been derived from the experimental tempering curves (Paragraph 2, Page 120); and by adjusting the time axis of the new curve, the two curved were made to coincide accurately for/

for the interval representing the tempering of a steel containing 1·19 per cent. of carbon, thus determining the units of the time axis of the theoretical curve. In parenthesis, it may be mentioned that, the fact that these curves could be made to coincide accurately justifies the assumption that the rate of separation of carbon from solution is proportional to the amount of carbon in solution. The theoretical general curve showing the rate at which the resistance of martensite falls at 240°C. due to separation of carbon, was derived in this manner, and is shown in Curve A, Figure 39. Curve A, Figure 40, showing the rate of separation of carbon from solution, was obtained from Figure 39 by multiplying the decreases of resistance per minute by 0·03334, this being the cotangent of the line OA in Figure 38. The general curve showing the rise of resistance due to accumulation of the precipitated carbide was obtained in a similar manner; it is shown in Figure 41. The corresponding curves for the nickel, chromium, and nickel-chromium steels, are also shown in Figures 39,40, and 41. From Figure 40, it will be seen that in any steel within the limits of those used in the present investigation, martensite tempers at approximately the same rate.

By means of the curves plotted in Figures 39,40 and 41, the tempering-resistance curves of a number of the/

the steels were reconstructed, and are shown, together with the experimental curves, in Figures 42, 43, 44, and 45; the full lines represent the derived curves, and the points are those experimentally determined. Except for a few points at the beginning of each curve, the agreement is satisfactory. The lack of agreement at the beginning of the reaction is possibly due to the method of deriving the theoretical curve. The line OA in Figure 38 was drawn through the points representing the resistance at 240°C. of the quenched steels. As was explained on Page 100, these points were calculated from the known resistance of the quenched steels at 17°C., and their temperature-coefficient of resistance. It is probable that the discrepancy at the beginning of the curves may be due to the slight error introduced by this method of determining OA.

The experimental curve for the carbon steel containing 1·61 per cent. of carbon (Curve A, Figure 46) does not agree with the theoretical tempering-curve derived by the method described above (Curve B, Figure 46). As this steel when quenched consisted partly of martensite, the experimental curve represents the tempering of a mixture of austenite and martensite in unknown proportions, and is therefore the resultant of two curves, which represent respectively the tempering of martensite containing/

containing 1·61 per cent. of carbon, and of austenite of similar composition. One of these components may be obtained from the theoretical tempering-curve of martensite in a carbon steel containing 1·61 per cent. of carbon (Curve B, Figure 46) by the following method:-

The theoretical tempering-curve (B) shows that if the steel consisted entirely of martensite, the specific resistance would fall 20·6 microhms (from 70·0 to 49·4 microhms) in the first 12 minutes of tempering; from Curve A (Figure 46), it will be seen that the resistance actually fell 16·5 microhms (from 70·0 microhms to 53·5 microhms). The greater part of this fall was due to the tempering of martensite, which tempers rapidly, but part was also due to the slower tempering of austenite. The fall of resistance due to the latter reaction was obtained by extrapolation of Curve A (Figure 46), and was found to be approximately 4 microhms. The fall of resistance on the experimental curve due to tempering of martensite is therefore 12·5 microhms (16·5 microhms minus 4 microhms); if the steel had been entirely martensitic, the fall would have been 20·6 microhms. The change of resistance due to tempering of martensite in the quenched steel is therefore $\frac{12\cdot5}{20\cdot6}$ of the change that would have occurred if the steel had been totally martensitic. So by multiplying the decreases/

decreases of resistance per minute on Curve B (Figure 46) by $\frac{12.5}{20.6}$, the martensitic component of the tempering-curve was obtained; and by subtracting the decreases per minute on this component-curve, from the actual decreases per minute on the experimental curve (A), the austenitic component was obtained. The relation that the austenitic component bears to the curve that would be obtained if the steel were entirely austenitic, is similar to that which the martensitic component bears to the curve for pure martensite. Consequently, the curve that would represent the fall of resistance if the steel was totally austenitic, was obtained by multiplying the decreases per minute on the austenitic component by $\frac{20.6}{8.1}$. Curve C (Figure 46) was obtained in this manner.

The austenitic alloy-steels were analysed by this method; the experimental curves, martensitic curves, and austenitic curves, of these steels are shown in Figures 47, 48, and 49.

The austenitic curves (Curve C in each of the above Figures) were resolved into components by the same method as was used for the martensitic steels. The component curves representing the fall of resistance due to separation of carbon from solution in austenite are shown in Figure 50.

As the fall of resistance due to separation of carbon/

carbon from solution is proportional to the amount of carbon that separates, the rate at which the carbon separates may be calculated from the curves in Figure 50. Curves showing this rate for the various austenitic steels were calculated, and are shown in Figure 51. When these curves were plotted on suitable time-axes, they were found to coincide with the general curve for the separation of carbon from martensitic steels, as will be seen from Figures 52, 53, 54, and 55; this indicates that the precipitation of carbon from solution in gamma iron proceeds in a manner similar to that of the precipitation of carbon from solution in alpha iron, though at a smaller rate. It was found by this method that the austenite in a quenched carbon-steel containing 1.61 per cent. of carbon tempers 10 times more slowly than a martensitic carbon steel of the same carbon content: that in a quenched nickel-steel containing 1.06 per cent. of carbon, the austenite tempers 18 times more slowly than a martensitic carbon steel of the same carbon content: that the austenite in a chromium steel containing 1.73 per cent. of carbon tempers 66 times more slowly than a martensitic carbon steel of the same carbon content: and that in a nickel-chromium steel containing 1.28 per cent. of carbon, it tempers 80 times more slowly than a martensitic carbon steel of the same carbon content.

On page 123, it was shown that the rate at which martensite tempers in a carbon steel is not materially affected by the addition of moderate amounts of special elements. On the other hand, the rate of tempering of austenite, that is the rate at which austenite decomposes into alpha iron + carbide, is considerably reduced by the addition of special elements; the increased facility with which the critical points of steel may be lowered when special elements are present, may be due to this reduction in the velocity of the transformation.

CONCLUSIONS

In the introduction, it was stated that slight variations of heat-treatment have a greater effect on the physical properties of steel than is generally believed; justification of this statement is to be found in the text. Thus, all the heat-treatments described in Section 2 could be classified as "annealing;" yet, on account of graphitisation, the choice of the correct method of annealing is clearly of importance. In order to avoid graphitisation, a temperature such as 850°C. should be chosen for annealing, and the specimens should be heated to that temperature, soaked for a suitable length of time, and cooled not too slowly to room-temperature. The present series of steels after heating to 850°C. for 6 hours was not measurably graphitised; but if the percentage of manganese were lower than in these steels (0.2 per cent.), graphitisation would probably commence in less than that time. It should be sufficient to heat to 850°C. for one hour, and cool at a moderately slow rate: if the cooling is too slow, graphite may be deposited during cooling from the annealing/

annealing-temperature.

As was stated in Section 4, different workers have obtained different values for the resistance of annealed steels. Matthiessen and Le Chatelier found that the specific resistance of annealed steels bore a linear relation to the carbon content: Gumlich found that the resistance-concentration curve of annealed steels consisted of two intersecting curves, with a discontinuity at the eutectoid: Tamann states that the relation of resistance to carbon concentration is represented by two straight lines intersecting at the eutectoid composition, and that this is due to the different disposition of the cementite in hypoeutectoid and hypereutectoid steels. There can be little doubt that the discrepancies are due to the use of different methods of annealing. Unfortunately the method employed is seldom specified.

The effect produced by varying the method of quenching was described in Section 3. It was shown that the results obtained by quenching after the temperature had fallen to the quenching-temperature, were not the same as those obtained by quenching without first heating to a higher temperature. It is probable that if the steels were soaked at the quenching-temperature for a sufficient length of time to allow the specimens to attain equilibrium, either method/

method would give the same results - so long as graphitisation did not occur. The experiments on graphitisation showed that steels of low manganese content are liable to graphitise if heated to a high temperature and soaked at a lower temperature.

The customary method of investigating the process of tempering consists in heating intermittently to the tempering-temperature, and making observations at room-temperature between the periods of tempering - the method therefore involves periods of heating and of cooling between each observation. It is assumed that the only effect of the heating and cooling is to increase slightly the effective time of tempering; but the results obtained by the method of continuous measurement of resistance during tempering, make it clear that the intermittent method does not give a true representation of the reactions that occur at the tempering-temperature. Numerous investigators have shown: (a) that when an austenitic, or partly-austenitic steel is tempered at about 240°C. and cooled to room-temperature, the austenite appears to have been converted into martensite, a large proportion of which remains in the specimen untempered at room-temperature: and (b) that when the tempering-temperature is lower than 240°C., martensite tempers, but austenite is not decomposed. From this it has been deduced that at temperatures/

temperatures lower than 240°C., martensite tempers into troostite, and that at 240°C., austenite is converted into martensite, which apparently does not temper. This means that at any temperature below 240°C. martensite tempers more rapidly than austenite, but that at 240°C., austenite tempers more rapidly than martensite. The results obtained by the writer have shown that at 240°C., martensite tempers rapidly, the reaction being completed in a few minutes, and that austenite tempers slowly and progressively. From the tempering-resistance curves, it was shown that in all steels, martensite tempers rapidly, and that the rate of tempering is not affected by the presence of alloying elements; on the other hand, austenite tempers slowly - the austenite in a carbon steel containing 1.61 per cent. of carbon tempers 10 times more slowly than martensite of the same carbon content, and the rate of tempering of austenite is decreased by the addition of alloying elements such as nickel or chromium. As martensite tempers much more rapidly than austenite at 240°C., there could not be an accumulation of martensite from austenite at that temperature; the martensite found in the tempered austenitic specimens after cooling to room-temperature must be formed during cooling, and not at 240°C. as the intermittent method of tempering indicates. The error introduced by the intermittent method/

method of tempering is similar to that made by some investigators in assuming that a true representation of the state of an alloy at a high temperature may be obtained by quenching - an instance was quoted on page 67. The only certain method of finding out the constitution of a substance at a high temperature lies in examination of the substance at that temperature; this is certainly so if a change of phase occurs during cooling. Similarly, the nature of the reactions that occur during tempering, can be investigated satisfactorily only by examination of the tempering specimen at the tempering-temperature.

From Figure 34, it will be seen that when an austenitic chromium-steel was withdrawn from the tempering-bath after tempering for 320 minutes at 240°C., a considerable amount of martensite was formed during cooling to room-temperature; for on reheating to 240°C. the resistance decreased rapidly due to tempering of the martensite thus formed. The time taken for the specimen to cool from 240°C. to room-temperature was not more than 5 minutes; during that period, as will be seen from the uninterrupted tempering-resistance curve of the specimen (Point X to Point Y, Figure 34), the amount of tempering that would have occurred at 240°C. would have been very small - yet a considerable amount of austenite tempered to martensite during cooling. Austenite must therefore temper/

temper much more rapidly during cooling than it does at the tempering-temperature. From this, it is clear that austenite can temper in two different ways: slowly at a constant temperature: more rapidly when the temperature is falling. At a constant temperature, austenite, in effect, appears to temper directly to troostite; martensite is retained only when austenite is decomposed by cooling. The rapid change during cooling is of the same nature as that which occurs at the lower change-point, named A" by French metallurgists; the slow change at constant temperature is similar to that which occurs at Ar_1 (A').

It was stated above that the electrode-potential curves show that quenching from a falling temperature gives results that differ from those obtained by quenching from a rising temperature. That deduction can legitimately be made from the potential-concentration curves. Although the electrode potential of quenched steels may be affected to some extent by the strain caused by quenching, the effect is probably negligible, for the increase of potential caused by straining is apparently within the limits of experimental error (Turner and Jevons(23), and Desch(42)); consequently it may be assumed that the electrode potential of a quenched steel is governed solely by the concentration of carbon in solution. In any/

any case, the strain produced by quenching after cooling slowly to 850°C. and soaking for 5 minutes, should not be less than that produced by quenching after heating to 850°C. and soaking for 30 minutes; the difference of potential caused by varying the method of quenching is clearly due to a difference in the concentration of carbon in solution in the iron.

Although the interpretation of some of the features of the potential-concentration curves is open to question, still, in the opinion of the writer, the general interpretation of the curves is correct. The electrode potential of a quenched steel is a measure of the concentration of carbon in the least concentrated area of martensite exposed to the electrolyte, and is not affected by the presence of austenite in the specimen. If martensite were a homogeneous substance, as Whiteley (43) believes it to be, the concentration of carbon in the least concentrated area of a totally martensitic steel would be equal to the average carbon content of the steel, and could not be reduced by raising the quenching-temperature. The electrode-potential curves of quenched hypoeutectoid steels indicate that the concentration of carbon in the least concentrated area <u>is</u> reduced by raising the quenching-temperature; from this it follows that martensite is a heterogeneous substance. This does not prove conclusively that the austenite from which heterogeneous martensite/

martensite is formed, is itself heterogeneous, for a change of phase occurs during quenching. By the method described in Section 1, consistent readings of the electrode potential of steel can be obtained, and further experiments by this method should yield useful information; for if martensite is a heterogeneous substance, electrode-potential measurements constitute the most convenient method, perhaps the only method, of differentiating between the least concentrated areas and the more concentrated areas, and investigating them quantitatively.

As was stated in Section 2, the results obtained in the experiments on graphitisation do not show whether carbon exists in solution as atomic carbon or as carbide. It is commonly suggested that if carbon existed in the atomic state in solution, it would be retained in that state by quenching, and would be deposited as graphite on tempering;* this is a further example of the fallacy discussed on page 134. The only satisfactory method of investigating the state of carbon in solution lies in investigation of the solution, and not of its decomposition products; examination with X-rays indicates that carbon exists in the atomic state in solution.

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* In this connection, it is interesting to note that Tamann states that when a quenched steel is tempered, carbon is precipitated as graphite (finely divided temper-carbon), and that the temper-carbon is converted into cementite when the temperature is raised above 400°C.

The experiments on graphitisation show that when a solid iron-cementite alloy is heated, cementite does not decompose directly into graphite; but that graphitisation occurs by solution of cementite in austenite, and deposition of graphite from the solid solution. The experiments indicate that the equilibrium between iron and carbon may be represented fully by means of two superimposed diagrams, representing respectively stable equilibrium between iron and carbon, and metastable equilibrium between iron and cementite.

In conclusion, the writer wishes to state that the investigation described in this thesis forms part of a general scheme of research carried out in conjunction with Mr. J. M. Robertson in the Metallurgical Department of the Royal Technical College. The measurement of the resistance of the quenched steels at room-temperature, was performed in order to obtain the first point on the tempering-resistance curves; the determination of these curves could not be carried out by one observer, for the determination of each curve necessitated simultaneous measurements of the resistance of two specimens by means of two potentiometers, at close intervals of time during several hours. These experiments were performed in collaboration with Mr. J. M. Robertson, and are described also in the thesis submitted by him.

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TABLES

TABLE 1.

Composition of Specimens.

	C.	Mn.	P.	S.	Si.	Ni.	Cr.
				CHROMIUM STEELS .			
C.	0·35	0·24			0·17		1·56
D.	0·62	0·22			0·18		1·67
E.	0·80	0·28			0·20		1·67
F.	0·98	0·28			0·22		1·68
B.	1·73	0·30			0·21		1·65
				CARBON STEELS .			
1.	0·22	0·08			0·15		
2.	0·48	0·18			0·11		
3.	0·71	0·16			0·10		
4.	0·90	0·20			0·14		
5.	1·19	0·18			0·11		
6.	1·61	0·18			0·13		
7.	1·73	0·20			0·08		

TABLE 1(b).

Composition of Specimens.

	C.	Mn.	P.	S.	Si.	Ni.	Cr.
				NICKEL STEELS.			
2.	0·26	0·27	·014	·025	0·10	3·64	
4.	0·44	0·33	·015	·038	0·12	3·65	
5.	0·67	0·40	·012	·027	0·11	3·64	
6.	0·87	0·41	·013	·025	0·15	3·64	
7.	1·06	0·42	·013	·026	0·15	3·67	
				NICKEL-CHROMIUM STEELS.			
B.	0·51	0·22	0·02	0·05	0·126	3·52	1·72
G1.	1·28	0·26	0·02		0·150	3·46	1·80
G2.	1·50	0·26			0·190	3·46	1·80

TABLE 2(a).

Electrode Potential of Carbon Steels.

Carbon Percent.	Electrode Potential (volts). Quenched 1100°C.	Quenched 1000°C.	Quenched 950°C.	Quenched 850°C.	Quenched 780°C.
0·22	0·514	0·514	0·515	0·517	0·516
0·28	0·503	0·502	0·501	0·5005	0·500
0·48	0·4935	0·492	0·491	0·4895	0·488
0·71	0·486	0·484	0·484	0·4815	0·480
0·90	0·482	0·483	0·482	0·4805	0·479
1·19	0·465	0·476	0·477	0·478	0·4775
1·61	0·457	0·471	0·472	0·475	0·477
1·73	0·456	0·470	0·471	0·474	0·477

	Quenched 1100°C., Immersed in Liquid-Air.	Quenched 780°C., Immersed in Liquid-Air .
0·22	0·515	0·516
0·28	0·503	0·500
0·48	0·493	0·487
0·71	0·486	0·480
0·90	0·482	0·479
1·19	0·465	0·4765
1·61	0·456	0·476
1·73	0·4555	0·476

TABLE 2(b).

Electrode Potential of Carbon Steels.

Carbon Percent.	Electrode Potential (volts).	
	Heated to 1100°C. Quenched 850°C.	Heated to 1100°C. Quenched 780°C
0·22	0·5170	0·5150
0·28	0·5030	0·4990
0·48	0·4900	0·4880
0·71	0·4815	0·4805
0·90	0·4815	0·4800
1·19	0·4690	0·4750
1·61	0·4680	0·4730
1·73	0·4680	0·4730

	Quenched at 1100°C.		
	Immediately after Quenching.	48 hours after Quenching.	21 days after Quenching.
0·22	0·514		
0·28	0·5030	0·502	0·499
0·48	0·4935	0·493	0·489
0·71	0·486	0·485	0·481
0·90	0·482	0·482	0·478
1·19	0·465	0·463	0·459
1·61	0·457	0·4535	0·451
1·73	0·456	0·4525	0·450

TABLE 2(c).

Electrode Potential of Chromium Steels.

Steel Number.	Electrode Potential (volts)			
	Quenched 1100°C.	Quenched 1000°C.	Quenched 900°C.	Quenched 800°C.
C.	0·495	0·496	0·494	0·494
D.	0·485	0·483	0·481	0·479
E.	0·482	0·481	0·480	0·478
F.	0·477	0·477	0·476	0·474
B.	0·452	0·464	0·4675	0·470

	Quenched at 1100°C., and Immersed in Liquid-Air.
C.	0·496
D.	0·485
E.	0·482
F.	0·477
B.	0·450

TABLE 3 (a).

Specific Volume of Carbon Steels (Specimens 1, 2, 3, and 4)

Exp†	Treatment.	Specific Volume.			
		1	2	3	4
1.	Series "A" as rolled.	0·12741	0·12738	0·12751	0·12800
2.	"A" heated to 850°C. 1 hour.	0·12739	0·12737	0·12750	0·12761
3.	"A" reheated to 1100°C. and soaked at 900°C. for 1 hour.	0·12740	0·12737	0·12751	0·12762
4.	"A" reheated to 1100°C. and soaked at 850°C. for 3 hours.	0·12739	0·12738	0·12749	0·12762
5.	"A" reheated to 850°C. for 3 hrs.	0·12742	0·12737	0·12751	0·12760
6.	"A" reheated to 850°C. for 3 hrs.	0·12741	0·12737	0·12752	0·12762
7.	"A" reheated to 850°C. for 1 hour in air.	0·12740	0·12738	0·12751	0·12760
8.	Series "B" as rolled.	0·12741	0·12737	0·12751	0·12762
9.	"B" reheated to 850°C. for 3 hrs.	0·12742	0·12736	0·12752	0·12762
10.	"B" reheated to 850°C. for 3 hrs.	0·12742	0·12737	0·12751	0·12762

TABLE 3 (b).

Specific Volume of Carbon Steels (Specimens 5, 6, and 7)

Exp^t.	Treatment.	Specific Volume. 5.	 6.	 7.
1.	Series "A" as rolled.	0·12775	0·12790	0·12800
2.	"A" reheated to 850°C. 1 hour.	0·12776	0·12792	0·12797
3.	"A" reheated to 1100°C., and soaked at 900°C. for 1 hour .	0·12794	0·12894	0·12963
4.	"A" reheated to 1100°C., and soaked at 850°C. for 3 hours .	0·12813	0·12983	0·13068
5.	"A" reheated to 850°C. for 3 hrs.	0·12818	0·13010	0·13099
6.	"A" reheated to 850°C. for 3 hrs.	0·12823	0·13023	0·13125
7.	"A" reheated to 850°C. for 1 hour in air .	0·12839	0·13063	0·13162
8.	Series "B" as rolled .	0·12775	0·12791	0·12801
9.	"B" reheated to 850°C. for 3 hrs.	0·12772	0·12787	0·12790
10.	"B" reheated to 850°C. for 3 hrs.	0·12771	0·12785	0·12791

TABLE 3 (c).

Specific Volume of Carbon Steels (Specimens 1, 2, 3, and 4)

Exp^{t}.	Treatment	Specific Volume. 1	2	3	4
11.	Series "C" as rolled .	0·12743	0·12738	0·12752	0·12761
12.	"C" heated to 1100°C., and soaked at 850°C. for 3 hours .	0·12740	0·12739	0·12751	0·12760
13.	Series "D" as rolled .	0·12741	0·12738	0·12751	0·12762
14.	"D" heated to 1100°C., and soaked at 850°C. for 3 hours .	0·12741	0·12739	0·12750	0·12761
15.	Series "E" heated to 1100°C. 1 hour and soaked at 850°C. for 3 hrs.	0·12741	0·12738	0·12750	0·12761
16.	Series "F" heated to 1100°C. and soaked at 950°C. for 3 hours .	0·12741	0·12738	0·12751	0·12763
17.	Series "G" soaked at 950°C 6 hours.	0·12741	0·12738	0·12752	0·12762
18.	Series "H" heated to 1100°C. and soaked at 750°C. for 3 hours .	0·12742	0·12737	0·12751	0·12762
19.	Series "K" soaked at 750°C 3 hours.	0·12743	0·12738	0·12753	0·12763
20.	Series "L" heated to 950°C 3 hours, soaked at 850°C. for 3 hours .	0·12742	0·12738	0·12751	0·12762

TABLE 3 (d).

Specific Volume of Carbon Steels (Specimens 5, 6, and 7).

Exp^t.	Treatment	Specific Volume. 5.	6.	7.
11.	Series "C" as rolled.	0·12775	0·12788	0·12802
12.	"C" heated to 1100°C., and soaked at 850°C. for 3 hours.	0·12788	0·12805	0·12885
13.	Series "D" as rolled.	0·12775	0·12790	0·12800
14.	"D" heated to 1100°C., and soaked at 850°C. for 3 hours.	0·12792	0·12899	0·12980
15.	Series "E" heated to 1100°C. 1 hour and soaked at 850°C. for 3 hours	0·12805	0·12872	0·12983
16.	Series "F" heated to 1100°C. and soaked at 950°C. for 3 hours.	0·12782	0·12911	0·13005
17.	Series "G" soaked at 950°C 6 hours.	0·12784	0·12791	0·12796
18.	Series "H" heated to 1100°C., and soaked at 750°C. for 3 hours.	0·12781	0·12889	0·13021
19.	Series "K" soaked at 750°C 3 hours.	0·12774	0·12788	0·12790
20.	Series "L" heated to 950°C 3 hours, and soaked at 850°C. for 3 hrs.	0·12832	0·12880	0·12814

TABLE 4.

RESISTANCE OF STANDARD SPECIMEN at 17°C.

Resistance Microhms.	Temperature Reading	Corrected Resistance Microhms.
0·0010506	0·28	0·0010506
0·0010507	0·40	0·0010514
0·0010515	0·49	0·0010510
0·0010504	0·80	0·0010495
0·0010520	1·15	0·0010500
0·0010515	1·57	0·0010486
0·0010528	1·67	0·0010496
0·0010511	2·30	0·0010505
0·0010556	2·38	0·0010508
0·0010556	2·46	0·0010508
0·0010553	2·42	0·0010507
0·0010553	2·42	0·0010507
0·0010550	2·34	0·0010504
0·0010548	2·31	0·0010502
0·0010551	2·28	0·0010505
0·0010545	2·17	0·0010499
0·0010543	1·83	0·0010502
0·0010537	1·76	0·0010501
0·0010534	1·53	0·0010504
0·0010533	1·49	0·0010505
0·0010522	1·29	0·0010499
0·0010508	0·63	0·0010501
0·0010500	0·29	0·0010500
0·0010506	0·47	0·0010504
0·0010504	0·53	0·0010499
0·0010503	0·62	0·0010496
0·0010515	0·79	0·0010503
0·0010520	0·94	0·0010505
0·0010525	1·00	0·0010502
0·0010526	1·23	0·0010502
0·0010532	1·43	0·0010506
0·0010538	1·76	0·0010497
0·0010542	2·04	0·0010500
0·0010546	2·36	0·0010498
0·0010548	2·25	0·0010502
0·0010545	2·00	0·0010505
0·0010538	1·88	0·0010510
0·0010536	1·85	0·0010504
0·0010523	1·32	0·0010499
0·0010524	1·17	0·0010503
0·0010528	1·04	0·0010510
0·0010512	0·58	0·0010505
0·0010499	0·20	0·0010501
0·0010504	0·15	0·0010507
Mean Resistance		0·00105024

(temperature 0·28 = 17°C.)

TABLE 5(a).

Electrical Resistance of Carbon Steels.

Resistance (microhms).

Treatment	1.	2.	3.	4.	5	6.
Quenched 1100°C .	15·34	21·93	28·88	34·87	42·36	52·57
Quenched 1000°C.	15·46	21·94	29·65	36·06	42·60	52·13
Quenched 900°C.	15·30	21·90	29·65	36·06	42·60	52·13
Quenched 800°C.	15·03	23·31	29·55	36·83	40·17	42·39
Tempered at 240°C. after quenching from:-						
1100°C.	15·12	17·73	19·05	22·59	22·69	24·37
1000°C.	15·14	17·47	19·00	22·53	22·75	24·57
900°C.	-	-	19·17	21·48	22·94	25·22
Annealed	14·02	15·56	16·73	18·88	19·13	21·17

TABLE 5(b).

Electrical Resistance of <u>Nickel Steels</u>.

	Resistance (microhms).				
Treatment.	2.	4.	5.	6.	7.
Quenched 1100°C.	25·47	30·69	35·37	42·14	47·00
Quenched 1000°C.	25·61	30·42	36·88	42·55	47·78
Quenched 900°C.	24·86	30·74	36·91	44·15	48·99
Quenched 800°C.	25·20	30·52	38·06	44·63	43·72
Tempered at 240°C. after quenching from:-					
1100°C.	24·10	26·11	28·02	30·44	31·46
1000°C.	24·54	26·48	28·05	29·98	30·91
Annealed	22·53	24·37	25·34	27·62	27·73

TABLE 5(c).

Electrical Resistance of Chromium Steels.

	Resistance (microhms).				
Treatment	C.	D.	E.	F.	B.
Quenched 1100°C.	27·21	34·88	40·10	46·46	52·60
Quenched 1000°C.	27·42	36·43	41·99	46·42	61·11
Quenched 900°C.	27·57	37·07	41·55	46·69	47·47
Quenched 800°C.	33·42	37·15	40·74		45·85
Tempered at 240°C. after quenching from:-					
1100°C.	25·44	29·69	33·06	34·32	36·45
1000°C.	25·34	29·81	32·71	33·87	35·40
900°C.	25·64			33·69	
Annealed	19·75	20·45	21·29	22·58	23·99

TABLE 5(d).

Electrical Resistance of Nickel-Chromium Steels.

Treatment	Resistance (microhms). B.	G1.	G2.
Quenched 1100°C.	41·00	58·35	64·42
Quenched 1000°C.	40·95	57·23	65·85
Quenched 900°C.	40·75	57·61	59·76
Quenched 800°C.	40·95	50·55	52·38
Tempered at 240°C. after quenching from:-			
1100°C.	35·89	44·44	65·19
1000°C.	35·57	47·56	66·96
Annealed	27·06	32·46	32·96

REMAINING
TABLES
in
SEPARATE
VOLUME.

SOME PROPERTIES OF STEEL

VOL. II.

TABLES.

M. S. FISHER, B.Sc., A.R.T.C.

TABLE 6

TEMPERATURE
COEFFICIENT
of
RESISTANCE.

TABLE 6.

TEMPERATURE COEFFICIENT of RESISTANCE.

CHROMIUM STEELS

NUMBER	TEMPERATURE COEFFICIENT		
	Annealed	Tempered	Quenched
C	0·0029310	0·0024462	0·0022129
D	0·0026930	0·0021082	0·0018249
E	0·0025934	0·0020234	0·0019151
F	0·0025316	0·0018342	0·0013137
B	0·0028420	0·0020050	0·0015878

NICKEL STEELS

NUMBER	Annealed	Tempered	Quenched
2	0·0026970	0·0024803	0·0024021
4	0·0025761	0·0023684	0·0020667
5	0·0024302	0·0021950	0·0017326
6	0·0022909	0·0021182	0·0014437
7	0·0022560	0·0020080	0·0014048

TEMPERATURE COEFFICIENT of RESISTANCE

CARBON STEELS

NUMBER	TEMPERATURE COEFFICIENT	
	ANNEALED	QUENCHED
1	·004770	·0038077
2	·004220	·0028059
3	·0038012	·0022818
4	·0034759	·0018790
5	·0033511	·0015654
6	·0033575	·00162765

NICKEL-CHROME STEELS

NUMBER	QUENCHED	TEMPERED
B	·0015028	·0017225
G1	·00134145	·0014962
G2	·0016642	·0016642

MEASUREMENT of RESISTANCE during TEMPERING.

TABLE 7A.

CARBON STEELS
Quenched 900°C.

CARBON 3 Quenched 900°C

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	46·48	21	33·95		
½	39·80	22	33·90		
1	38·45	23	33·85		
1½	37·58	24	33·82		
2	37·23	25	33·79		
2½	36·91	27½	33·79		
3	36·53	30	33·75		
3½	36·28	35	33·70		
4	36·04	40	33·65		
4½	35·82	45	33·60		
5	35·71	50	33·55		
5½	35·58	55	33·55		
6	35·49	60	33·50		
6½	35·41	65	33·50		
7	35·28	70	33·50		
7½	35·20	75	33·46		
8	35·16	85	33·46		
8½	35·08	100	33·41		
9	35·02	115	33·41		
9½	34·92	125	33·41		
10	34·84	133	33·38		
11	34·77	149	33·38		
12	34·62	170	33·36		
13	34·50	182	33·36		
14	34·44	201	33·34		
15	34·34	216	33·33		
16	34·22	250	33·32		
17	34·18	302	33·31		
18	34·12	320	33·30		
19	34·06				
20	34·00				

CARBON 4 Quenched 900°C.

Time	Res.	Time	Res.
0	51·77	21	36·62
½	45·27	22	36·58
1	42·64	23	36·54
1½	41·36	24	36·51
2	40·75	25	36·48
2½	40·31	26	36·45
3	39·97	27	36·43
3½	39·62	28	36·41
4	39·41	29	36·39
4½	39·23	30	36·37
5	39·00	32½	36·33
5½	38·82	35	36·30
6	38·71	40	36·24
6½	38·58	45	36·16
7	38·48	50	36·08
7½	38·35	60	35·95
8	38·25	70	35·86
8½	38·14	85	35·79
9	38·02	100	35·79
9½	37·93	125	35·75
10	37·81	148	35·67
11	37·64	170	35·62
12	37·49	180	36·60
13	37·34	198	36·60
14	37·23	220	36·60
15	37·11	231	36·60
16	37·01	251	35·58
17	36·92	276	35·56
18	36·81	302	35·56
19	36·73	324	35·53
20	36·67	340	35·53

CARBON 5 Quenched 900°C.

Time	Res.	Time	Res.	Time	Res.	Time	Res.
0	59·22	21	40·75	321	38·64		
½	57·00	22	40·64	350	38·62		
1	54·03	23	40·57	362	38·60		
1½	51·51	24	40·51	377	38·60		
2	49·52	25	40·47				
2½	48·21	27½	40·39				
3	46·92	30	40·28				
3½	45·90	32½	40·16				
4	45·43	35	40·01				
4½	45·06	37½	39·86				
5	44·75	40	39·70				
5½	44·35	42½	39·61				
6	44·20	45	39·55				
6½	43·92	50	39·35				
7	43·73	60	39·22				
7½	43·54	70	39·16				
8	43·36	75	39·03				
8½	43·14	100	38·92				
9	42·86	110	38·87				
9½	42·75	120	38·82				
10	42·62	126	38·81				
11	42·44	133	38·80				
12	42·19	151	38·80				
13	41·90	176	38·78				
14	41·63	191	38·72				
15	41·47	220	38·70				
16	41·33	230	38·69				
17	41·20	234	38·69				
18	40·97	250	38·69				
19	40·93	275	38·64				
20	40·85	300	38·64				

CARBON 6 Quenched 900°C.

C6 900·

Time (mins)	Res. Microhms	Time	Res.
0	66·77	21	46·22
½	57·78	22	46·12
1	55·35	23	45·96
1½	54·25	24	45·85
2	53·58	25	45·75
2½	53·00	27½	45·51
3	52·52	30	45·28
3½	52·04	32½	44·95
4	51·70	35	44·84
4½	51·41	40	44·52
5	50·98	45	44·23
5½	50·73	50	43·96
6	50·49	60	43·58
6½	50·21	70	43·21
7	49·87	90	42·55
7½	49·73	102	42·28
8	49·50	115	42·13
8½	49·37	125	42·02
9	49·16	147	41·93
9½	48·97	155	41·85
10	48·77	198	41·66
11	48·51	214	41·64
12	48·23	225	41·60
13	47·96	249	41·57
14	47·71	260	41·55
15	47·45	273	41·55
16	47·22	311	41·55
17	47·00	349	41·53
18	46·78		
19	46·56		
20	46·35		

TABLE 7B

CARBON STEELS
Quenched 1000°C.

CARBON 3 Quenched 1000°C.

Time (mins)	Res. (micr.)	Time	Res.	Time	Res.
0	44·87	21	33·79		
½	41·25	22	33·76		
1	39·00	23	33·75		
1½	37·47	24	33·72		
2	36·81	25	33·69		
2½	36·48	26	33·67		
3	36·23	27	33·65		
3½	36·02	28	33·64		
4	35·88	29	33·63		
4½	35·73	30	33·62		
5	35·60	35	33·58		
5½	35·48	37½	33·58		
6	35·35	42	33·56		
6½	35·22	47½	33·52		
7	35·10	50	33·48		
7½	35·00	52½	33·48		
8	34·85	55	33·44		
8½	34·76	60	33·44		
9	34·70	65	33·40		
9½	34·64	80	33·36		
10	34·60	91	33·33		
10½	34·52	100	33·33		
11	34·48	120	33·28		
12	34·40	160	33·23		
13	34·30	175	33·23		
14	34·21	218	33·20		
15	34·02	250	33·19		
16	33·97	275	33·18		
17	33·93	295	33·17		
18	33·90	315	33·15		
19	33·87	375	33·11		
20	33·83				

CARBON 4 Quenched 1000°C.

Time (mins)	Res. Microhms	Time	Res.
0	51·29	21	38·06
½	43·35	22	38·03
1	41·62	23	38·00
1½	40·87	24	37·96
2	40·50	25	37·93
2½	40·22	26	37·89
3	39·99	27	37·83
3½	39·75	28	37·77
4	39·57	29	37·74
4½	39·45	30	37·72
5	39·32	35	37·53
5½	39·23	40	37·44
6	39·11	45	37·40
6½	39·02	50	47·35
7	38·95	55	37·31
7½	38·89	60	37·25
8	38·82	65	37·21
8½	38·75	70	37·18
9	38·68	75	37·16
9½	38·62	80	37·16
10	38·56	85	37·14
11	38·51	100	37·11
12	38·48	115	37·10
13	38·44	125	37·08
14	38·38	135	37·06
15	38·34	175	37·06
16	38·30	200	37·04
17	38·25	220	37·04
18	38·21	275	37·02
19	38·15	295	37·02
20	38·09	350	37·01

CARBON 6 Quenched 1000°C.

C6 1000.

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	71·21	21	48·28	240	40·90
½	64·80	22	48·06	250	40·90
1	62·25	23	47·86	265	40·88
1½	60·49	24	47·72	280	40·84
2	59·32	25	47·53	296	40·84
2½	58·33	26	47·43	325	40·78
3	57·65	27	47·31	350	40·78
3½	57·13	28	47·20		
4	56·56	29	47·08		
4½	56·17	31	46·84		
5	55·78	32	46·65		
5½	55·24	33	46·53		
6	54·81	34	46·42		
6½	54·39	36	46·14		
7	53·81	40	45·71		
7½	53·05	45	45·11		
8	52·51	50	44·59		
8½	52·08	61	43·56		
9	51·78	70	42·63		
9½	51·50	80	42·23		
10	51·37	90	41·78		
11	50·96	100	41·53		
12	50·55	109	41·45		
13	50·06	120	41·33		
14	49·78	137	41·19		
15	49·52	150	41·08		
16	49·34	167	41·01		
17	49·09	181	40·98		
18	48·92	195	40·95		
19	48·69	210	40·93		
20	48·46	225	40·93		

TABLE 7C.

CARBON STEELS
Quenched 1100°C.

C1 1100°

CARBON 1 Quenched 1100°C.

Time (mins)	Res. Microhms	Time	Res.
0	28·40	21	30·55
½	29·50	22	30·53
1	30·53	23	30·52
1½	31=08	24	30·52
2	31·35	25	30·50
2½	31·30	30	30·43
3	31·25	35	30·37
3½	31·12	40	30·28
4	31·04	45	30·25
4½	31·00	50	30·23
5	30·85	55	30·19
5½	30·83	60	30·19
6	30·80	70	30·18
6½	30·79	80	30·18
7	30·79	90	30·17
7½	30·78	100	30·16
8	30·77	110	30·16
8½	30·75	120	30·16
9	30·75	140	30·15
9½	30·73	160	30·13
10	30·73	180	30·12
11	30·71	200	30·11
12	30·70	220	30·11
13	30·70	240	30·09
14	30·67	260	30·08
15	30·65		
16	30·62		
17	30·62		
18	30·60		
19	30·57		
20	30·56		

Completely

Tempered.....25·41

C2 1100°

CARBON 2 Quenched 1100°C.

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	35·95	21	32·43	250	31·79
½	35·48	25	32·39	260	31·77
1	3~~5~~4·90	30	32·34		
1½	34·73	35	32·28		
2	34·42	40	32·23		
2½	34·21	45	32·20		
3	34·01	50	32·19		
3½	33·79	55	32·17		
4	33·68	60	32·17		
4½	33·57	65	32·15		
5	33·46	70	32·13		
5½	33·40	75	32·12		
6	33·37	80	32·12		
6½	33·32	85	32·08		
7	33·26	90	32·08		
7½	33·20	95	32·06		
8	33·16	100	32·05		
8½	33·10	110	32·02		
9	33·06	120	32·01		
9½	33·00	130	31·97		
10	32·94	140	31·97		
11	32·82	150	31·94		
12	32·77	160	31·93		
13	32·70	170	31·91		
14	32·63	180	31·90		
15	32·57	190	31·88		
16	32·52	200	31·85		
17	32·49	210	31·84		
18	32·46	220	31·82		
19	32·45	230	31·80		
20	32·44	240	31·79		

Completely

Tempered..28·77

CARBON 4 Quenched 1100°C.

Time	Res.	Time	Res.	Time	Res.
0	49·60	21	38·54	185	37·27
½	46·67	22	38·51	195	37·23
1	44·83	23	38·48	205	37·20
1½	43·60	24	38·43	215	37·17
2	42·85	25	38·38	225	37·13
2½	42·32	26	38·34	250	37·10
3	41·75	27	38·29	275	37·06
3½	41·46	28	38·24		
4	41·21	29	38·19		
4½	40·93	30	38·14		
5	40·76	35	37·96		
5½	40·60	37½	37·90		
6	40·45	40	37·81		
6½	40·32	42½	37·79		
7	40·18	45	37·76		
7½	40·02	47½	37·75		
8	39·94	50	37·73		
8½	39·80	53	37·73		
9	39·71	57½	37·68		
9½	39·60	60	37·64		
10	39·51	67	36·60		
11	39·44	75	37·60		
12	39·31	80	37·54		
13	39·19	90	37·54		
14	39·04	100	37·50		
15	38·95	110	37·46		
16	38·87	116	37·46		
17	38·80	135	37·35		
18	38·73	155	37·32		
19	38·64	165	37·31		
20	38·58	175	37·30		

CARBON 5 Quenched 1100°C.

Time	Res.	Time	Res.	Time	Res.
0	57·24	21	40·43	240	38·90
½	53·05	22	40·36		
1	49·97	23	40·30		
1½	47·53	24	40·26		
2	46·00	25	40·22		
2½	44·82	26	40·20		
3	44·32	27½	40·16		
3½	43·81	30	40·09		
4	43·53	32½	40·00		
4½	43·25	35	39·90		
5	43·01	40	39·66		
5½	42·77	42½	39·56		
6	42·56	45	39·48		
6½	42·40	50	39·40		
7	42·17	55	39·31		
7½	41·99	60	39·24		
8	41·80	65	39·21		
8½	41·63	70	39·18		
9	41·51	80	39·16		
9½	41·42	95	39·10		
10	41·30	108	38·98		
11	41·18	120	38·98		
12	41·08	130	38·98		
13	41·02	140	38·96		
14	40·91	150	38·94		
15	40·84	160	38·94		
16	40·77	180	38·94		
17	40·70	195	38·94		
18	40·63	210	38·92		
19	40·56	219	38·92		
20	40·49	230	38·90		

C6 1100°

CARBON 6 Quenched 1100°C.

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	74·35	21	50·16	Completely	
½	67·51	22	49·83	Tempered...	
1	65·26	23	49·57		37·11
1½	63·32	24	49·35		
2	61·96	25	49·11		
2½	60·75	26	48·82		
3	59·73	27	48·69		
3½	58·84	28	48·53		
4	58·13	29	48·38		
4½	57·62	30	48·26		
5	57·02	35	47·41		
5½	56·51	40	46·58		
6	56·09	45	45·83		
6½	55·68	50	45·25		
7	55·39	60	44·28		
7½	55·04	70	43·52		
8	54·72	80	42·84		
8½	54·51	90	42·49		
9	54·22	100	42·21		
9½	54·01	110	42·03		
10	53·79	140	41·79		
11	53·41	160	41·74		
12	53·05	180	41·68		
13	52·59	200	41·52		
14	52·36	220	41·47		
15	52·03	230	41·43		
16	51·70	240	41·40		
17	51·30	250	41·37		
18	51·03	260	41·35		
19	50·67	306	41·24		
20	50·43				

TABLE 7D

NICKEL STEELS
Quenched 1000°C.

N 2 1000

NICKEL 2 Quenched 1000°C.

Time (mins)	Res. Microhms	Time	Res.
0	39·49	21	38·60
½	39·75	22	38·59
1	40·02	23	38·59
1½	39·80	24	38·57
2	39·69	25	38·56
2½	39·49	26	38·55
3	39·26	27	38·55
3½	39·11	28	38·55
4	39·03	29	38·55
4½	38·87	30	38·53
5	38·78	35	38·50
5½	38·76	40	38·47
6	38·75	45	38·44
6½	38·74	50	38·44
7	38·73	55	38·38
7½	38·73	70	38·32
8	38·71	80	38·32
8½	38·71	98	38·27
9	38·70	120	38·22
9½	38·70	131	38·22
10	38·70	139	38·18
11	38·68	170	38·15
12	38·68	185	38·14
13	38·68	200	38·12
14	38·68	215	38·12
15	38·65	226	38·12
16	38·65	249	38·11
17	38·62	273	38·11
18	38·62	294	38·09
19	38·62	312	38·09
20	38·61		

Completely
Tempered....35·93

NICKEL 4 Quenched 1000°C.

N4 1000.

Time (min)	Res. Microhms	Time	Res.	Time	Res.
0	44·59	21	41·49	230	40·72
½	43·94	22	41·44	247	40·71
1	43·48	23	41·41	255	40·70
1½	43·19	24	41·39	270	40·68
2	42·93	25	41·35	295	40·68
2½	42·79	26	41·32	302	40·66
3	42·58	27	41·30	320	40·65
3½	42·49	28	41·28	Completely	
4	42·34	29	41·26	Tempered...	38·44
4½	42·29	30	41·25		
5	42·22	32½	41·23		
5½	42·17	35	41·20		
6	42·12	40	41·18		
6½	42·06	45	41·16		
7	41·99	50	41·13		
7½	41·93	55	41·11		
8	41·88	60	41·09		
8½	41·82	70	41·06		
9	41·80	75	41·03		
9½	41·78	90	40·98		
10	41·75	100	40·92		
11	41·73	115	40·88		
12	41·71	126	40·85		
13	41·66	145	40·83		
14	41·64	150	40·81		
15	41·61	160	40·80		
16	41·59	174	40·79		
17	41·57	186	40·77		
18	41·55	195	40·76		
19	41·53	201	40·74		
20	41·52	220	40·72		

NICKEL 5 Quenched 1000°C.

N5 1000°

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	47·15	21	41·86	210	40·61
½	45·18	22	41·81	220	40·58
1	44·51	23	41·75	230	40·56
1½	44·08	24	41·69	240	40·54
2	43·81	25	41·62	250	40·51
2½	43·68	26	41·57	260	40·50
3	43·49	27	41·55	Completely	
3½	43·33	28	41·52	Tempered...	
4	43·22	29	41·50		39·18
4½	43·11	30	41·48		
5	43·05	40	41·44		
5½	42·97	45	41·41		
6	42·94	50	41·37		
6½	42·86	55	41·32		
7	42·82	60	41·27		
7½	42·77	65	41·23		
8	42·71	70	41·18		
8½	42·65	75	41·15		
9	42·59	80	41·11		
9½	42·54	90	41·88		
10	42·50	100	41·03		
11	42·46	110	40·99		
12	42·40	120	40·96		
13	42·34	130	40·91		
14	42·27	140	40·85		
15	42·20	150	40·81		
16	42·15	160	40·77		
17	42·09	170	40·73		
18	42·04	180	40·69		
19	42·00	190	40·66		
20	41·94	200	40·64		

NICKEL 6 Quenched 1000°C.

N6 1000.

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	56·38	21	47·33	284	44·20
½	54·46	22	47·20	310	44·18
1	52·63	23	47·10	Completely	
1½	51·29	24	47·01	Tempered..	41·87
2	50·57	25	46·92		
2½	50·08	26	46·85		
3	49·83	27	46·79		
3½	49·69	28	46·74		
4	49·54	29	46·66		
4½	49·43	30	46·61		
5	49·36	32½	46·45		
5½	49·28	35	46·23		
6	49·20	37½	46·09		
6½	49·08	40	46·01		
7	49·01	45	45·91		
7½	49·94	50	45·80		
8	48·86	55	45·74		
8½	48·78	60	45·62		
9	48·70	65	45·48		
9½	48·62	70	45·40		
10	48·53	75	45·32		
11	48·41	90	45·18		
12	48·29	100	45·01		
13	48·18	122	44·83		
14	48·06	140	44·67		
15	47·94	159	44·52		
16	47·80	182	44·43		
17	47·69	200	44·36		
18	47·58	232	44·27		
19	47·49	248	44·25		
20	47·44	270	44·23		

NICKEL 7 Quenched 1000°C.

N7 1000.

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	62·88	21	49·45	314	44·81
½	62·52	22	49·26	Completely	
1	57·21	23	49·14	Tempered...	41·80
1½	54·36	24	49·02		
2	53·62	25	48·93		
2½	53·27	26	48·76		
3	53·01	27	48·69		
3½	52·79	28	48·56		
4	52·64	29	48·51		
4½	52·45	30	48·44		
5	52·30	35	48·02		
5½	52·22	40	47·68		
6	52·03	45	47·39		
6½	51·92	50	47·02		
7	51·80	55	46·63		
7½	51·71	64	46·42		
8	51·59	70	46·08		
8½	51·51	81	45·71		
9	51·44	92	45·50		
9½	51·33	101	45·43		
10	51·26	114	45·33		
11	51·09	126	45·26		
12	50·91	141	45·19		
13	50·76	159	45·08		
14	50·58	171	45·01		
15	50·43	186	44·98		
16	50·26	200	44·94		
17	50·09	214	44·90		
18	49·92	245	44·90		
19	49·75	269	44·86		
20	49·63	295	44·83		

TABLE 7E.

NICKEL STEELS
Quenched 1100°C.

NICKEL 2 Quenched 1100°C.

N2 1100·

Time (mins)	Res. Microhms	Time	Res.
0	39·25	21	37·63
½	39·47	22	37·63
1	39·16	23	37·63
1½	38·28	24	37·64
2	38·02	25	37·63
2½	37·83	26	37·63
3	37·80	30	37·62
3½	37·76	35	37·61
4	37·74	40	37·62
4½	37·72	50	37·59
5	37·70	60	37·59
5½	37·70	75	37·58
6	37·70	104	37·58
6½	37·69	140	37·56
6½	37·69	191	37·56
7	37·69	225	37·56
7½	37·68	250	37·55
8	37·68	275	37·55
8½	37·68	Completely	
9	37·70	Tempered...35·93	
9½	37·67		
10	37·67		
11	37·67		
12	37·67		
13	37·65		
14	37·66		
15	37·65		
16	37·64		
17	37·64		
18	37·64		
19	37·64		
20	37·64		

NICKEL 4 Quenched 1100°C.

N4 1100·

Time	Res.	Time	Res.
(min)	Microhms		
0	45·98	21	40·98
½	44·69	22	40·96
1	43·78	23	40·92
1½	43·02	24	40·90
2	42·52	25	40·86
2½	42·29	27½	40·81
3	42·09	30	40·74
3½	41·97	35	40·70
4	41·92	40	40·62
4½	41·81	45	40·56
5	41·73	50	40·51
5½	41·69	55	40·47
6	41·62	64	40·43
6½	41·56	70	40·40
7	41·51	76	40·37
7½	41·48	82	40·32
8	41·41	94	40·30
8½	41·34	102	40·26
9	41·30	124	40·18
9½	41·28	140	40·15
10	41·24	169	40·12
11	41·20	190	40·08
12	41·23	204	40·06
13	41·17	225	40·04
14	41·14	249	40·02
15	41·09	260	40·00
16	41·08	Completely	
17	41·06	Tempered...	38·44
18	41·04		
19	41·92		
20	41·00		

NICKEL 5 Quenched 1100°C.

N5 1100

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	49·42	21	44·13	200	43·08
½	47·49	22	44·10	210	43·06
1	46·50	23	44·08	220	43·06
1½	45·85	24	44·06	230	43·04
2	45·53	25	44·04	240	43·03
2½	45·44	26	44·02	250	43·03
3	45·27	27	44·00	260	43·02
3½	45·13	30	43·96	Completely	
4	45·04	35	43·90	Tempered...	
4½	44·98	40	43·82		39·18
5	44·93	45	43·75		
5½	44·87	50	43·70		
6	44·82	55	43·66		
6½	44·76	60	43·61		
7	44·69	65	43·57		
7½	44·63	70	43·54		
8	44·58	75	43·50		
8½	44·55	80	43·45		
9	44·53	85	43·41		
9½	44·50	90	43·38		
10	44·48	95	43·33		
11	44·46	100	43·30		
12	44·42	110	43·28		
13	44·38	120	43·25		
14	44·35	130	43·22		
15	44·32	140	43·19		
16	44·27	150	43·16		
17	44·23	160	43·14		
18	44·21	170	43·13		
19	44·19	180	43·11		
20	44·16	190	43·09		

NICKEL 6 Quenched 1100°C.

N6 1100.

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	55·82	21	47·42	215	44·86
½	53·01	22	47·36	239	44·81
1	52·13	23	47·32	258	44·76
1½	51·69	24	47·27	283	44·71
2	51·34	25	47·23	Completely	
2½	51·04	26	47·17	Tempered...	41·87
3	50·86	27	47·08		
3½	50·67	28	47·02		
4	50·50	29	46·90		
4½	50·36	30	46·84		
5	50·22	32½	46·76		
5½	50·03	35	46·64		
6	49·94	37½	46·51		
6½	49·76	40	46·43		
7	49·63	42½	46·36		
7½	49·51	45	46·24		
8	49·39	47½	46·11		
8½	49·25	50	45·98		
9	49·13	55	45·87		
9½	49·02	65	45·80		
10	48·91	70	45·73		
11	48·72	77	45·61		
12	48·51	88	45·42		
13	48·32	99	45·31		
14	48·22	106	45·26		
15	48·04	125	45·12		
16	47·92	147	45·01		
17	47·81	160	44·98		
18	47·69	166	44·96		
19	47·56	181	44·93		
20	47·49	198	44·90		

NICKEL 7 Quenched 1100°C.

N7 1100C.

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	61·87	21	51·22	231	46·11
½	61·79	22	51·11	241	45·93
1	59·23	23	51·03	250	45·84
1½	56·76	24	50·94	266	45·74
2	55·21	25	50·82	275	45·69
2½	54·48	26	50·76	302	45·65
3	54·02	27	50·69	324	45·60
3½	53·78	28	50·60		
4	53·61	29	50·52		
4½	53·50	30	50·45		
5	53·44	35	50·10		
5½	53·31	40	49·70		
6	53·23	45	49·42		
6½	53·12	50	49·11		
7	53·06	55	48·82		
7½	53·01	60	48·70		
8	52·95	65	48·46		
8½	52·88	70	48·27		
9	52·79	75	48·14		
9½	52·73	86	47·88		
10	52·65	93	47·76		
11	52·55	102	47·49		
12	52·40	115	47·35		
13	52·25	125	47·19		
14	52·08	132	47·10		
15	51·93	145	47·01		
16	51·77	155	46·83		
17	51·64	167	46·72		
18	51·52	185	46·54		
19	51·40	200	46·39		
20	51·28	219	46·30		

Completely

Tempered...41·80

TABLE 7F

CHROMIUM STEELS
Quenched 900°C.

CHROME C Quenched 900°C.

Time (mins)	Res. Microhms	Time	Res.
0	41·28	21	40·81
½	43·50	22	40·76
1	44·52	23	40·70
1½	45·05	24	40·65
2	44·95	25	40·60
2½	44·79	27½	40·50
3	44·68	30	40·43
3½	44·22	35	40·34
4	44·03	40	40·22
4½	43·77	45	40·14
5	43·52	50	40·08
5½	43·31	55	39·97
6	43·16	60	39·90
6½	43·01	65	39·86
7	42·81	70	39·81
7½	42·63	85	39·76
8	42·51	95	39·70
8½	42·44	110	39·63
9	42·27	126	39·60
9½	42·14	141	39·56
10	42·03	152	39·53
11	41·82	175	39·53
12	41·71	202	39·50
13	41·60	227	39·48
14	41·51	250	39·48
15	41·38	270	39·45
16	41·26	303	39·42
17	41·14	Completely	
18	41·02	Tempered...	32·77
19	40·96		
20	40·90		

CHROME C Quenched 900°C.

C 900.

Time (mins)	Res. Microhms	Time	Res.
0	41·28	21	40·81
½	43·50	22	40·76
1	44·52	23	40·70
1½	45·05	24	40·65
2	44·95	25	40·60
2½	44·79	27½	40·50
3	44·68	30	40·43
3½	44·32	35	40·34
4	44·03	40	40·22
4½	43·77	45	40·14
5	43·52	50	40·08
5½	43·31	55	39·97
6	43·16	60	39·90
6½	43·01	65	39·86
7	42·81	70	39·81
7½	42·63	85	39·76
8	42·51	95	39·70
8½	42·44	110	39·63
9	42·27	126	39·60
9½	42·14	141	39·56
10	42·03	152	39·53
11	41·82	175	39·53
12	41·71	202	39·50
13	41·60	227	39·48
14	41·51	250	39·48
15	41·38	270	39·45
16	41·26	303	39·42
17	41·14	Completely	
18	41·02	tempered...32·77	
19	40·96		
20	40·90		

CHROME F Quenched 900°C.

F 900°

Time (mins)	Res. Microhms	Time	Res.
0	62·56	21	50·32
½	62·49	22	50·25
1	61·90	23	50·20
1½	54·75	24	50·14
2	53·70	25	50·06
2½	53·29	26	50·01
3	53·01	27	49·94
3½	52·82	28	49·86
4	52·66	30	49·72
4½	52·52	32½	49·57
5	52·38	35	49·44
5½	52·24	37½	49·25
6	52·12	40	49·09
6½	52·03	42½	49·00
7	51·90	45	48·92
7½	51·78	50	48·74
8	51·69	58	48·58
8½	51·60	65	48·41
9	51·52	70	48·33
9½	51·44	81	48·26
10	51·33	100	48·02
11	51·22	116	47·93
12	51·03	145	47·84
13	50·91	156	47·80
14	50·77	180	47·71
15	50·68	200	47·62
16	50·62	218	47·56
17	50·56	254	47·53
18	50·51	303	47·50
19	50·45	316	47·50
20	50·38	Completely Tempered..35·41	

TABLE 7G.

CHROMIUM STEELS
Quenched 1000°C.

Chrome D Quenched 1000°C.

D 1000

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	51·36	21	46·70	409	43·98
½	51·22	22	46·70	425	43·95
1	50·79	23	46·67	460	43·92
1½	49·64	24	46·65		
2	49·06	25	46·60		
2½	48·72	26	46·54		
3	48·48	27	46·50		
3½	48·32	28	46·50		
4	48·16	29	46·47		
4½	48 04	30	46·42		
5	47·93	35	46·36		
5½	47·82	40	46·20		
6	46·75	45	46·03		
6½	47·63	50	45·90		
7	47·52	55	45·84		
7½	47·47	65	45·66		
8	47·42	75	45·45		
8½	47·35	90	45·31		
9	47·31	100	45·25		
9½	47·25	115	45·09		
10	47·19	130	44·93		
11	47·12	152	44·66		
12	47·04	175	44·54		
13	47·01	201	44·45		
14	46·97	220	44·40		
15	46·90	245	44·35		
16	46·83	298	44·28		
17	46·79	325	44·16		
18	46·77	350	44·09		
19	46·75	366	44·05		
20	46·72	381	44·02		

Completely Tempered... 33·23

CHROME E Quenched 1000°C.

E 1000

Time (mins)	Res. Microhms	Time	Res.
0	60·06	21	48·78
½	55·25	22	48·73
1	53·48	23	48·69
1½	52·07	25	48·63
2	51·32	27½	48·58
2½	50·86	30	48·52
3	50·56	35	48·45
3½	50·32	40	48·36
4	50·11	45	48·24
4½	49·93	50	48·08
5	49·78	55	47·96
5½	49·62	75	47·73
6	49·57	90	47·61
6½	49·53	110	47·52
7	49·50	125	47·40
7½	49·46	135	47·40
8	49·40	145	47·35
8½	49·37	165	47·30
9	49·32	186	47·26
9½	49·28	200	47·17
10	49·24	221	47·03
11	49·19	251	47·00
12	49·12	270	46·95
13	49·05	Completely	
14	49·00	Tempered...	34·12
15	48·95		
16	48·92		
17	48·90		
18	48·90		
19	48·85		
20	48·81		

F 1000.

CHROME F Quenched 1000°C.

Time (mins)	Res. Microhms	Time	Res.
0	62·21	21	50·49
½	62·00	23	50·41
1	61·47	24	50·35
1½	55·42	25	50·27
2	54·01	27½	50·09
2½	53·66	30	50·01
3	53·42	35	49·87
3½	53·17	40	49·78
4	52·94	45	49·64
4½	52·79	50	49·51
5	52·68	55	49·39
5½	52·52	60	49·30
6	52·41	68	49·21
6½	52·31	78	49·02
7	52·23	90	48·82
7½	52·12	99	48·61
8	52·01	110	48·51
8½	51·90	125	48·47
9	51·78	145	48·41
9½	51·67	171	48·33
10	51·56	200	48·17
11	51·44	230	48·09
12	51·30	261	48·07
13	51·18	286	48·03
14	51·06	316	48·00
15	50·96	326	48·00
16	50·83	342	47·98
17	50·77	359	47·98
18	50·70	372	47·95
19	50·65	Completely	
20	50·57	Tempered..	35·41

CHROME B Quenched 1000°C.

B 1000·

Time (mins)	Res. Microhms	Time	Res.	Time	Res.	Time	Res.
0	82·90	21	76·08	134	68·23	531	52·61
½	80·70	22	76·01	140	67·76	550	52·59
1	79·98	23	75·94	146	67·39	571	52·36
1½	79·48	24	75·85	151	67·04	Completely	
2	79·06	25	75·78	163	66·45	Tempered...	
2½	78·81	26	75·69	173	66·01		39·61
3	78·64	27	75·56	190	65·01		
3½	78·50	28	75·49	206	64·03		
4	78·35	29	75·42	220	63·18		
4½	78·20	30	75·34	225	62·92		
5	78·03	31	75·26	233	62·61		
5½	77·90	33	75·16	243	62·05		
6	77·77	36	74·85	249	61·55		
6½	77·69	40	74·53	274	60·32		
7	77·60	45	74·16	290	59·53		
7½	77·52	47½	74·02	302	58·96		
8	77·47	50	73·91	308	58·59		
8½	77·38	55	73·50	320	58·04		
9	77·27	60	73·22	335	57·49		
9½	77·16	65	72·85	357	56·62		
10	77·07	70	72·47	363	56·41		
11	76·97	75	72·18	376	55·89		
12	76·78	81	71·72	385	55·57		
13	76·63	90	71·04	392	55·43		
14	76·59	95	70·88	400	55·18		
15	76·53	100	70·47	416	54·65		
16	76·47	105	70·16	424	54·43		
17	76·41	110	69·75	435	54·14		
18	76·34	115	69·49	444	53·93		
19	76·25	120	69·03	476	53·35		
20	76·16	124	68·95	508	52·89		
		134					

TABLE 7H

CHROMIUM STEELS
Quenched 1100°C.

CHROME D Quenched 1100°C

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	49·22	21	44·78	250	43·79
½	49·00	22	44·77	269	43·78
1	47·85	23	44·77	280	43·76
1½	47·18	24	44·75	300	43·76
2	46·67	25	44·75	335	43·74
2½	46·21	26	44·72	Completely tempered..	33·23
3	45·82	27	44·72		
3½	45·65	28	44·72		
4	45·53	29	44·70		
4½	45·48	30	44·70		
5	45·42	35	44·61		
5½	45·35	40	44·44		
6	45·30	45	44·36		
6½	45·25	50	44·30		
7	45·22	55	44·25		
7½	45·20	60	44·21		
8	45·18	65	44·16		
8½	45·15	70	44·08		
9	45·11	75	44·05		
9½	45·10	91	44·01		
10	45·08	102	43·97		
11	45·06	124	43·94		
12	45·04	139	43·94		
13	45·03	150	43·93		
14	45·00	163	43·90		
15	44·97	173	43·89		
16	44·93	180	43·89		
17	44·90	195	43·89		
18	44·86	201	43·84		
19	44·83	215	43·82		
20	44·78	231	43·80		

CHROME E Quenched 1100°C.

E 1100°

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	57·32	21	48·27	220	46·55
½	55·15	22	48·23	230	46·54
1	53·21	23	48·19	240	46·52
1½	52·09	24	48·17	250	46·50
2	51·38	25	48·13	260	46·49
2½	50·78	26	48·09	Completely	
3	50·33	27	48·03	Tempered....	
3½	50·01	28	47·99		34·12
4	49·70	29	47·96		
4½	49·59	30	47·92		
5	49·50	35	47·80		
5½	49·42	40	47·72		
6	49·33	45	47·69		
6½	49·27	50	47·54		
7	49·19	55	47·48		
7½	49·13	60	47·43		
8	49·08	70	47·34		
8½	48·95	80	47·25		
9	48·91	90	47·16		
9½	48·86	100	47·08		
10	48·80	110	47·01		
11	48·74	120	46·96		
12	48·67	130	46·85		
13	48·60	140	46·79		
14	48·54	150	46·75		
15	48·48	160	46·73		
16	48·44	170	46·69		
17	44·40	180	46·66		
18	48·37	190	46·62		
19	48·33	200	46·59		
20	48·30	210	46·57		

CHROME F Quenched 1100°C.

F 1100.

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	62·23	21	50·41	410	47·70
½	61·95	22	50·34	Completely Tempered..	35·41
1	61·58	23	50·28		
1½	56·32	24	50·21		
2	54·81	25	50·14		
2½	54·25	26	50·09		
3	53·83	27	50·03		
3½	53·66	28	49·96		
4	53·38	29	49·89		
4½	53·14	30	49·83		
5	52·79	34	49·72		
5½	52·55	40	49·46		
6	52·37	52	49·13		
6½	52·16	66	48·91		
7	52·01	75	48·80		
7½	51·84	86	48·65		
8	51·76	100	48·52		
8½	51·59	115	48·43		
9	51·48	125	48·38		
9½	51·39	140	48·26		
10	51·30	151	48·20		
11	51·21	169	48·14		
12	51·07	180	48·06		
13	50·93	210	47·97		
14	50·81	225	47·95		
15	50·74	236	47·90		
16	50·70	263	47·87		
17	50·64	288	47·83		
18	50·55	330	47·79		
19	50·50	351	47·75		
20	50·44	381	47·74		

CHROME B Quenched 1100°C.

B 1100

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	78·19	21	67·47	234	59·32
½	75·50	22	67·39	245	58·93
1	73·93	23	67·31	260	58·49
1½	72·39	24	67·28	275	58·33
2	71·28	25	67·23	290	57·88
2½	70·52	26	67·17	306	57·43
3	70·03	27	67·09	325	56·83
3½	69·68	28	67·03	350	56·54
4	69·42	29	66·95	375	56·08
4½	69·19	30	66·90	390	55·93
5	69·06	32½	66·85	400	55·82
5½	68·94	35	66·77	432	55·46
6	68·83	40	66·62	449	55·19
6½	68·72	45	66·51	478	54·96
7	68·59	50	66·39	500	54·72
7½	68·51	60	66·12	524	54·50
8	68·44	69	65·87	538	54·40
8½	68·33	75	65·52	560	54·13
9	68·28	85	65·12		
9½	68·23	99	64·48		
10	68·20	110	64·01		
11	68·12	120	63·72		
12	68·03	130	63·11		
13	67·96	140	62·55		
14	67·90	157	62·14		
15	67·85	165	61·75		
16	67·78	174	61·47		
17	67·72	188	60·81		
18	67·63	200	60·48		
19	67·56	216	59·81		
20	67·52	225	59·52		

Completely
Tempered..39·61

TABLE 7J.

NICKEL-CHROMIUM STEELS
Quenched 1000°C.

Nickel-Chrome B Quenched 1000°C.

NCB 1000.

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	54·82	21	50·38	301	48·93
½	52·79	22	50·35	321	48·93
1	52·43	23	50·33	353	48·92
1½	52·03	24	50·30	404	48·90
2	51·69	25	50·28		
2½	51·54	26	50·26		
3	51·43	27	50·24		
3½	51·32	28	50·22		
4	51·26	29	50·19		
4½	51·20	30	50·16		
5	51·14	31	50·14		
5½	51·08	35	50·07		
6	51·04	40	50·01		
6½	51·01	45	49·96		
7	50·96	50	49·91		
7½	50·90	61	49·85		
8	50·85	70	49·78		
8½	50·79	82	49·69		
9	50·75	90	49·60		
9½	50·72	101	49·54		
10	50·69	112	49·43		
11	50·66	124	49·37		
12	50·62	140	49·30		
13	50·59	157	49·26		
14	50·56	185	49·21		
15	50·53	199	49·16		
16	50·51	208	49·13		
17	50·48	225	49·07		
18	50·46	240	49·02		
19	50·43	263	48·96		
20	50·41	280	48·94		

Completely

Tempered...

NICKEL-CHROME G1 Quenched 1000°C. NCG1 1000.

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	74·91	21	67·95	396	61·23
½	71·53	22	67·92	442	60·89
1	69·94	23	67·90	507	60·48
1½	69·48	24	67·87	555	60·19
2	69·22	25	67·85	601	59·87
2½	69·13	26	67·82	626	59·69
3	69·01	27	67·79		
3½	68·92	28	67·74		
4	68·81	29	67·70		
4½	68·75	30	67·65		
5	68·70	35	67·45		
5½	68·67	40	67·25		
6	68·63	45	67·19		
6½	68·60	50	66·93		
7	68·55	55	66·75		
7½	68·51	65	66·29		
8	68·48	73	66·04		
8½	68·46	90	65·72		
9	68·43	102	65·49		
9½	68·39	122	65·15		
10	68·36	140	64·89		
11	68·33	171	64·21		
12	68·29	195	63·75		
13	68·26	218	63·45		
14	68·23	245	63·02		
15	68·20	260	62·83		
16	68·17	275	62·54		
17	68·14	304	62·26		
18	68·10	328	61·91		
19	68·04	350	61·69		
20	68·00	375	61·51		

NICKEL-CHROME G2 Quenched 1000°C.

G2 1000.

Time (mins)	Res. Microhms	Time	Res.
0	90·52	21	92·08
½	91·46	22	92·08
1	91·83	23	92·10
1½	92·01	24	92·10
2	92·05	25	92·07
2½	92·08	27½	92·05
3	92·12	30	92·05
3½	92·12	35	92·07
4	92·12	40	92·05
4½	92·12	45	92·05
5	92·12	50	92·07
5½	92·12	60	92·04
6	92·12	65	92·02
6½	92·12	70	92·03
7	92·12	75	92·03
7½	92·12	100	92·02
8	92·12	120	92·02
8½	92·12	125	92·02
9	92·12	146	92·02
9½	92·12	171	91·98
10	92·12	200	91·99
11	92·12	222	91·99
12	92·12	250	91·99
13	92·12	266	91·98
14	92·12	287	91·97
15	92·12	301	91·97
16	92·08	359	91·96
17	92·09		
18	92·08		
19	92·08		
20	92·10		

TABLE 7K

NICKEL-CHROMIUM STEELS
Quenched 1100°C.

NICKEL-CHROME B Quenched 1100°C.

NCrB 1100.

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	54·89	21	50·91	378	49·75
½	54·14	22	50·87		
1	53·53	23	50·84		
1½	53·08	24	50·81		
2	52·82	25	50·78		
2½	52·52	26	50·74		
3	52·35	27	50·72		
3½	52·18	28	50·69		
4	52·01	29	50·67		
4½	51·82	30	50·65		
5	51·76	35	50·53		
5½	51·67	40	50·45		
6	51·58	45	50·38		
6½	51·54	50	50·30		
7	51·50	55	50·25		
7½	51·47	60	50·20		
8	51·40	65	50·20		
8½	51·35	70	50·18		
9	51·33	80	50·14		
9½	51·30	85	50·06		
10	51·25	90	50·04		
11	51·22	100	50·00		
12	51·19	125	49·95		
13	51·15	150	49·90		
14	51·13	175	49·88		
15	51·11	201	49·86		
16	51·09	222	49·83		
17	51·04	249	49·80		
18	51·01	270	49·80		
19	50·97	301	49·78		
20	50·95	352	49·77		

NICKEL-CHROME G1 Quenched 1100°C.

NCG1 11CC

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	76·37	21	68·58	226	63·80
½	73·19	22	68·53	253	63·53
1	71·81	23	68·50	260	63·41
1½	71·04	24	68·45	281	62·97
2	70·47	25	68·41	320	62·54
2½	70·10	26	68·37	350	62·35
3	69·81	27	68·34	378	62·01
3½	69·62	28	68·32	400	61·73
4	69·51	29	68·28	427	61·45
4½	69·45	30	68·25	441	61·34
5	69·41	35	68·12	465	61·13
5½	69·33	40	68·00	488	60·92
6	69·27	45	67·94		
7½	69·23	50	67·85		
8	69·18	55	67·67		
8½	69·15	60	67·52		
9	69·11	65	67·38		
9½	69·07	70	67·19		
10	69·03	75	67·03		
11	69·00	80	66·96		
12	68·97	85	66·85		
13	68·95	95	66·59		
14	68·90	100	66·48		
15	68·85	110	66·23		
16	68·80	120	65·98		
17	68·76	130	65·75		
18	68·71	150	65·45		
19	68·67	170	64·88		
20	68·62	200	64·25		

NICKEL-CHROME G2 Quenched 1100°C.

G3 1100.

Time (mins)	Res. Microhms	Time	Res.
0	88·53	35	90·06
½	89·56	40	90·04
1	90·11	45	90·06
1½	90·19	55	90·06
2	90·19	60	90·04
2½	90·18	70	90·04
3	90·16	90	90·02
3½	90·17	100	90·00
4	90·16	110	90·00
4½	90·17	125	89·97
5	90·16	150	89·99
5½	90·17	160	89·93
6	90·15	185	89·93
6½	90·15	200	89·90
7	90·15	250	89·85
7½	90·13	275	89·85
8	90·15	300	89·82
8½	90·15	325	89·79
9	90·14	350	89·77
9½	90·14	400	89·73
10	90·13	425	89·70
11	90·13	450	89·67
14	90·12	475	89·64
16	90·12		
18	90·10		
20	90·10		
25	90·06		
30	90·06		

TABLE 7L.

INTERRUPTED TEMPERING CURVES.

Chromium B
Interrupted tempering.

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	83·55	21	76·33	327	58·71
½	81·05	22	76·25	328	58·51
1	80·22	23	76·16	329	58·31
1½	79·74	24	76·05	330	58·22
2	79·36	25	76·00	331	58·14
2½	79·04	30	75·53	332	58·07
3	78·78	35	75·11	333	58·01
3½	78·56	40	74·79	334	57·95
4	78·39	45	74·51	335	57·88
4½	78·25	50	74·23	340	57·57
5	78·06	55	73·85	345	57·33
5½	77·94	60	73·40	350	57·18
6	77·81	70	72·65	355	57·00
6½	77·63	80	71·94	360	56·81
7	77·63	90	71·18	365	56·69
7½	77·52	100	70·53	370	56·55
8	77·46	110	69·77	375	56·47
8½	77·38	130	68·43	390	55·94
9	77·34	150	67·29	400	55·66
9½	77·30	170	66·02	425	55·22
10	77·25	200	64·12	450	54·81
11	77·20	225	62·23	475	54·54
12	77·13	250	61·14	500	54·49
13	77·05	275	60·13	525	54·37
14	76·93	300	59·22	550	54·22
15	76·81	320	58·68	575	54·11
16	76·75	Withdrawn and Replaced.		600	54·00
17	76·68	325	59·63	625	53·91
18	76·59	325½	59·25		
19	76·50	326	59·08		
20	76·43	326½	58·83		

Nickel-chromium G1. Quenched 1100°C.
Interrupted Tempering

Time (mins)	Res. Microhms	Time	Res.	Time	Res.
0	76·53	70	67·62	300	62·01
½	74·45	80	67·39	325	61·83
1	72·79	90	67·22	350	61·58
1½	72·36	100	67·03	375	61·31
2	70·74	126	66·51	400	61·13
2½	70·37	150	65·93	425	60·91
3	70·16	175	65·45	450	60·61
3½	70·04	200	64·84	Withdrawn and Replaced	
4	69·93	225	64·42		
5	69·78	Withdrawn and Replaced.		455	61·54
6	69·69			456	61·24
7	69·54	230	65·90	457	61·09
8	69·49	231	65·54	458	60·76
9	69·37	232	65·27	459	60·65
10	69·28	233	64·99	460	60·52
11	69·18	234	64·76	465	60·28
12	69·11	235	64·51	470	60·01
13	69·06	236	64·29	475	59·89
14	69·03	237	64·15	480	59·69
15	68·95	238	64·02	485	59·55
17	68·84	239	63·89	490	59·45
20	68·72	240	63·74	495	59·36
25	68·63	245	63·23	500	59·28
30	68·52	250	62·91	505	59·13
35	68·39	255	62·78	555	58·72
40	68·28	260	62·67	605	58·31
45	68·20	265	62·54	655	57·96
50	68·11	270	62·43		
55	67·80	275	62·31		
60	67·71	280	62·24		

OTHER
TABLES
in
VOLUME
I

SOME PROPERTIES of STEEL

VOL. III.

FIGURES

M. S. FISHER, B.Sc., A.R.T.C.

FIGURES 1 to 7A.

ELECTRODE POTENTIAL

FIGURE 1.

ELECTRODE POTENTIAL of QUENCHED CARBON STEELS.

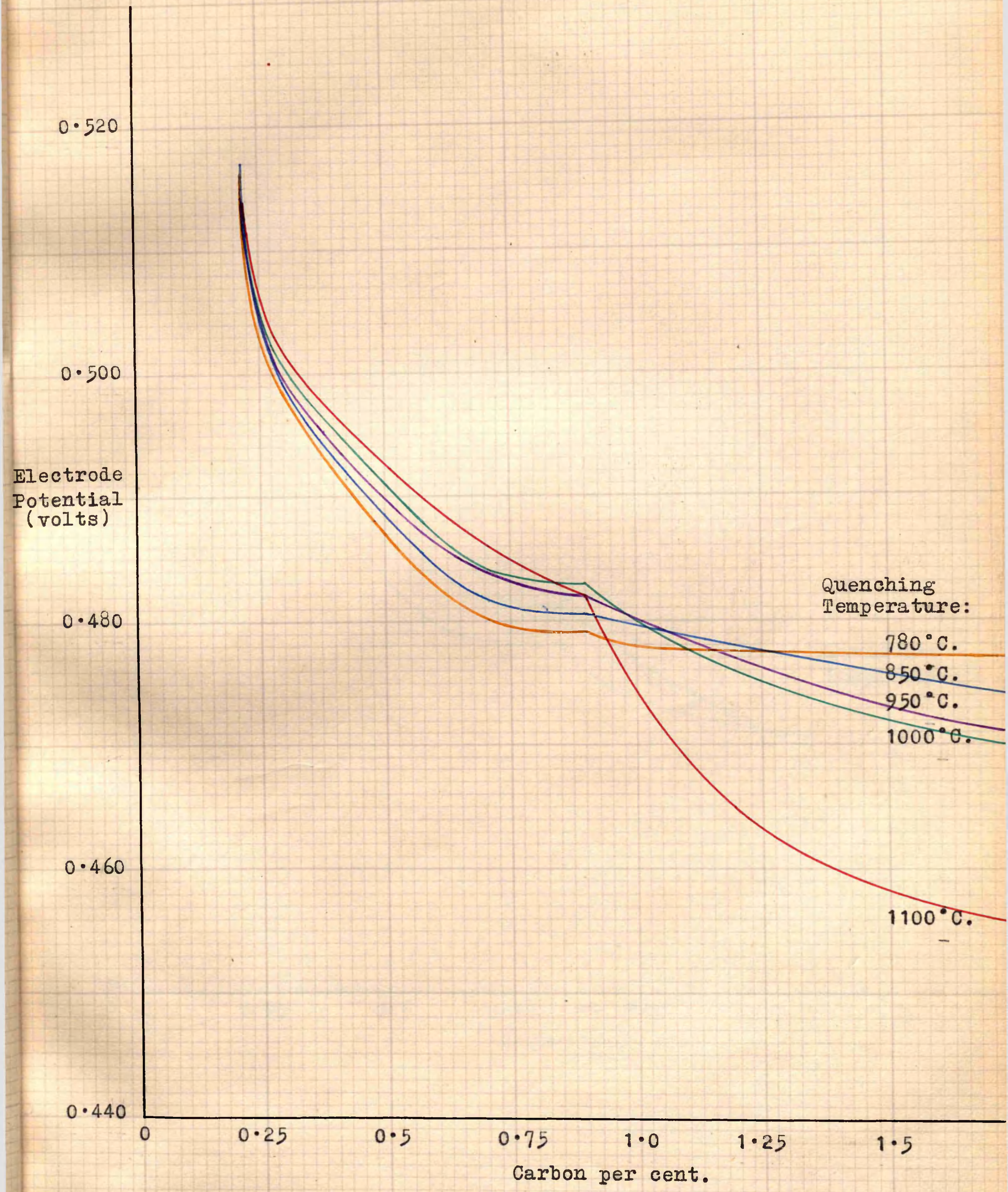


FIGURE 2.

ELECTRODE POTENTIAL of QUENCHED CHROMIUM-STEELS.

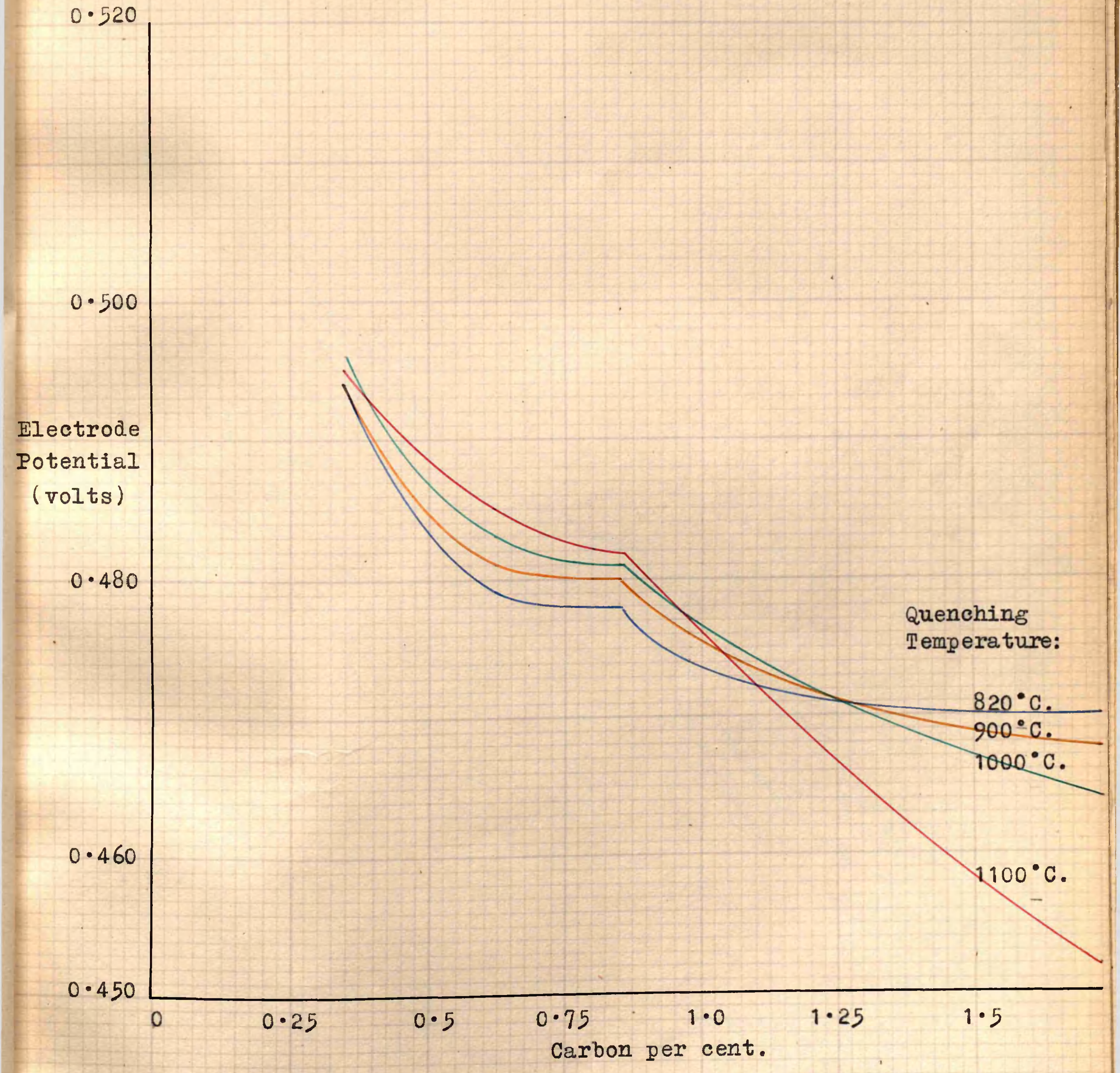


FIGURE 3.

VARIATION of ELECTRODE POTENTIAL
with TIME.

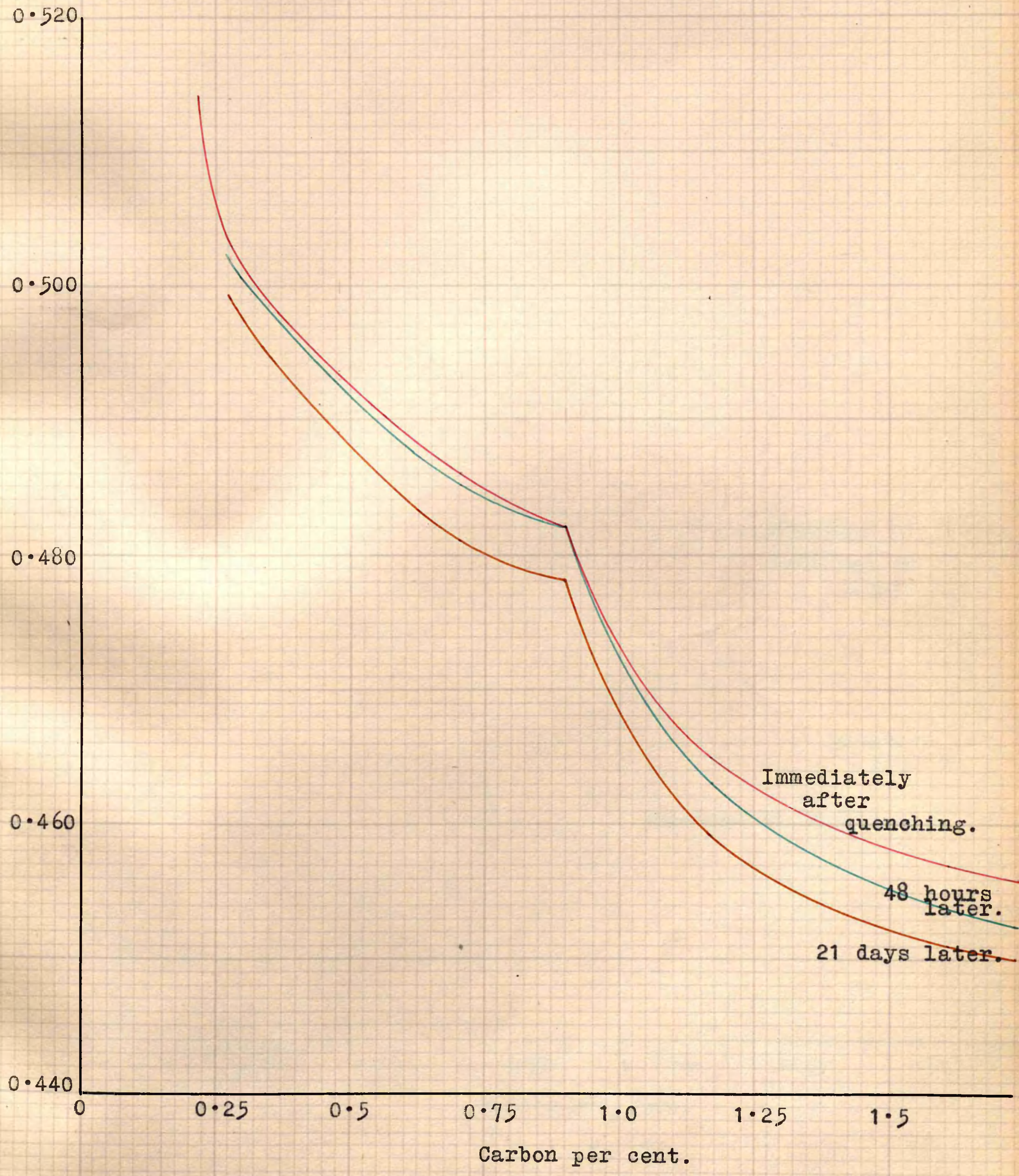


FIGURE 4.

ELECTRODE POTENTIAL

Effect of Immersion in Liquid-Air.

CARBON STEELS.

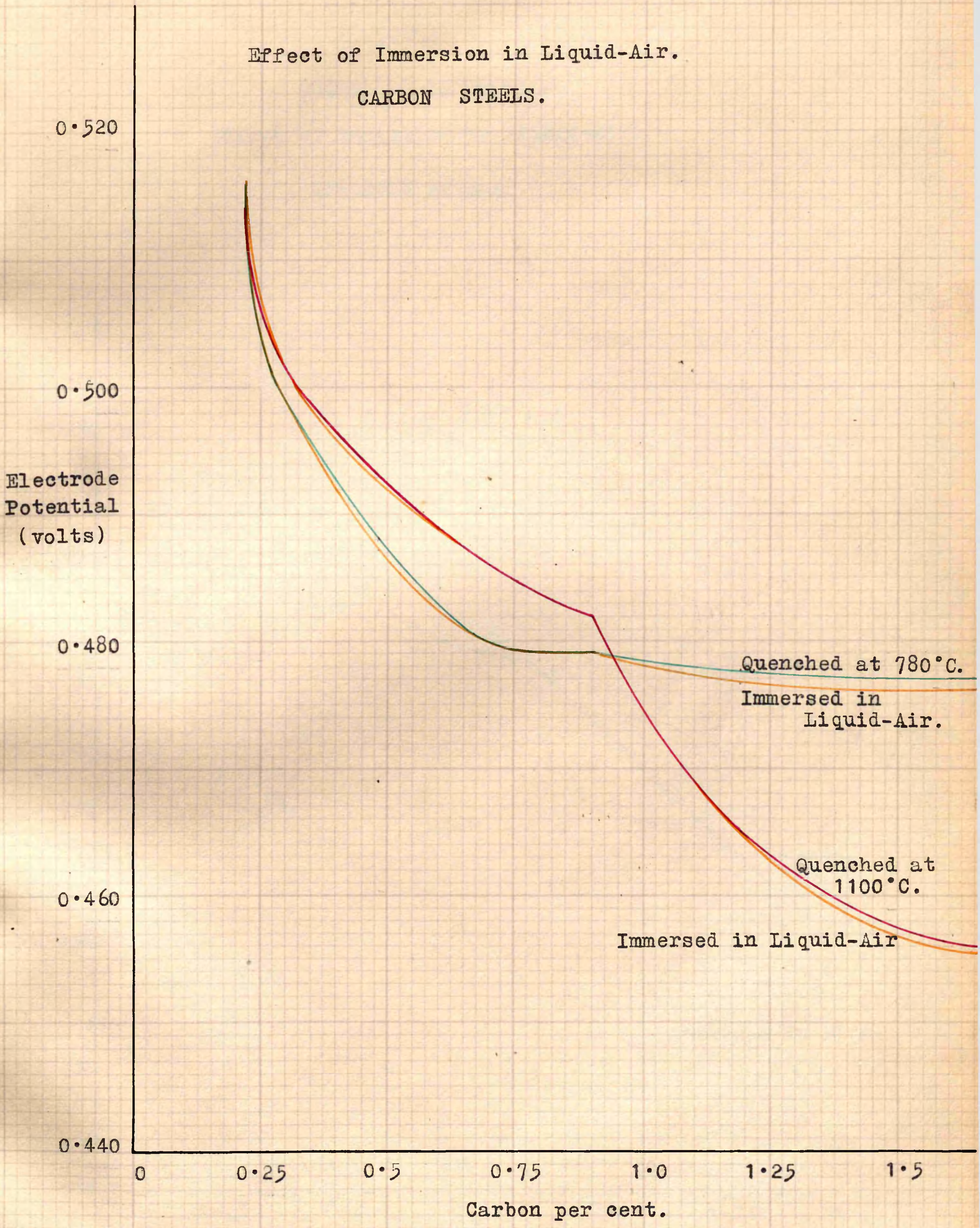


FIGURE 5.

ELECTRODE POTENTIAL

Effect of Immersion in Liquid-Air.

CHROMIUM STEELS.

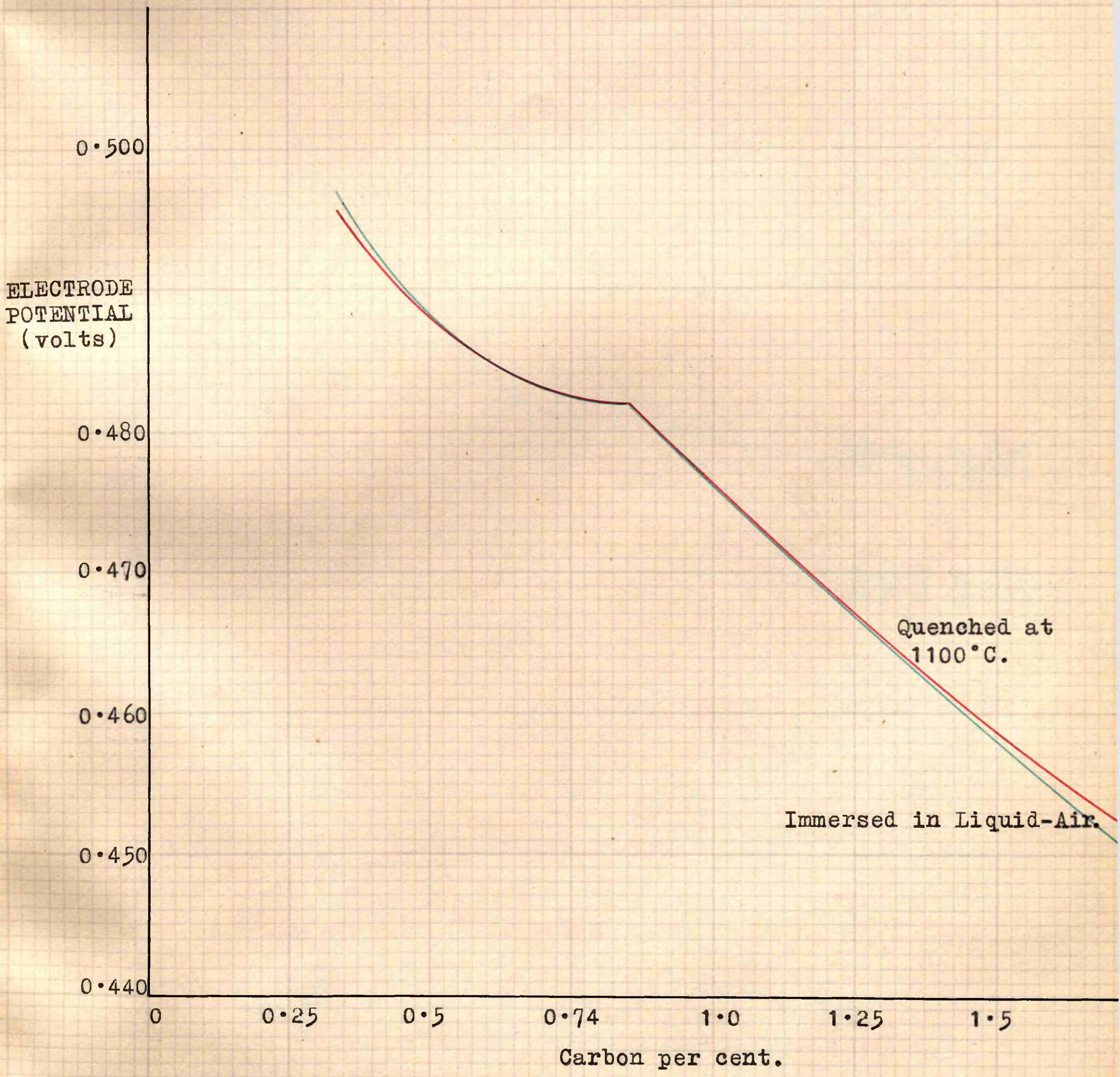


FIGURE 6.

ELECTRODE POTENTIAL

Effect of Varying
the Method of Quenching.

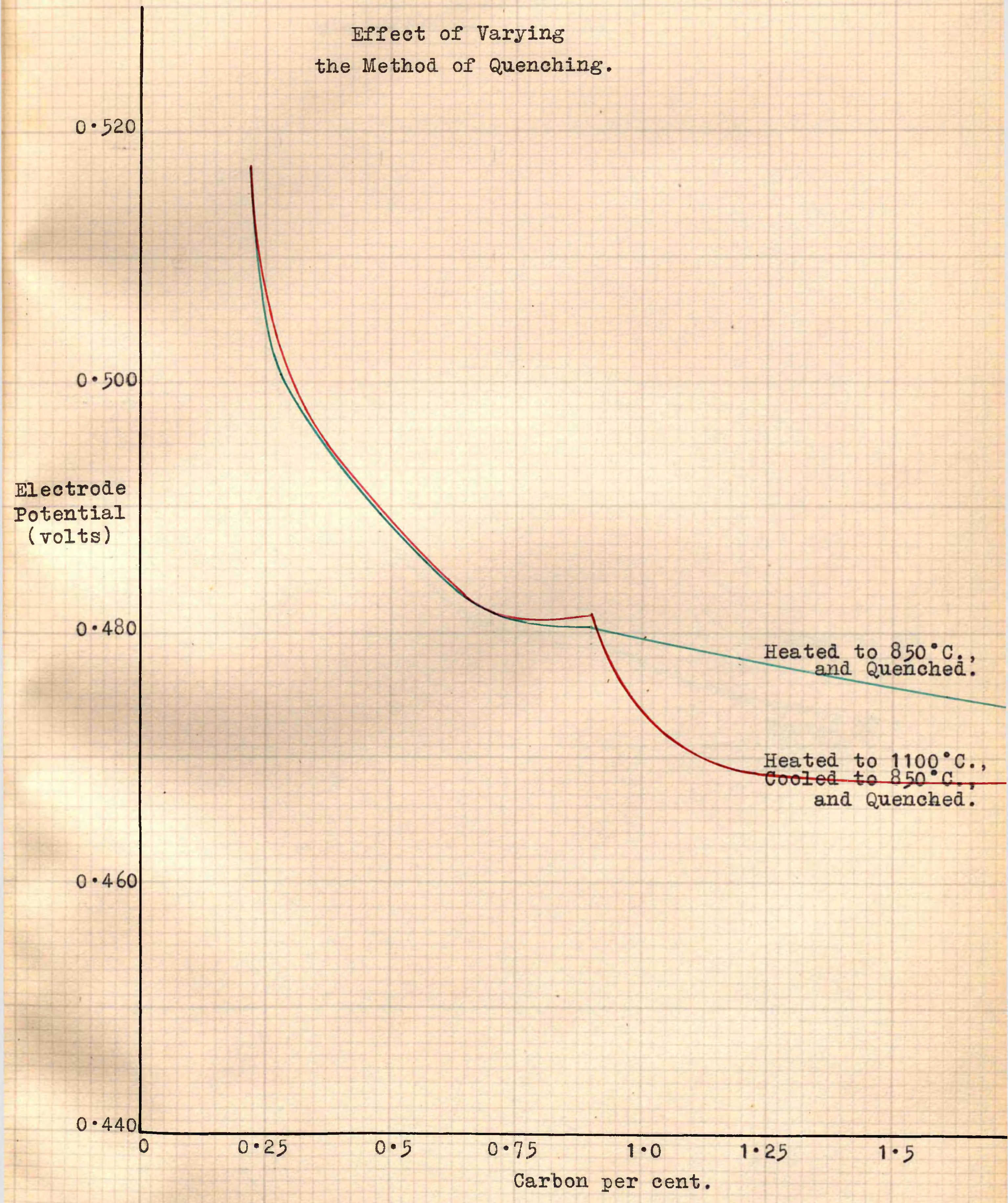


FIGURE 7.

ELECTRODE POTENTIAL

Effect of Varying
the Method of Quenching.

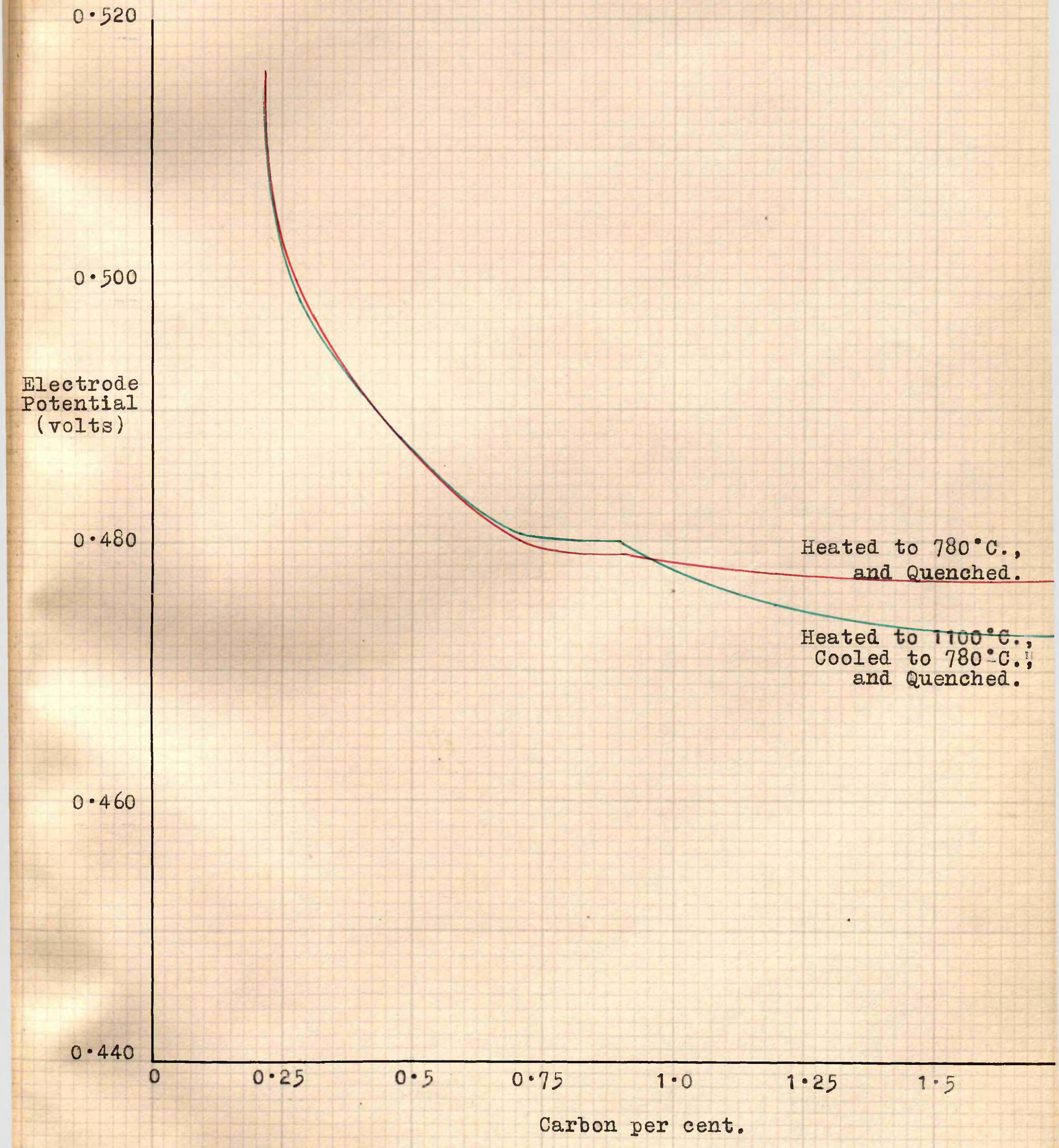
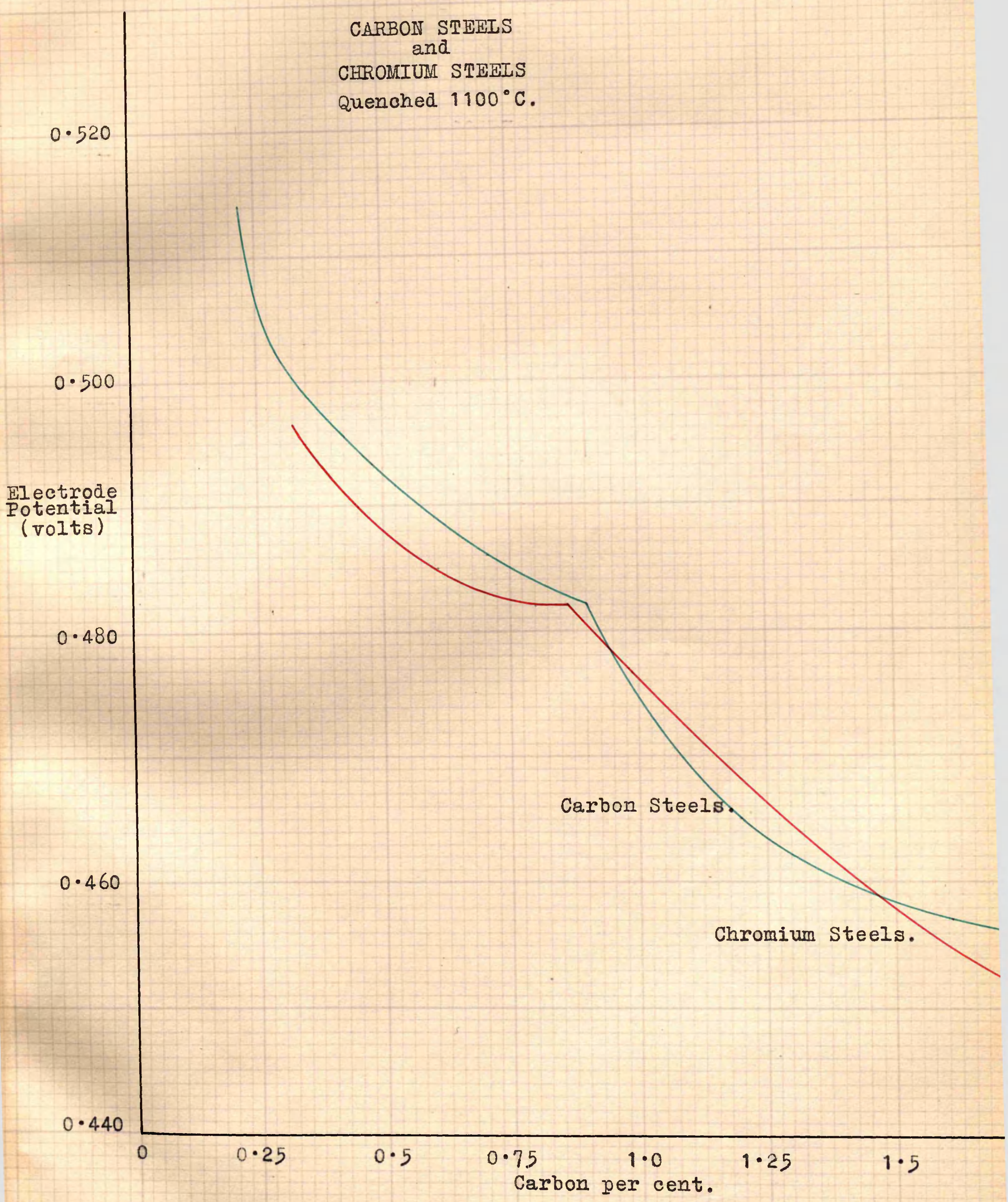


FIGURE 7A

ELECTRODE POTENTIAL

CARBON STEELS
and
CHROMIUM STEELS
Quenched 1100°C.



FIGURES 8 to 14

GRAPHITISATION

FIGURE 8

GRAPHITISATION

Experiments 1 to 7.

For Description of Experiments,
See Vol. I, Pages 45 and 46.

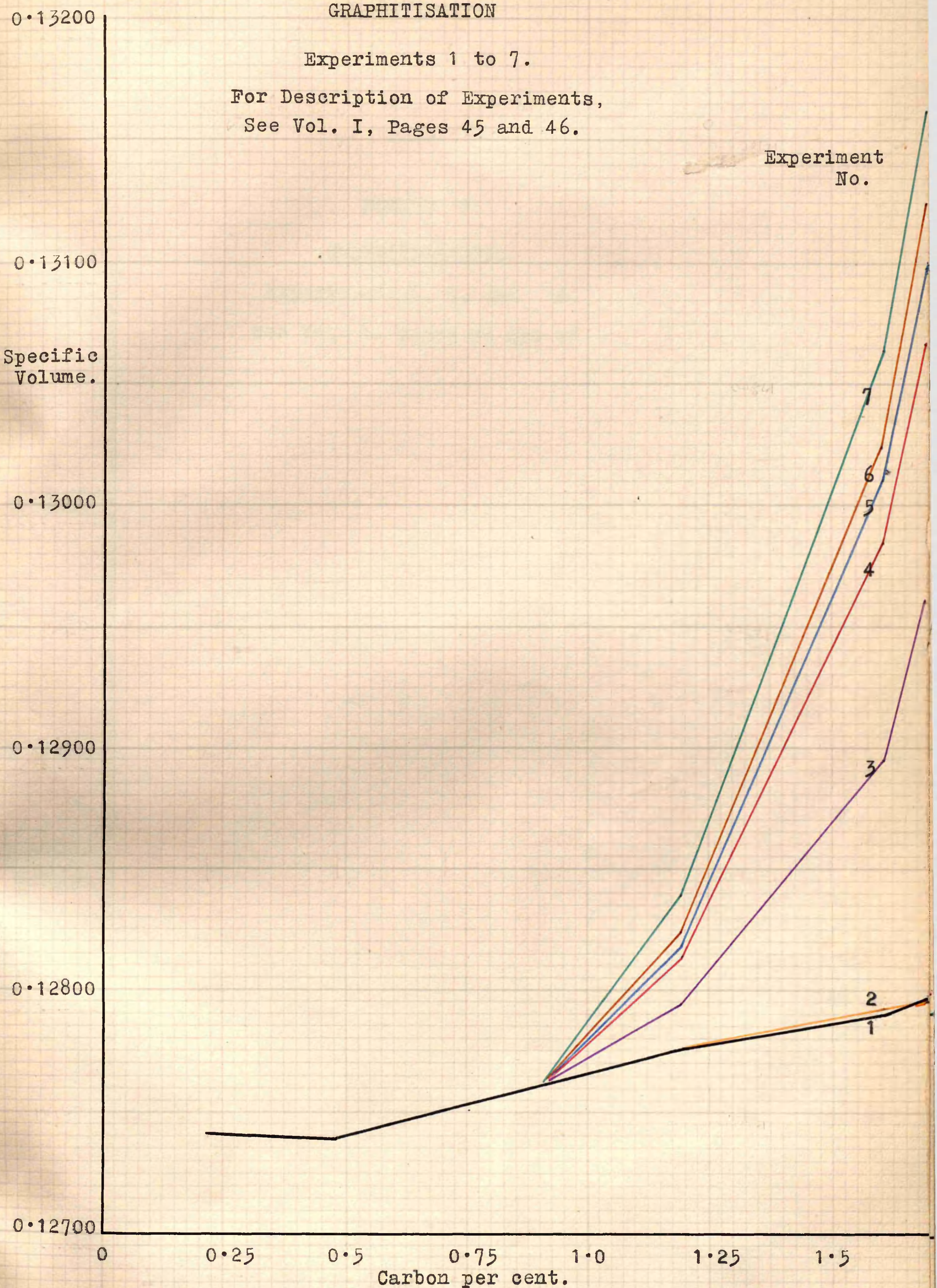


FIGURE 9

GRAPHITISATION.

Experiments 8, 9, and 10.

See Vol. I, Pages 46 and 47.

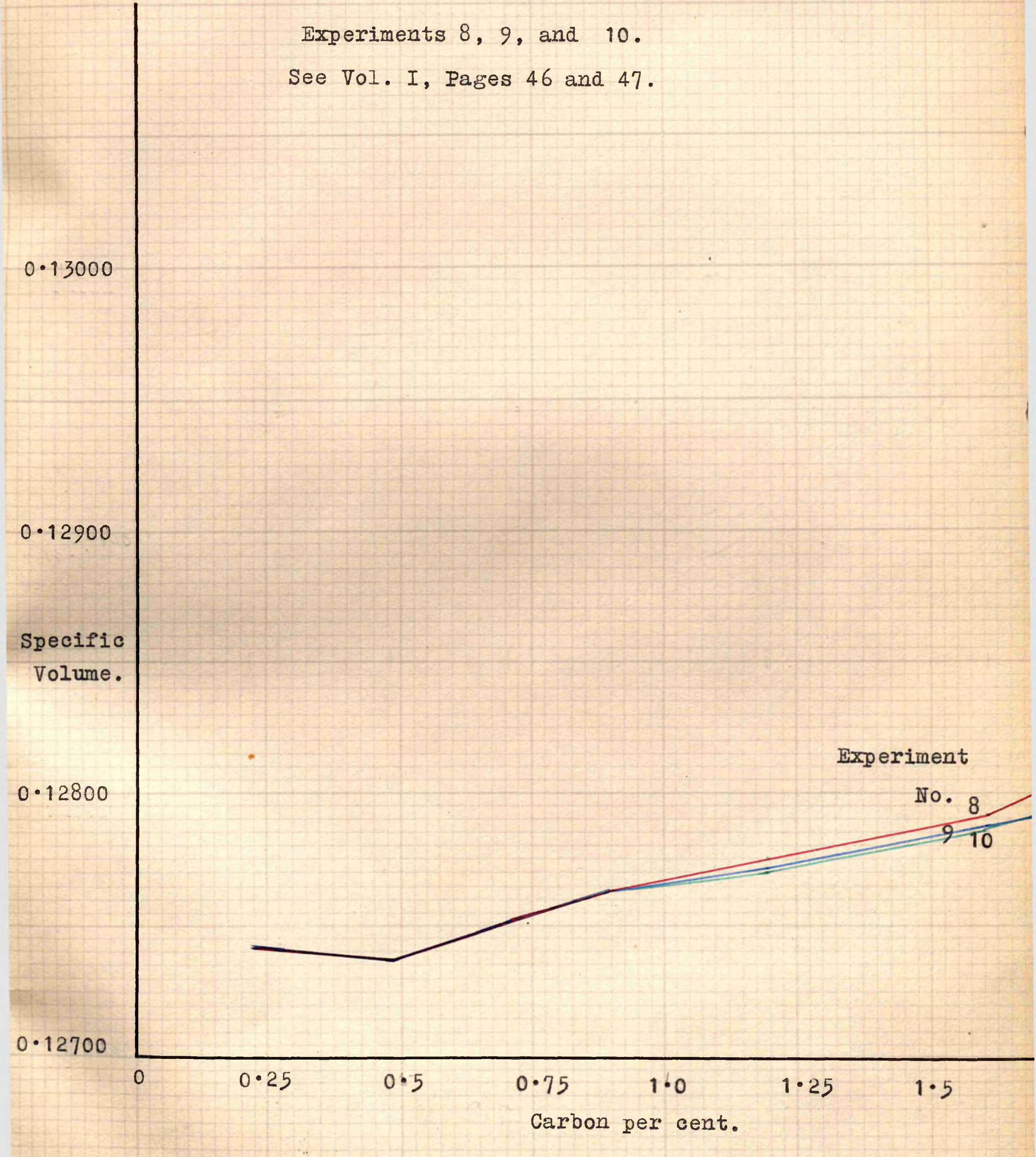


FIGURE 10
GRAPHITISATION

Experiments 11, 12, and 3.

See Vol. I, Pages 47 and 45.

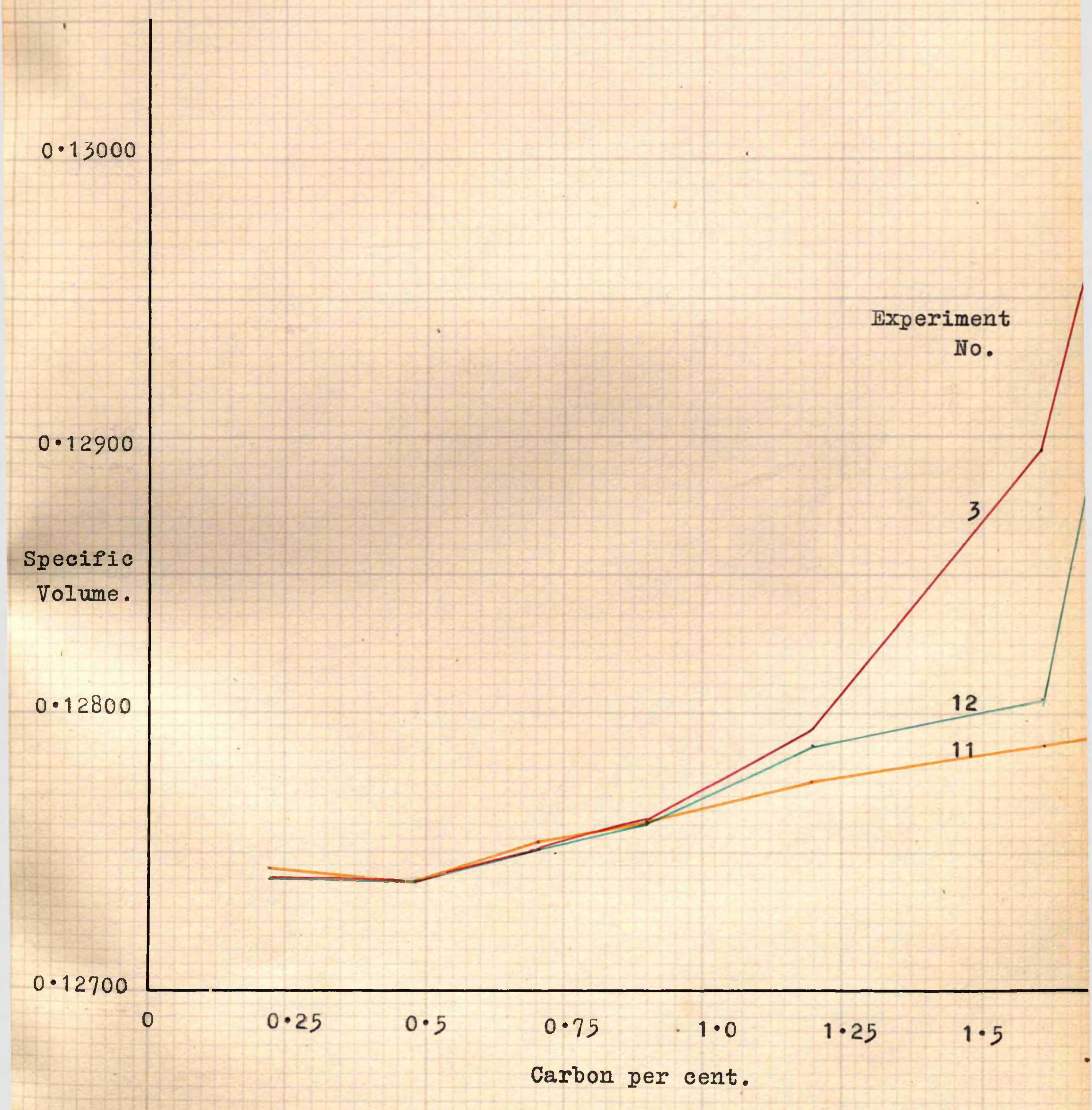


FIGURE 11

GRAPHITISATION

Experiment 13,

Experiments 12, 14, and 3.

See Vol. I, Pages 45, 47, 48, and 49.

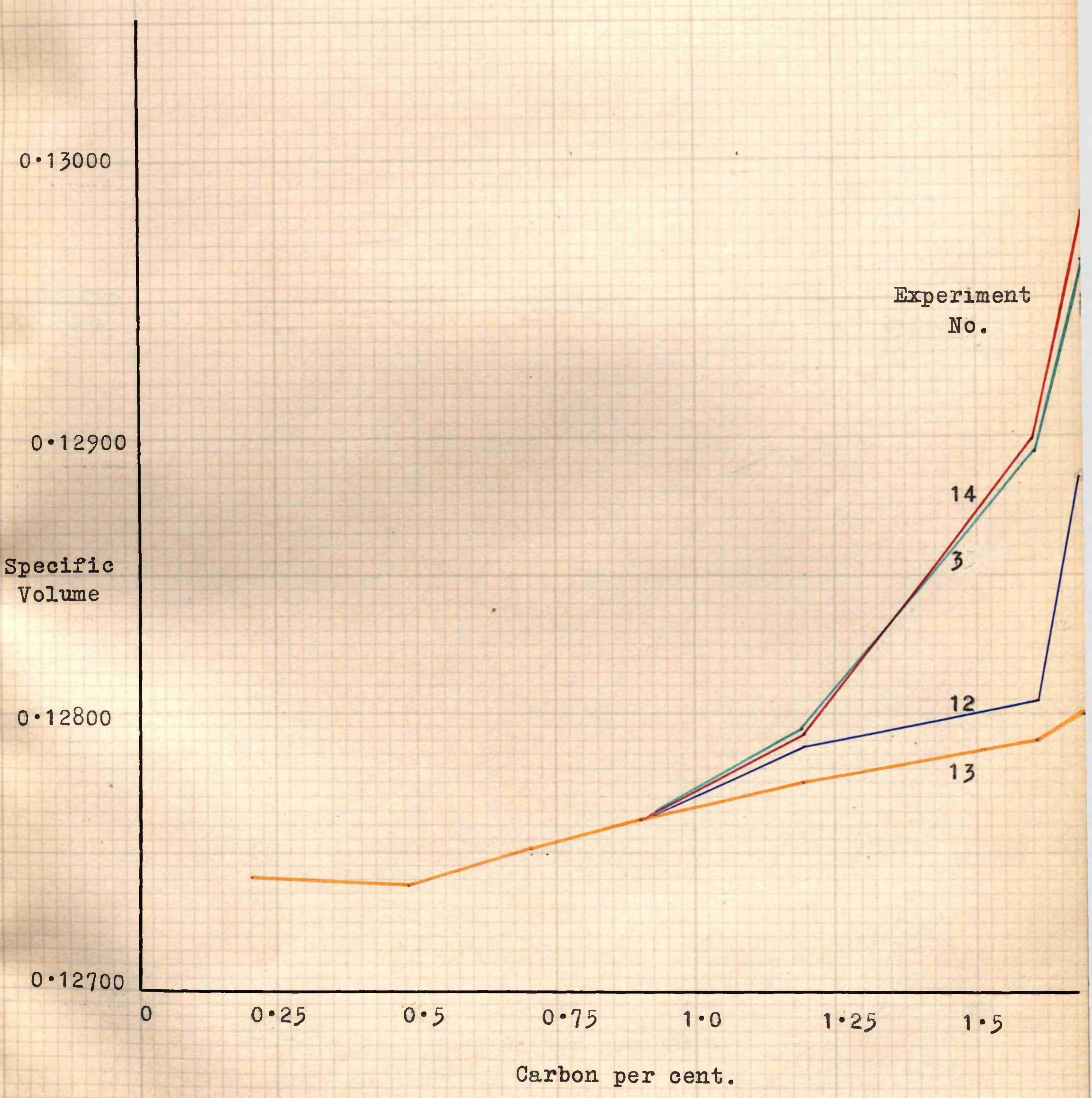


FIGURE 12

GRAPHITISATION

Experiments 14 and 15.

See Vol. I, Page 49.

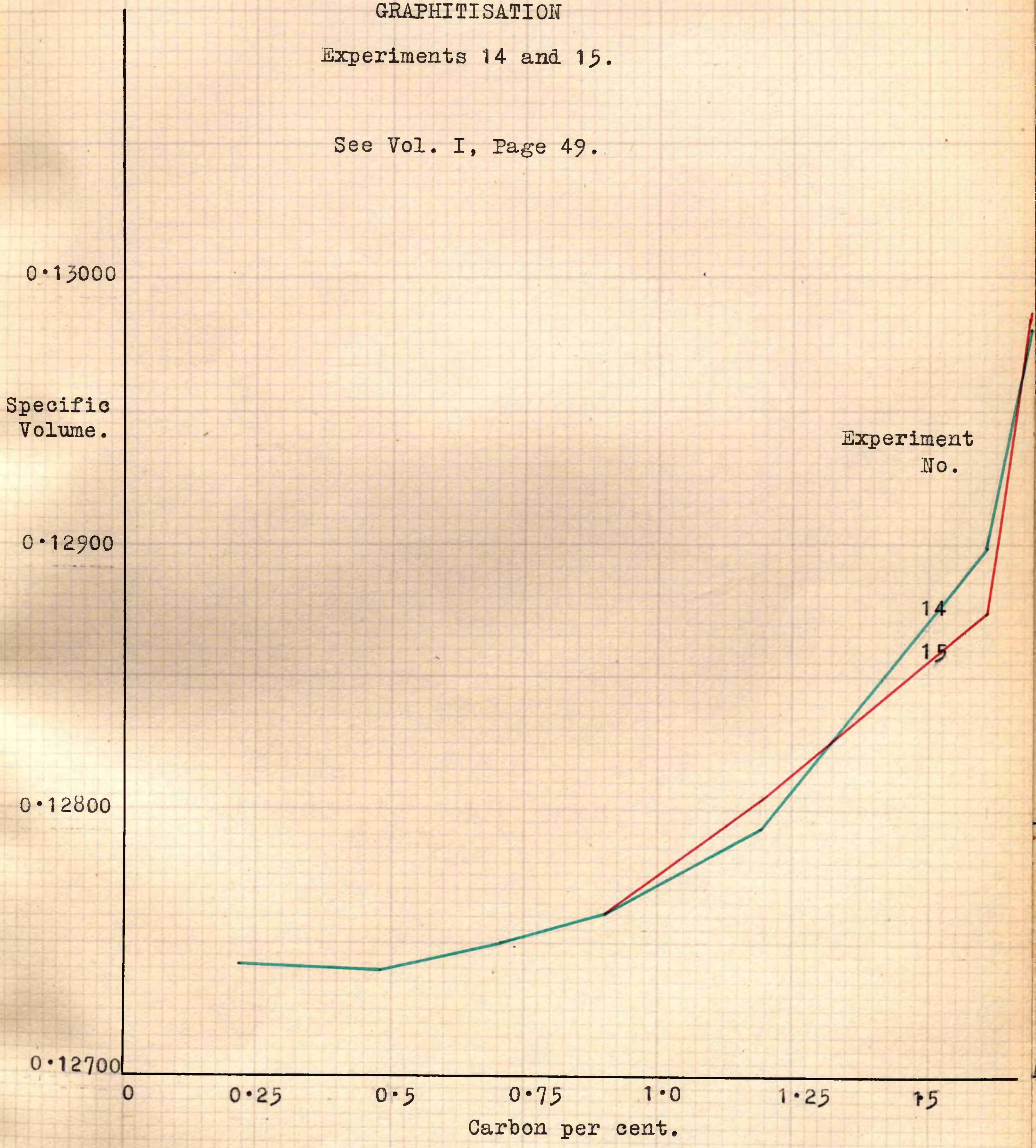


FIGURE 13.

GRAPHITISATION

Experiments 14, 16, and 18:
Experiments 17 and 19.
See Pages 49, and 50,
Vol. I.

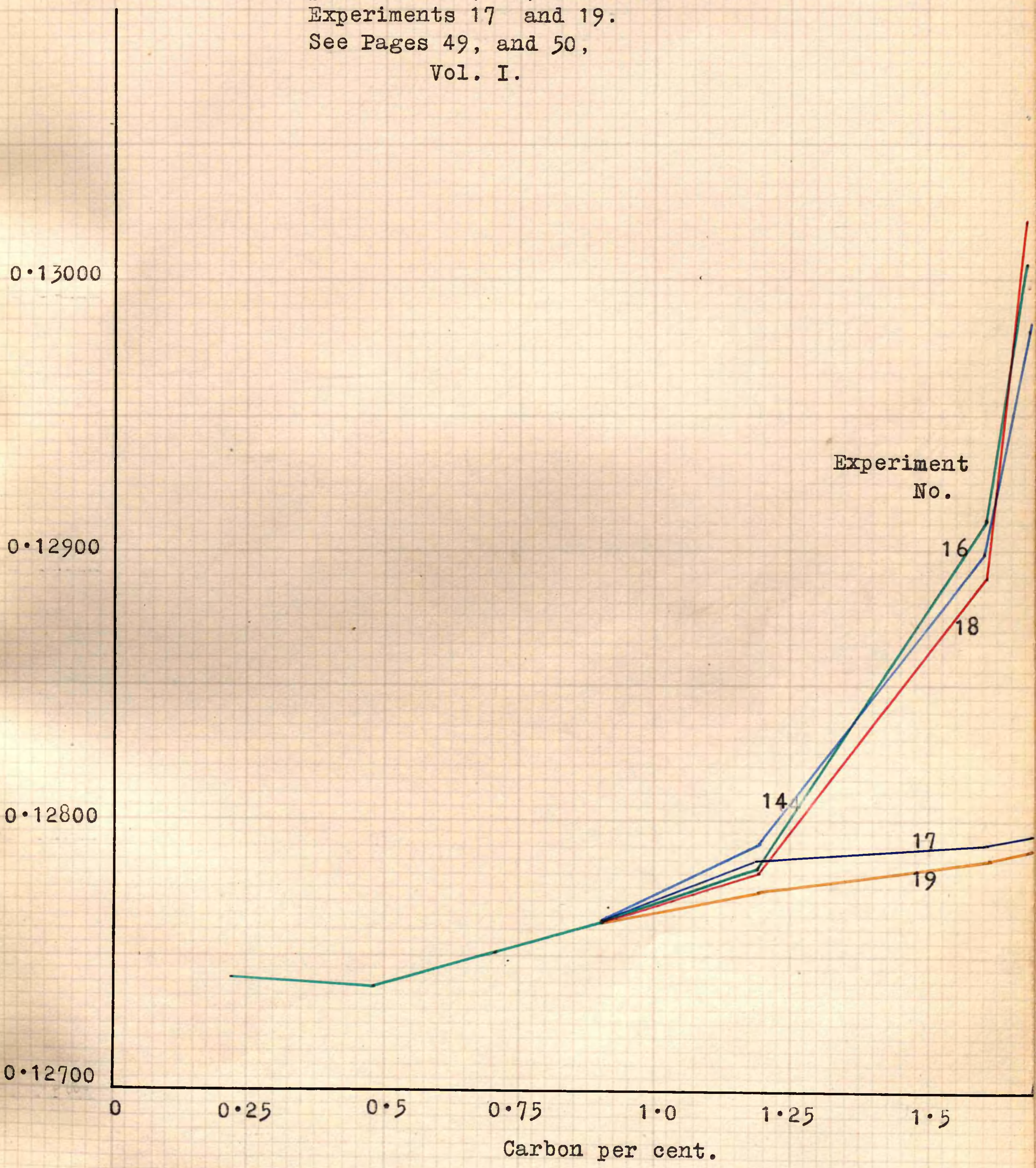
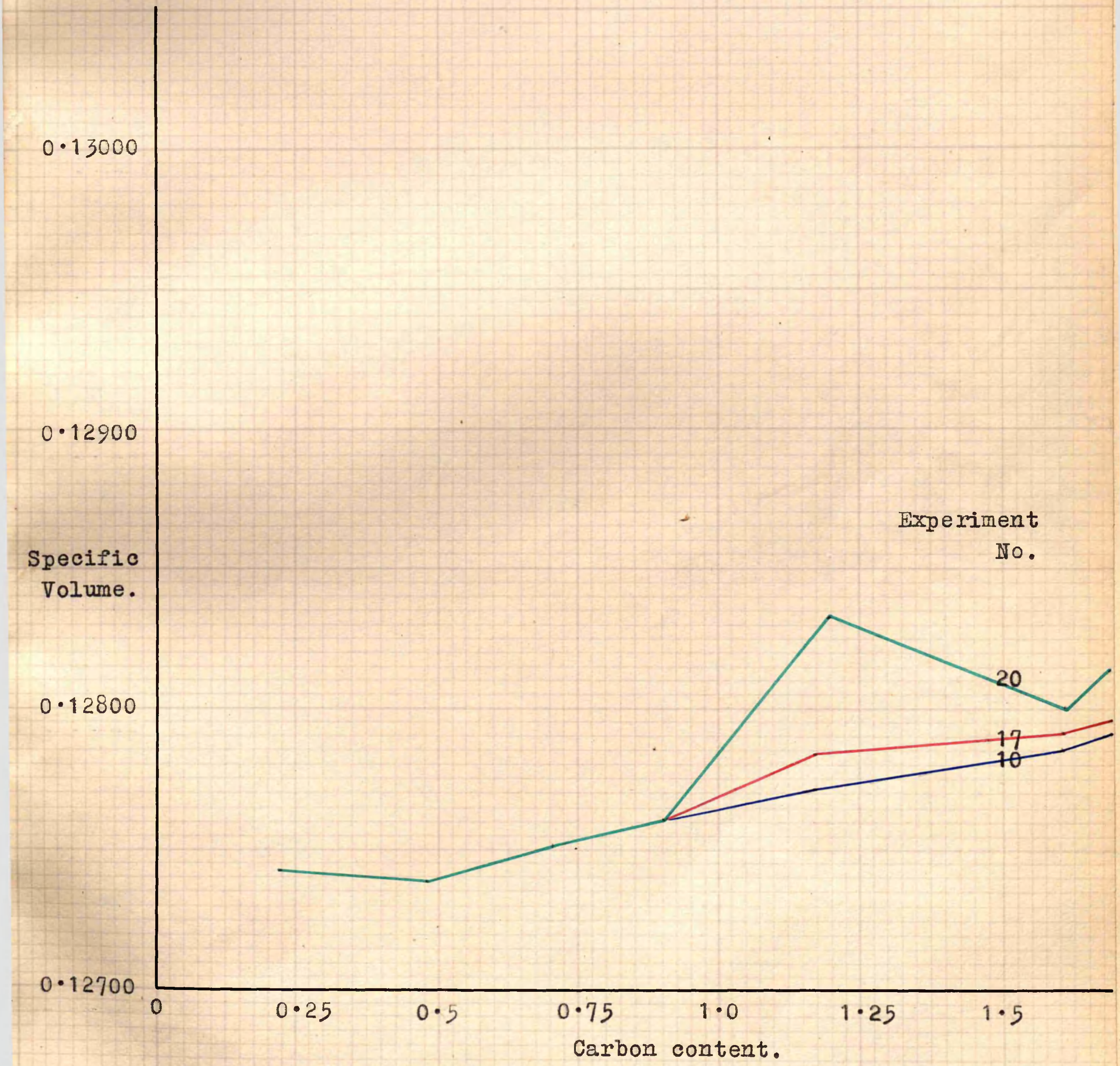


FIGURE 14.

GRAPHITISATION.

Experiments 10, 17, and 20.

See Vol. I; Pages 47, 50, and 51.



FIGURES 15 to 21

ELECTRICAL RESISTANCE

FIGURE 15

APPARATUS USED for

RESISTANCE MEASUREMENTS.

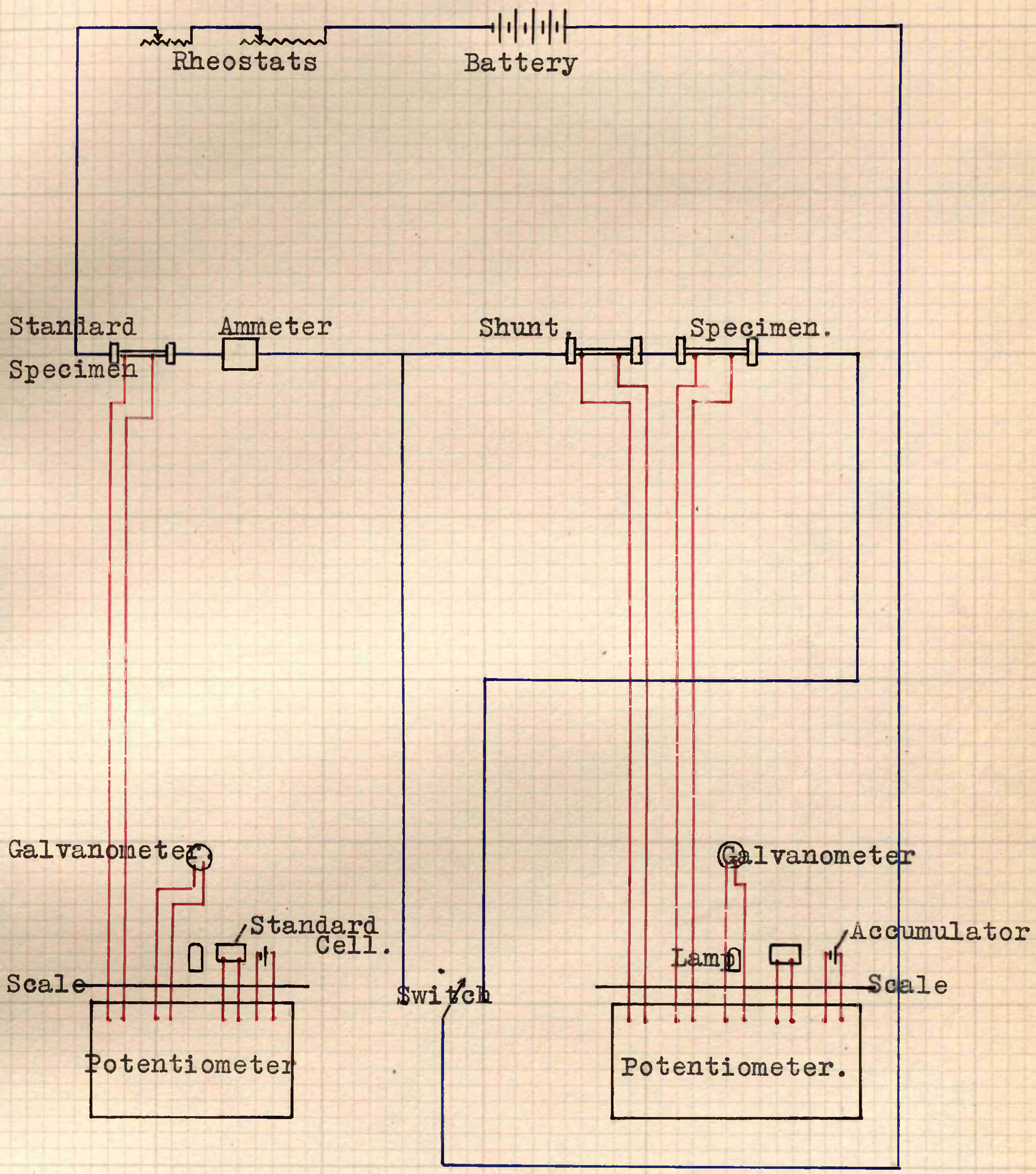


FIGURE 16

ELECTRICAL RESISTANCE of CARBON STEELS, QUENCHED, TEMPERED, and ANNEALED.

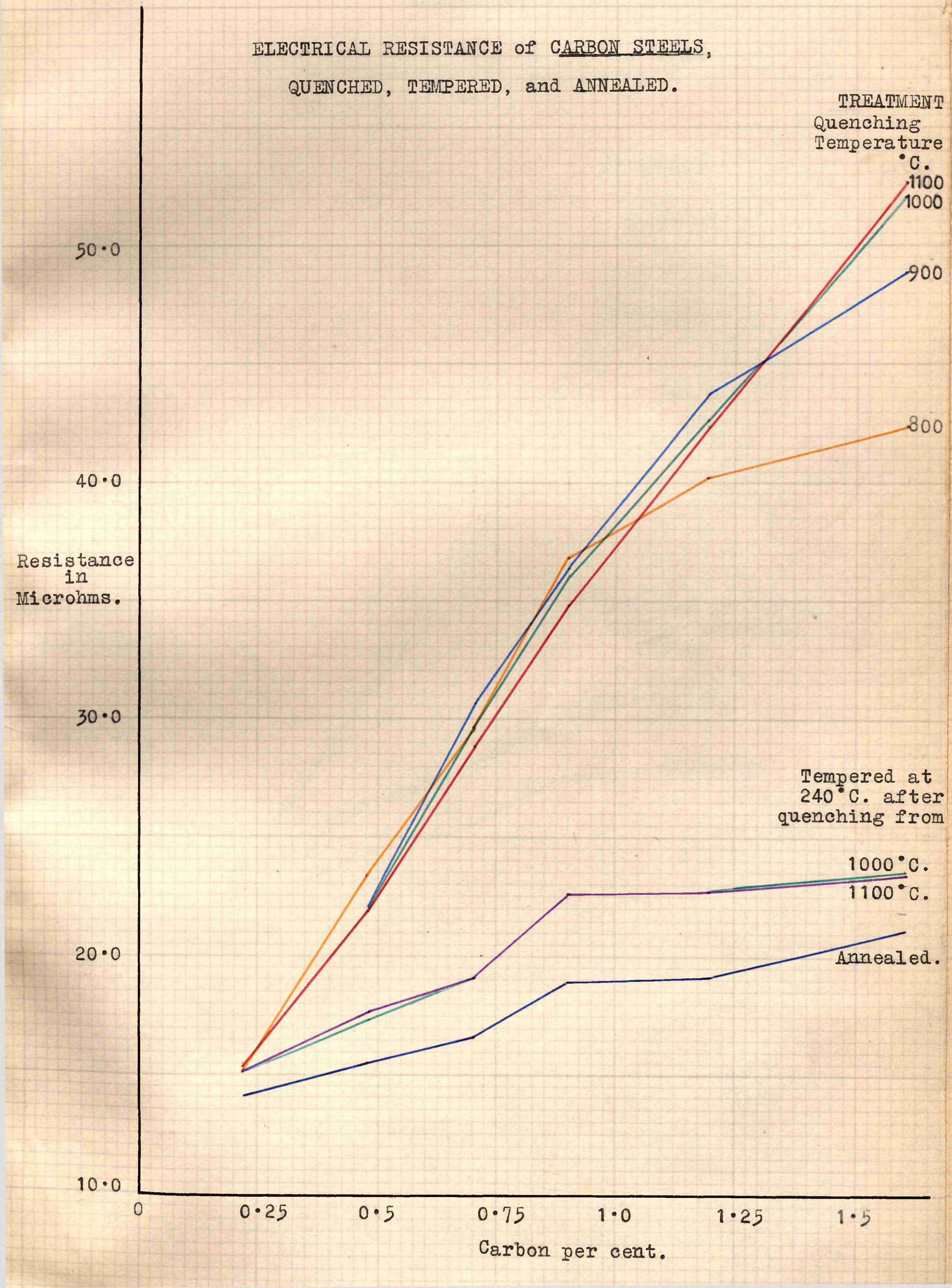


FIGURE 17

ELECTRICAL RESISTANCE of NICKEL STEELS
QUENCHED, TEMPERED, and ANNEALED.

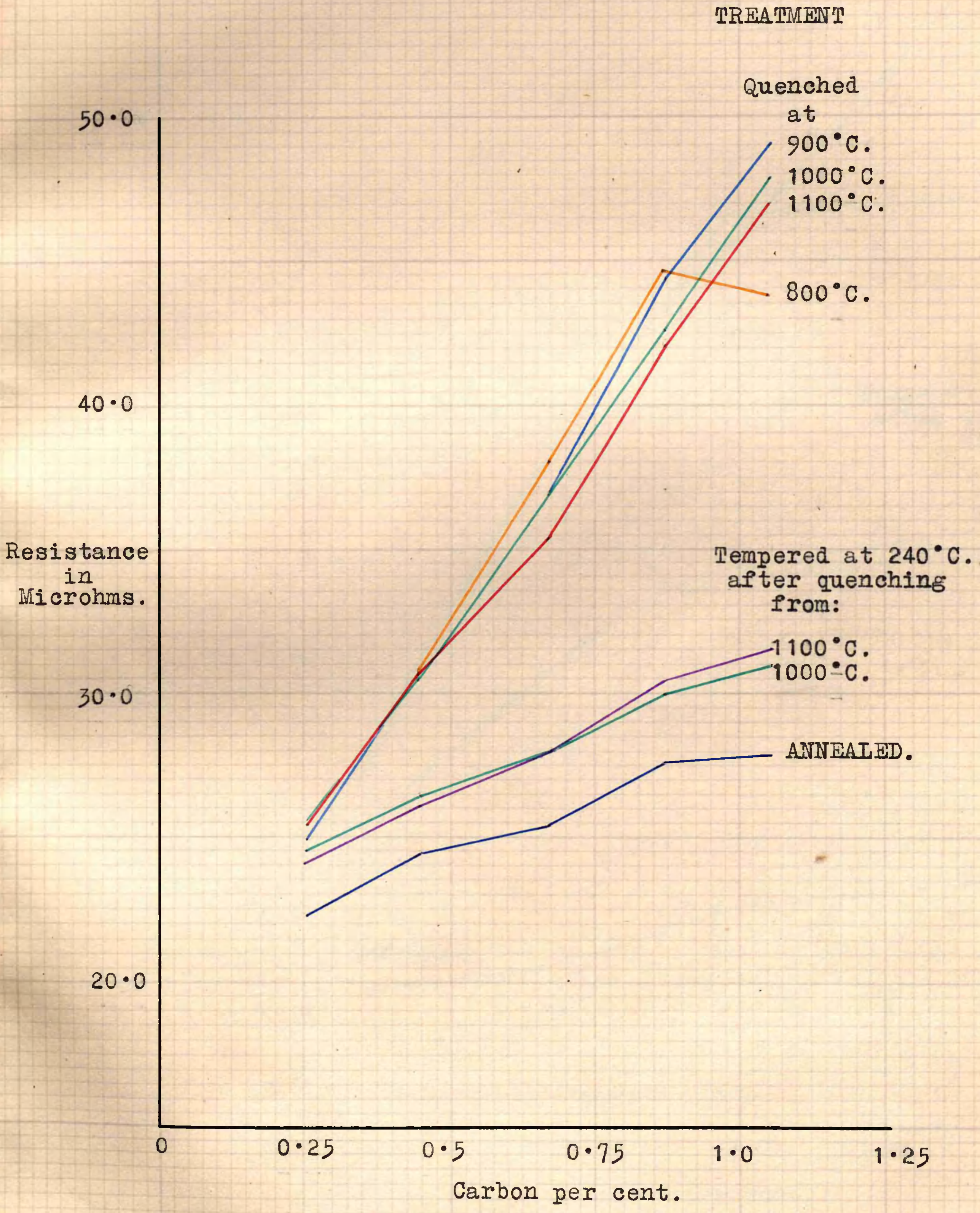
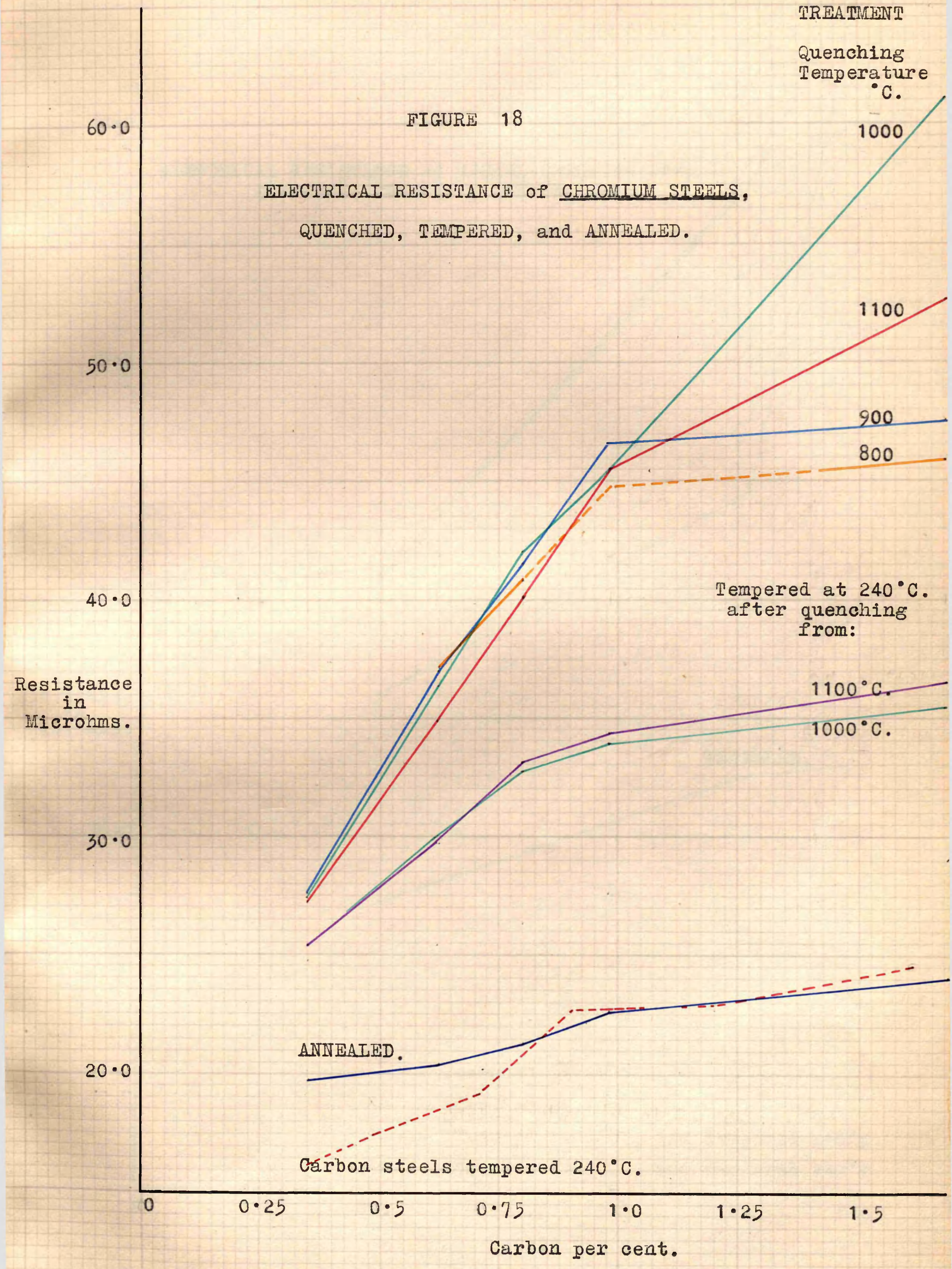


FIGURE 18

ELECTRICAL RESISTANCE of CHROMIUM STEELS, QUENCHED, TEMPERED, and ANNEALED.



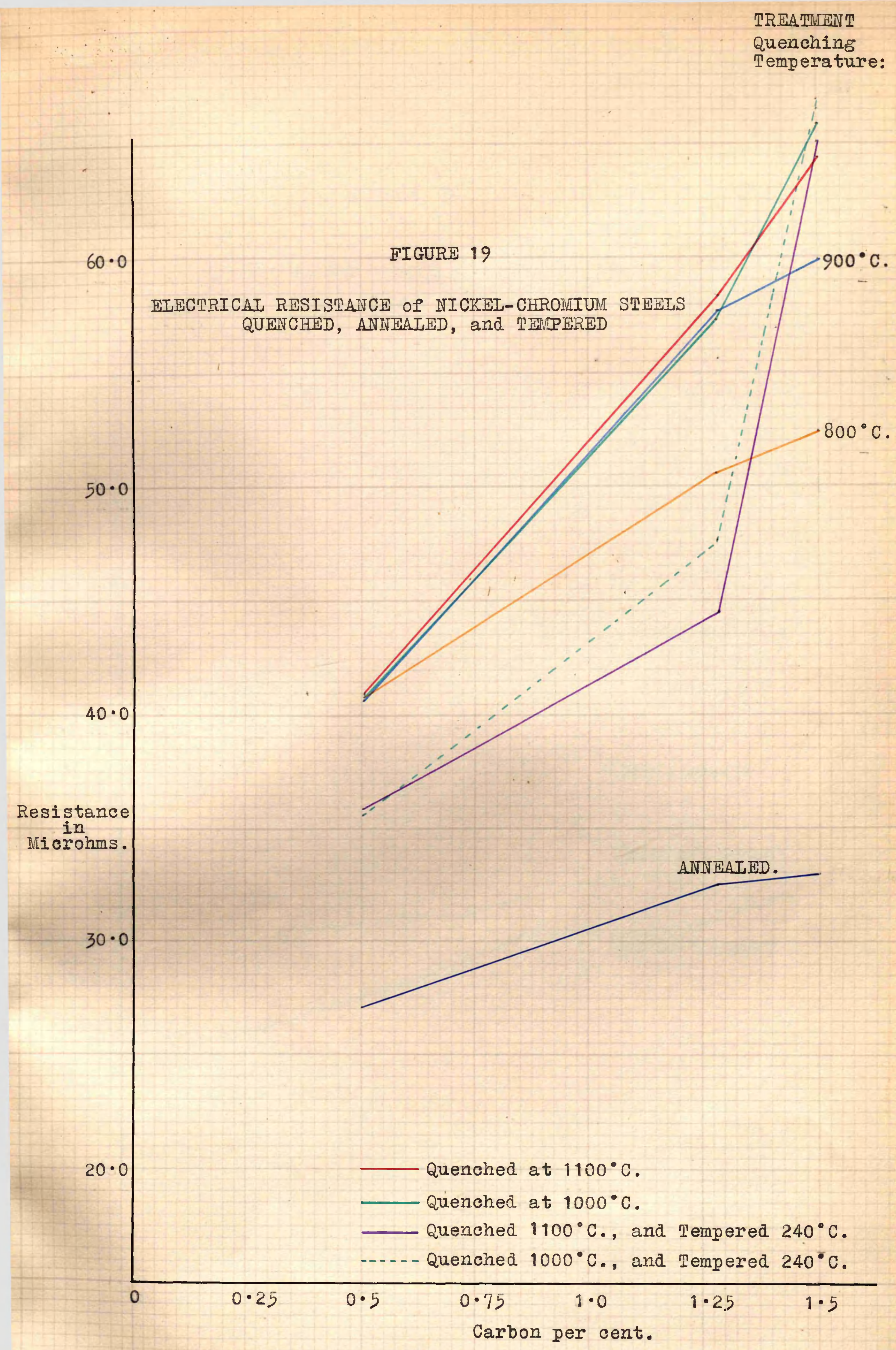
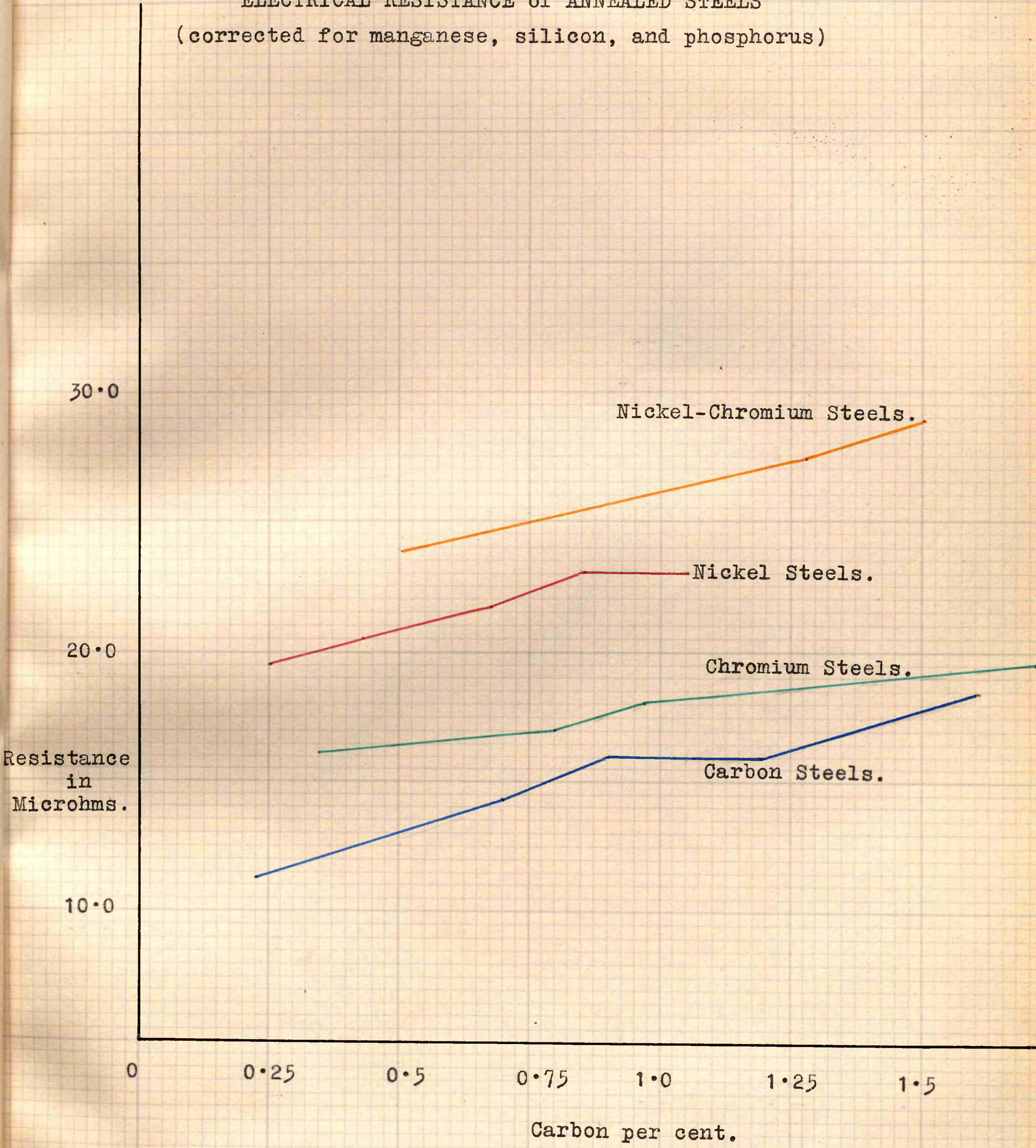


FIGURE 19

ELECTRICAL RESISTANCE of NICKEL-CHROMIUM STEELS QUENCHED, ANNEALED, and TEMPERED

FIGURE 20.

ELECTRICAL RESISTANCE of ANNEALED STEELS
(corrected for manganese, silicon, and phosphorus)



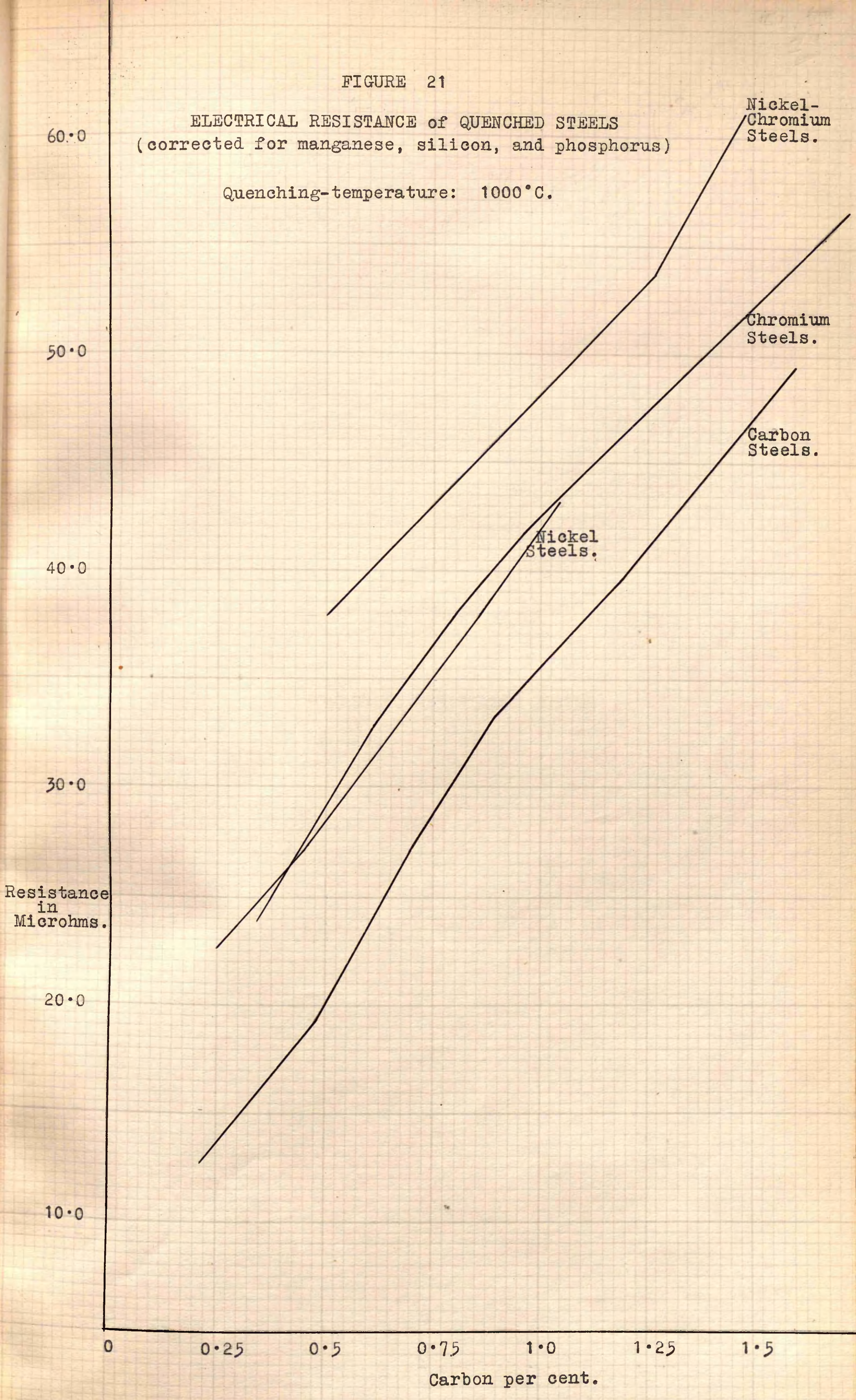


FIGURE 21

ELECTRICAL RESISTANCE of QUENCHED STEELS
(corrected for manganese, silicon, and phosphorus)

Quenching-temperature: 1000°C.

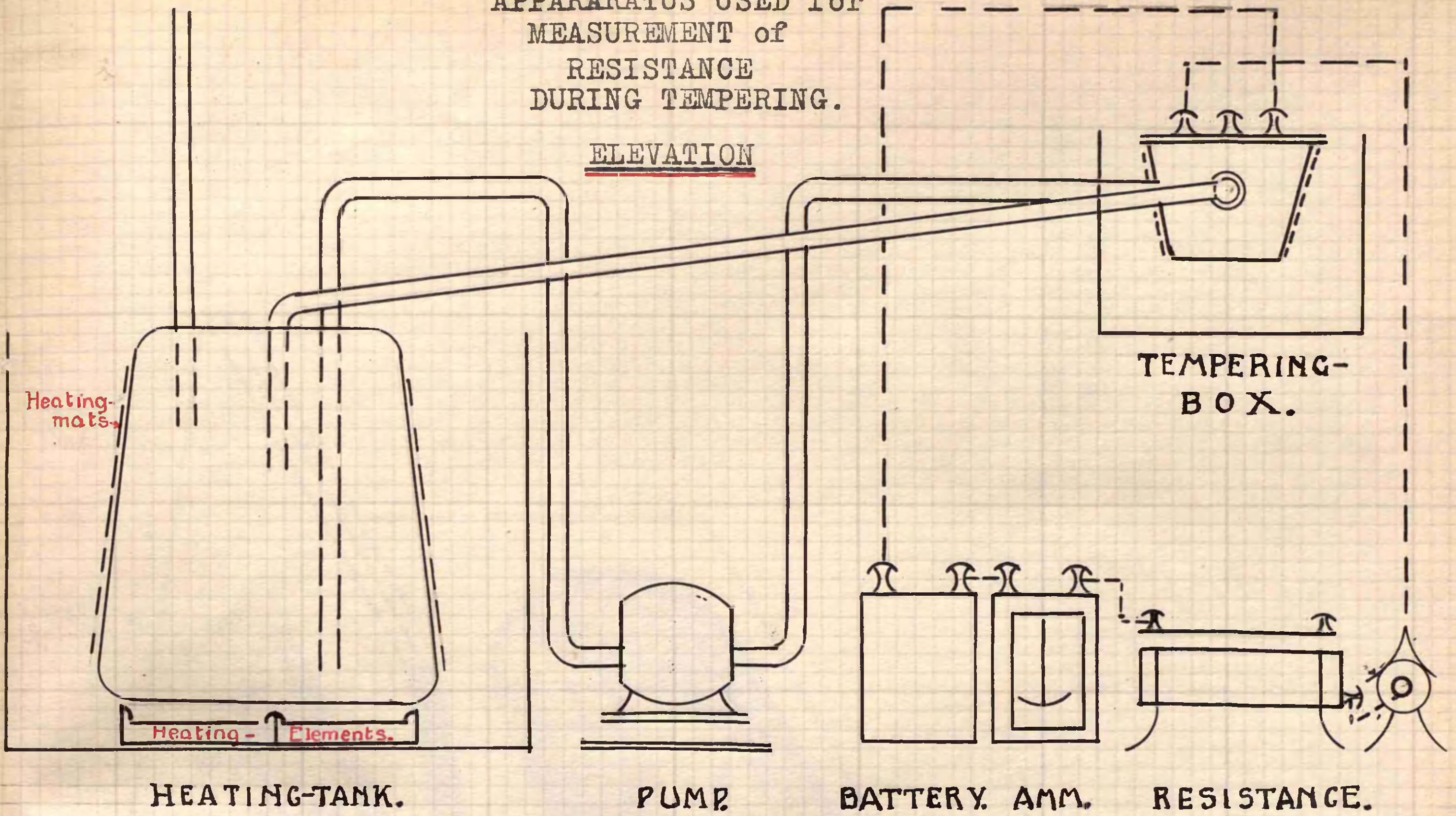
FIGURES 22 to 55

TEMPERING-RESISTANCE CURVES

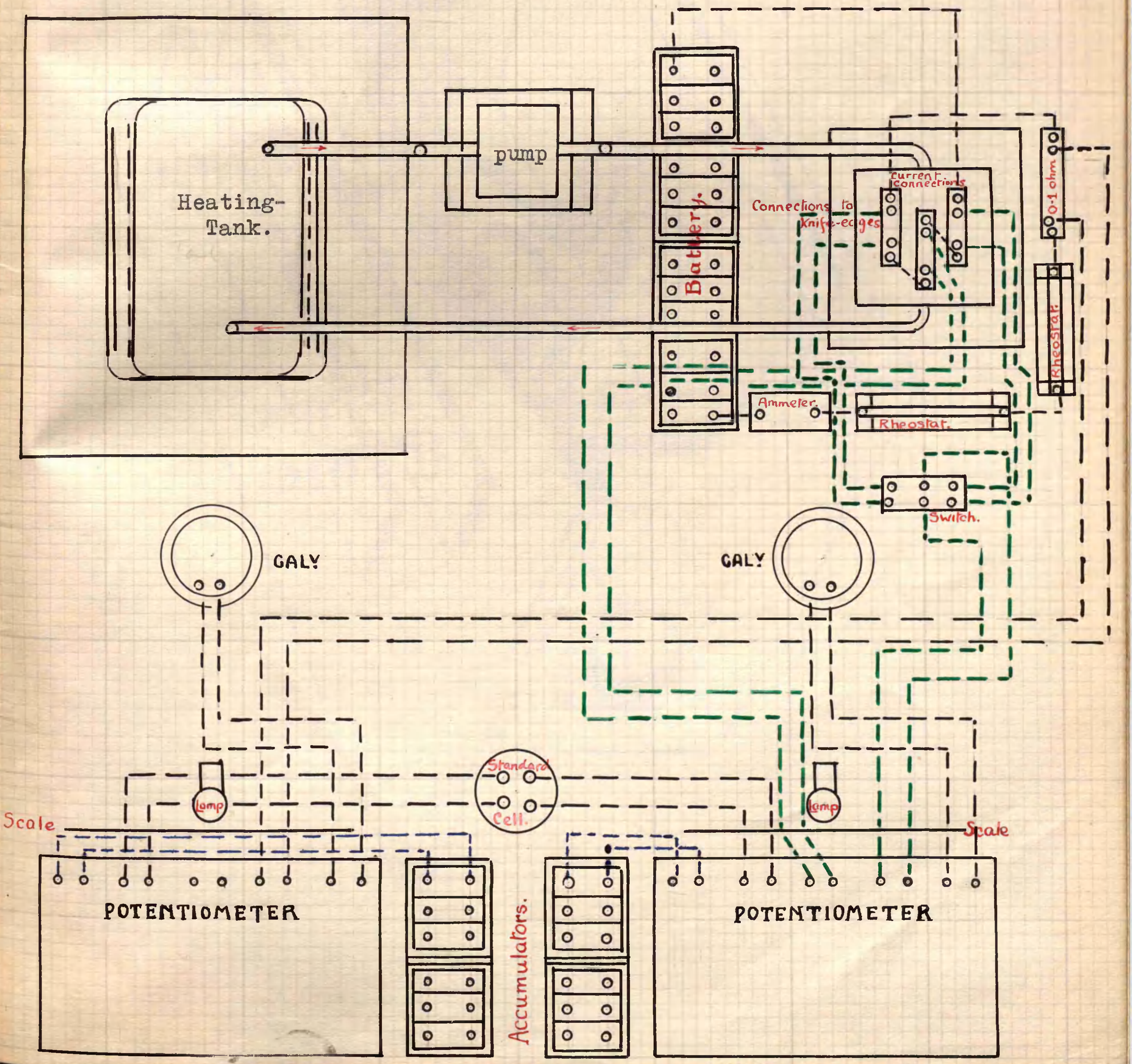
FIGURE 22.

APPARARATUS USED for
MEASUREMENT of
RESISTANCE
DURING TEMPERING.

ELEVATION



PLAN



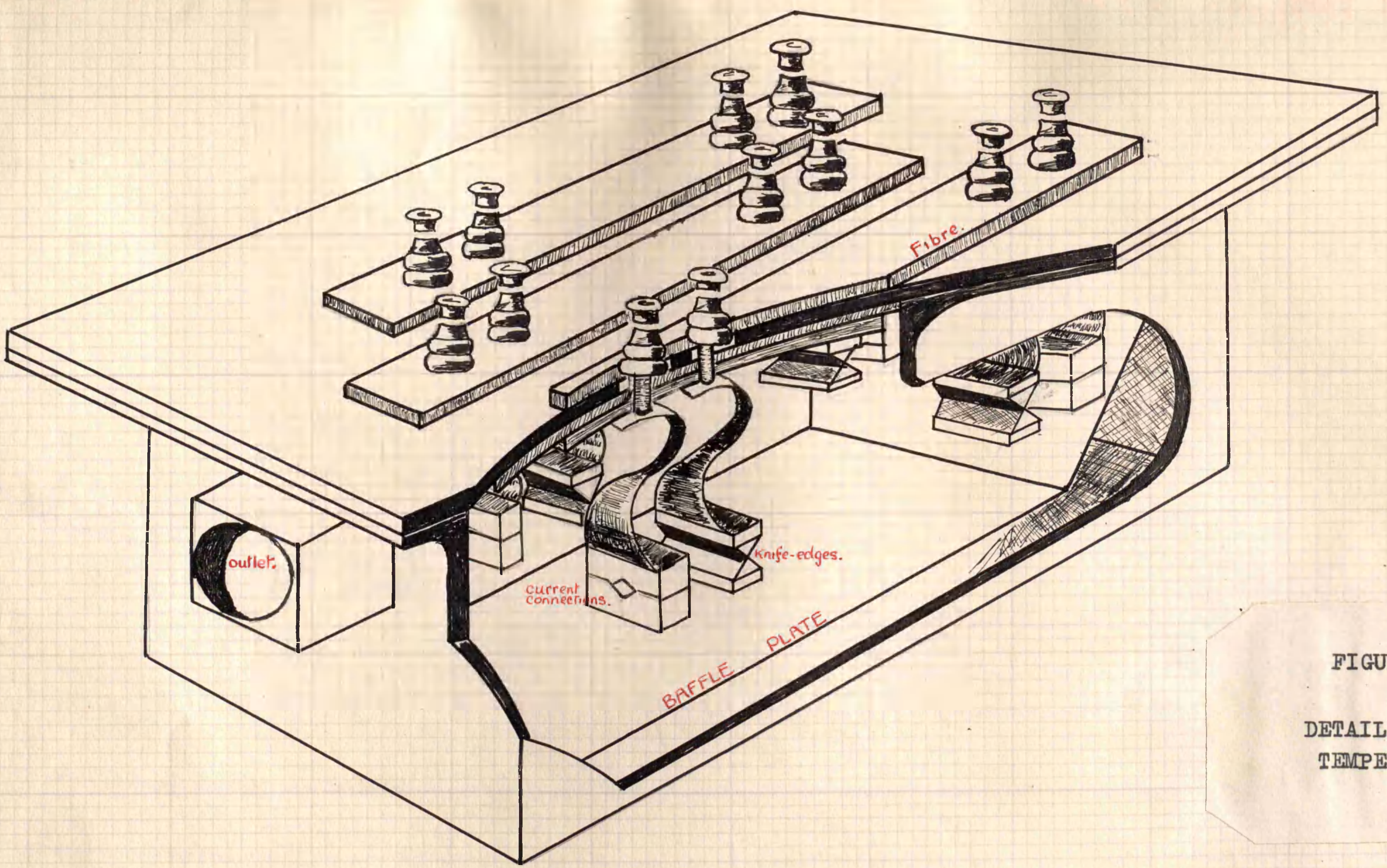


FIGURE 23

DETAILED VIEW of TEMPERING-BOX.

Figure 24.

Carbon Steels Quenched 900°C.

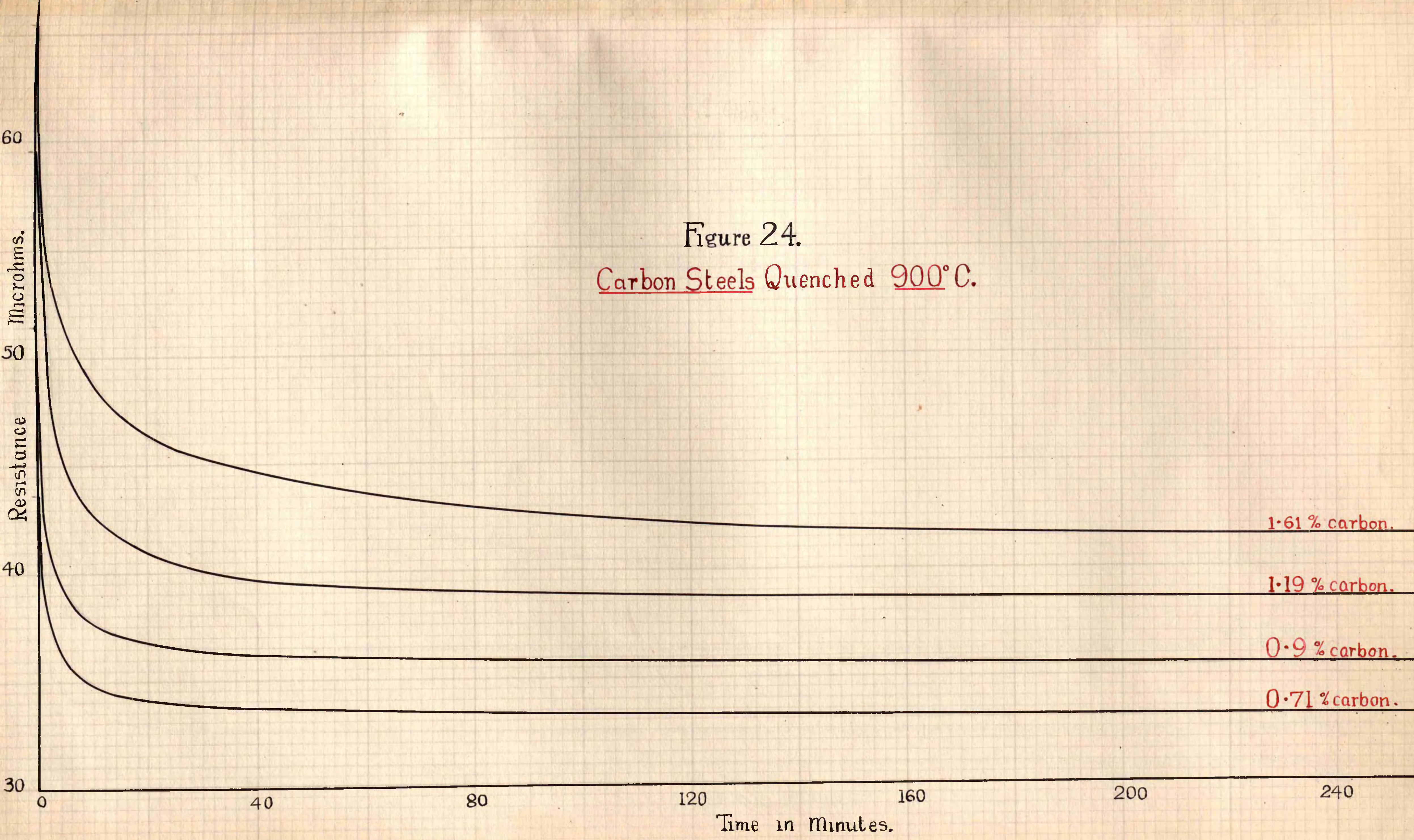


FIGURE 25.

CARBON STEELS QUENCHED 1000° C.

Points Calculated; See Vol. I Page 114.

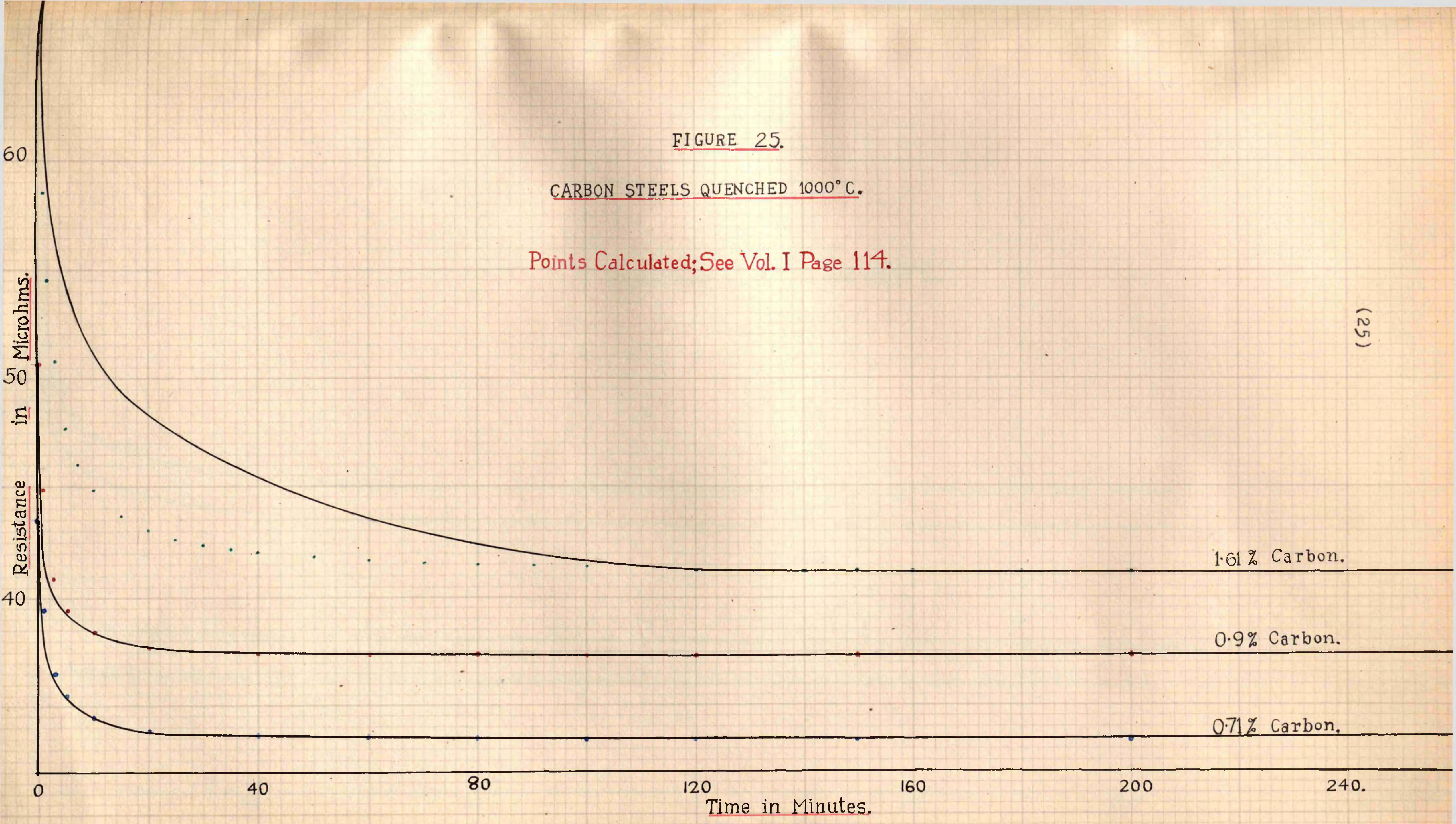
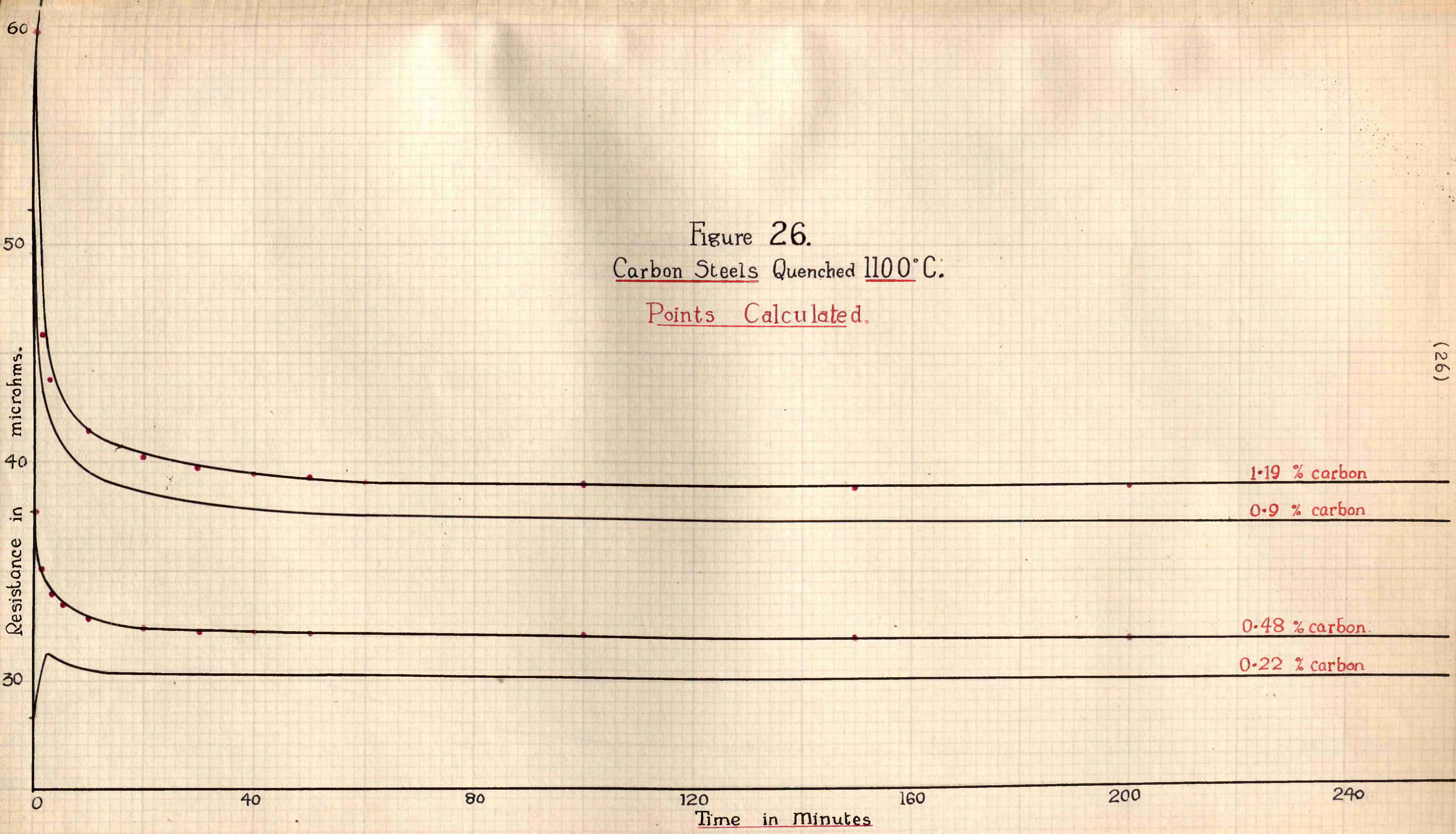


Figure 26.

Carbon Steels Quenched 1100°C.

Points Calculated.



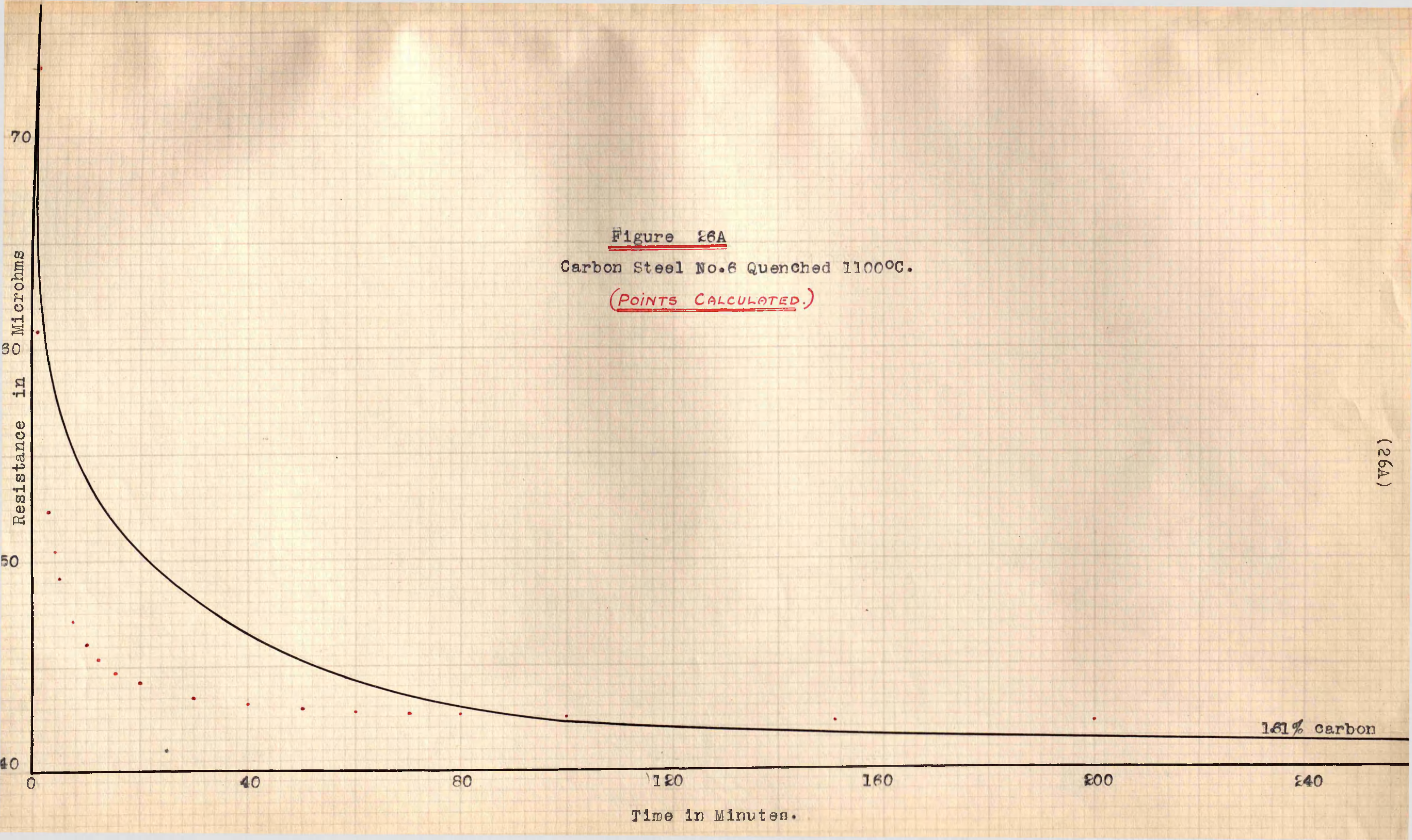


Figure 26A

Carbon Steel No.6 Quenched 1100°C.

(POINTS CALCULATED.)

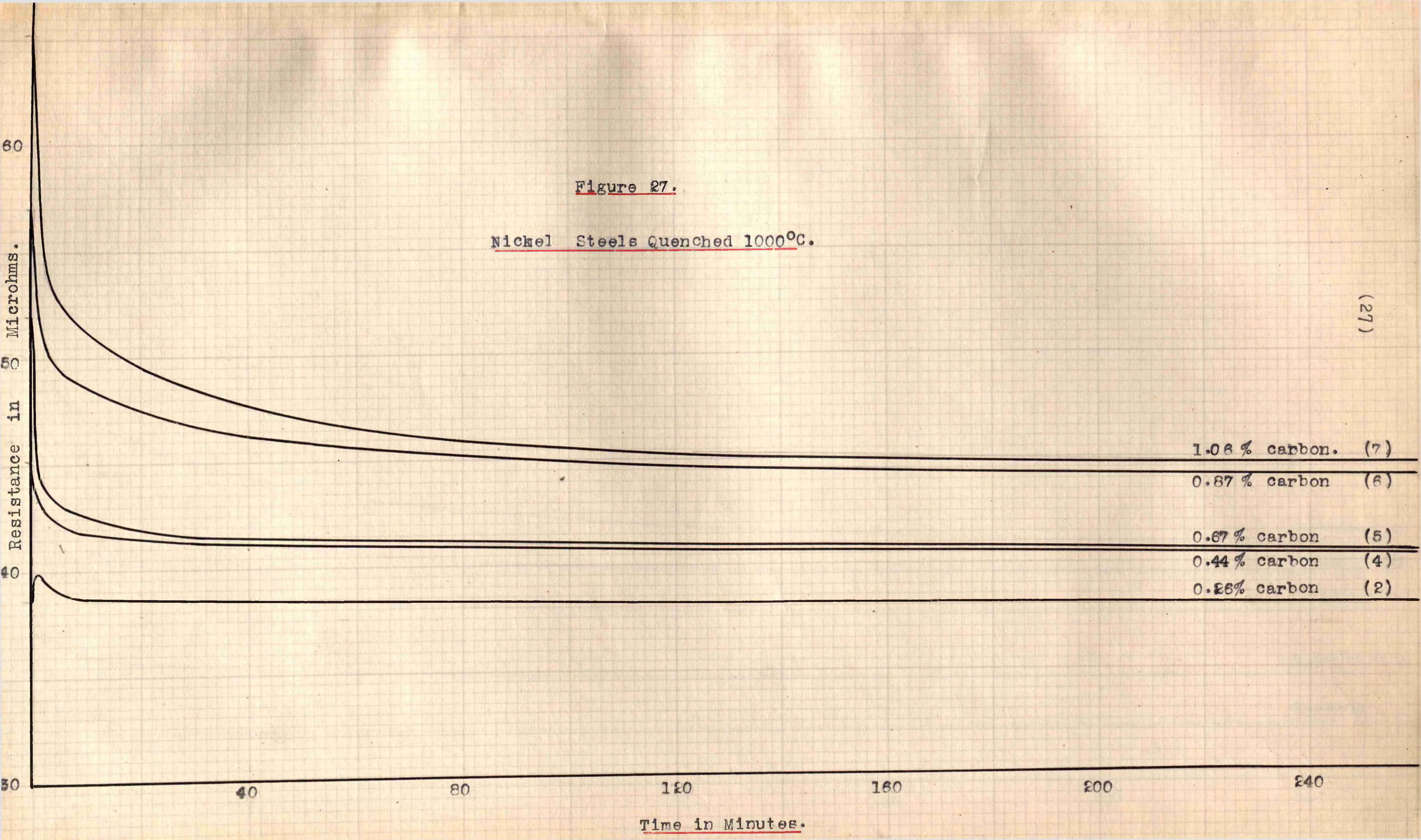


Figure 27.

Nickel Steels Quenched 1000°C.

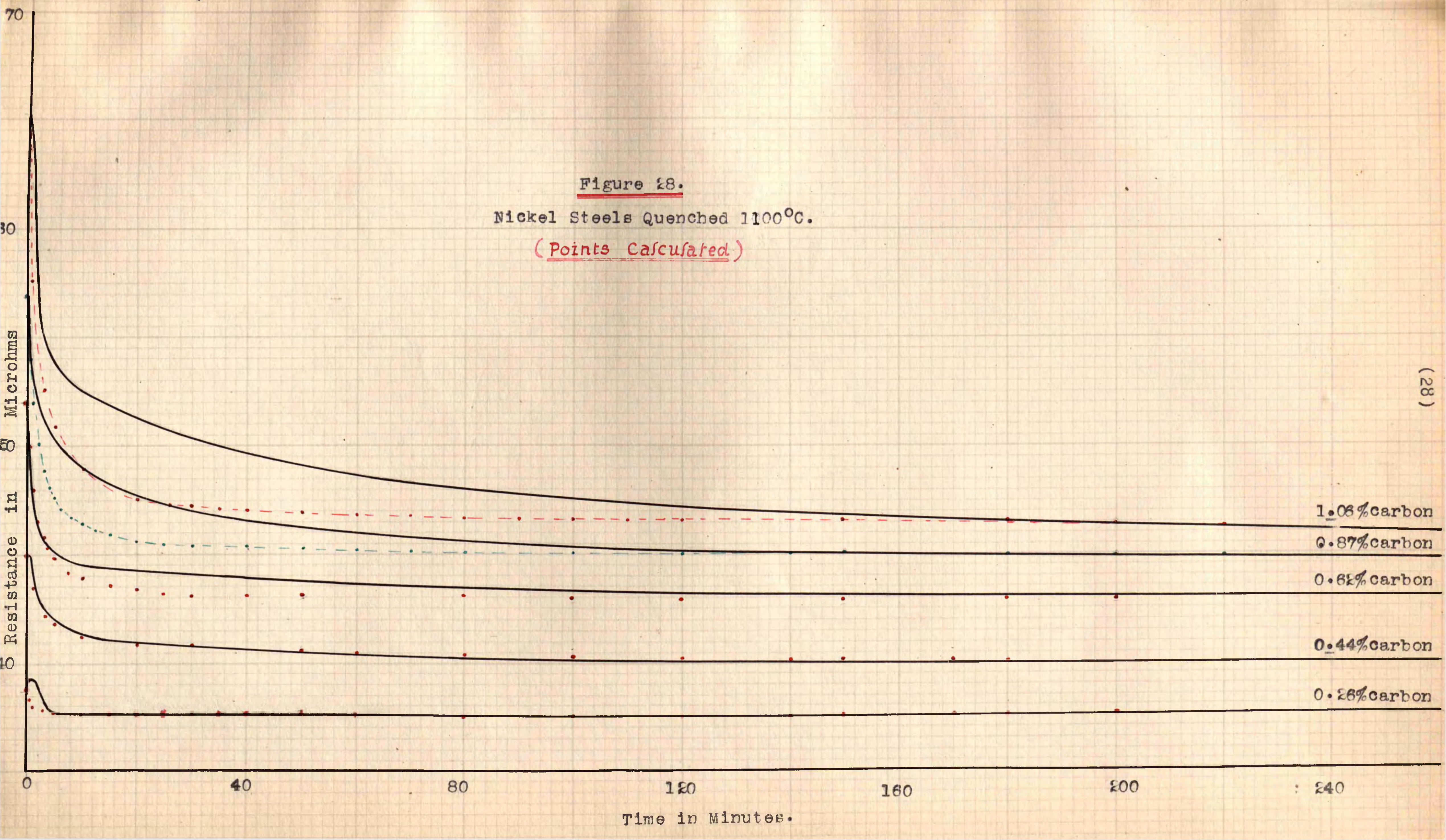


Figure 28.

Nickel Steels Quenched 1100°C.

(Points Calculated)

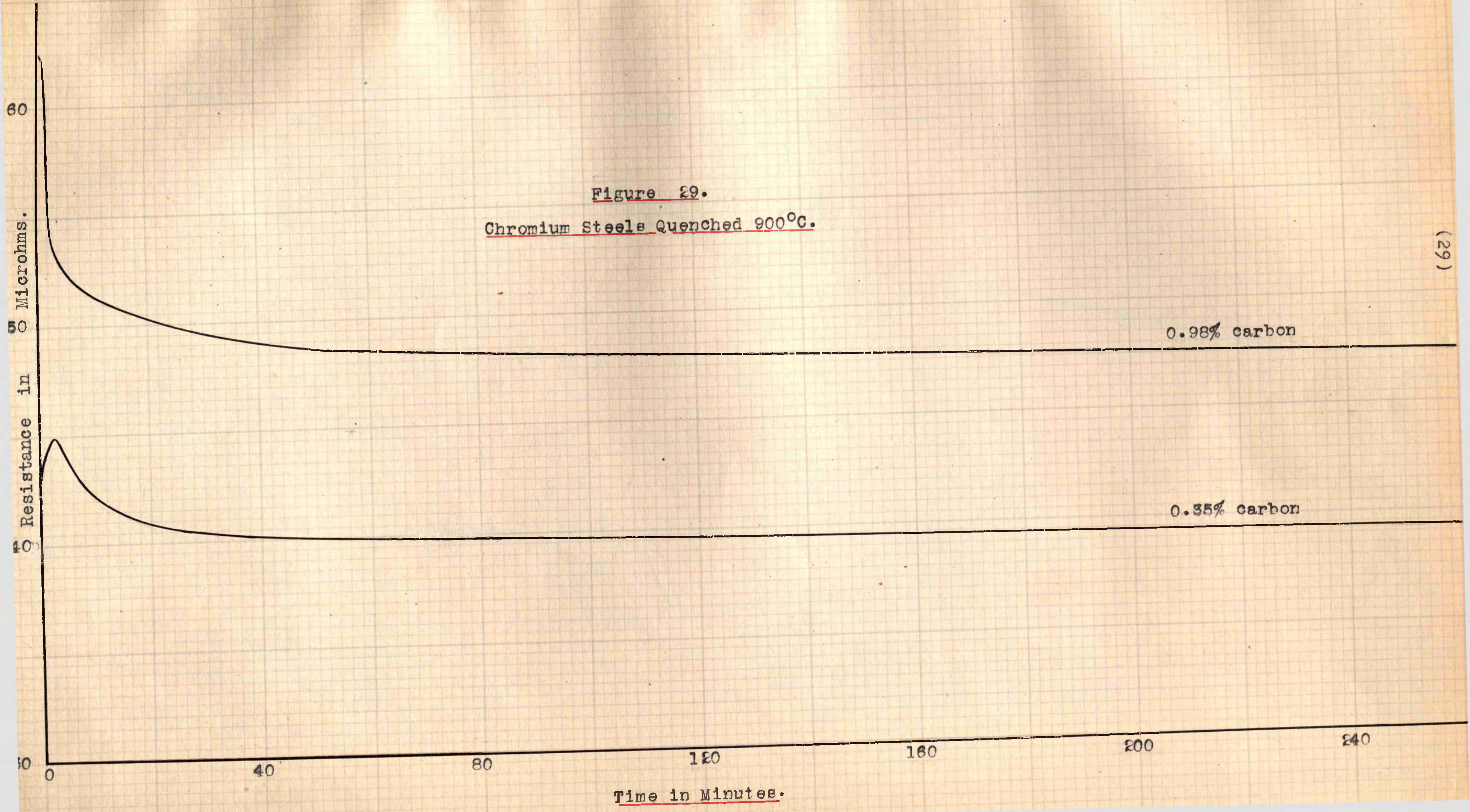
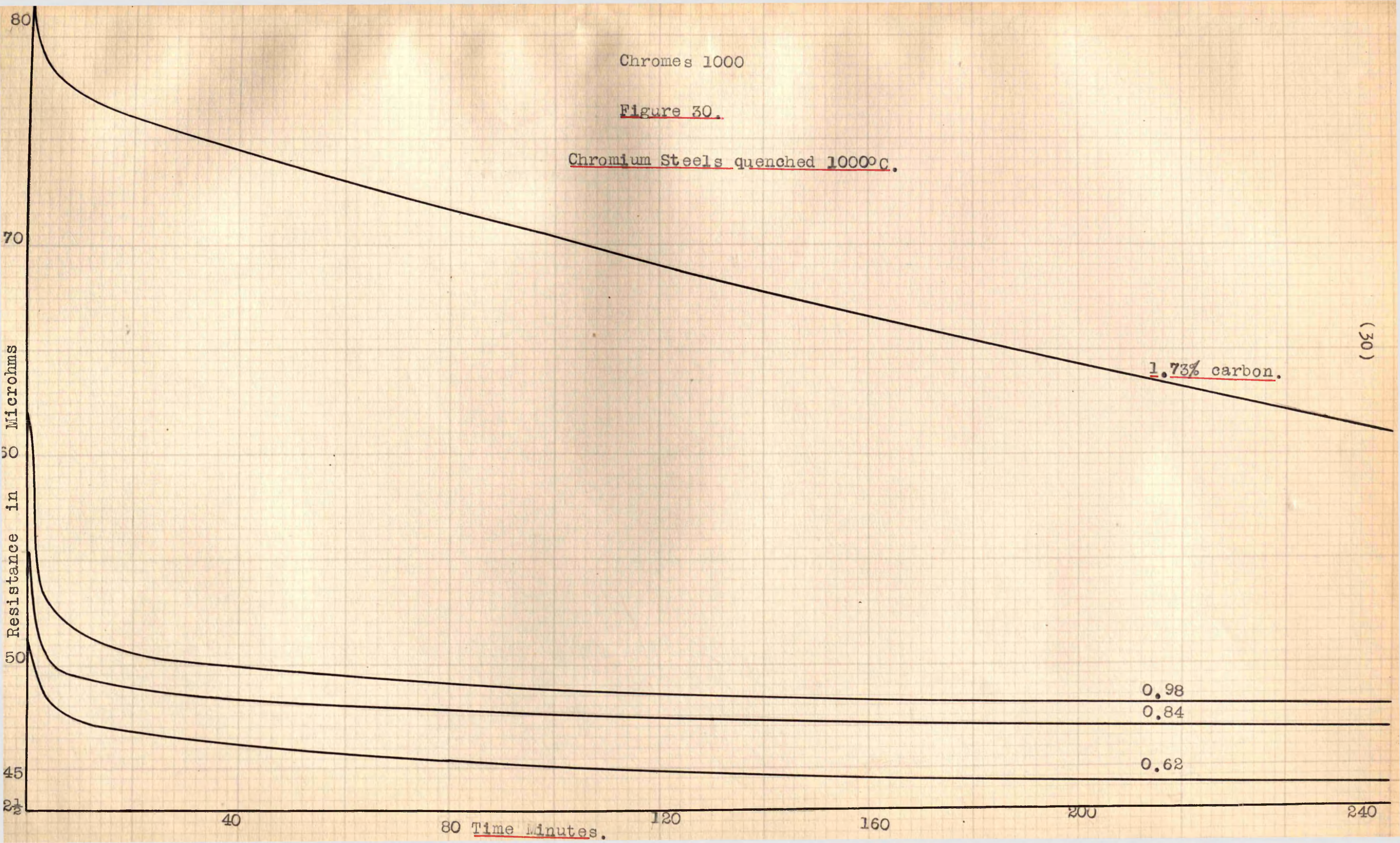


Figure 29.

Chromium Steels Quenched 900°C.



Chromes 1000

Figure 30.

Chromium Steels quenched 1000°C.

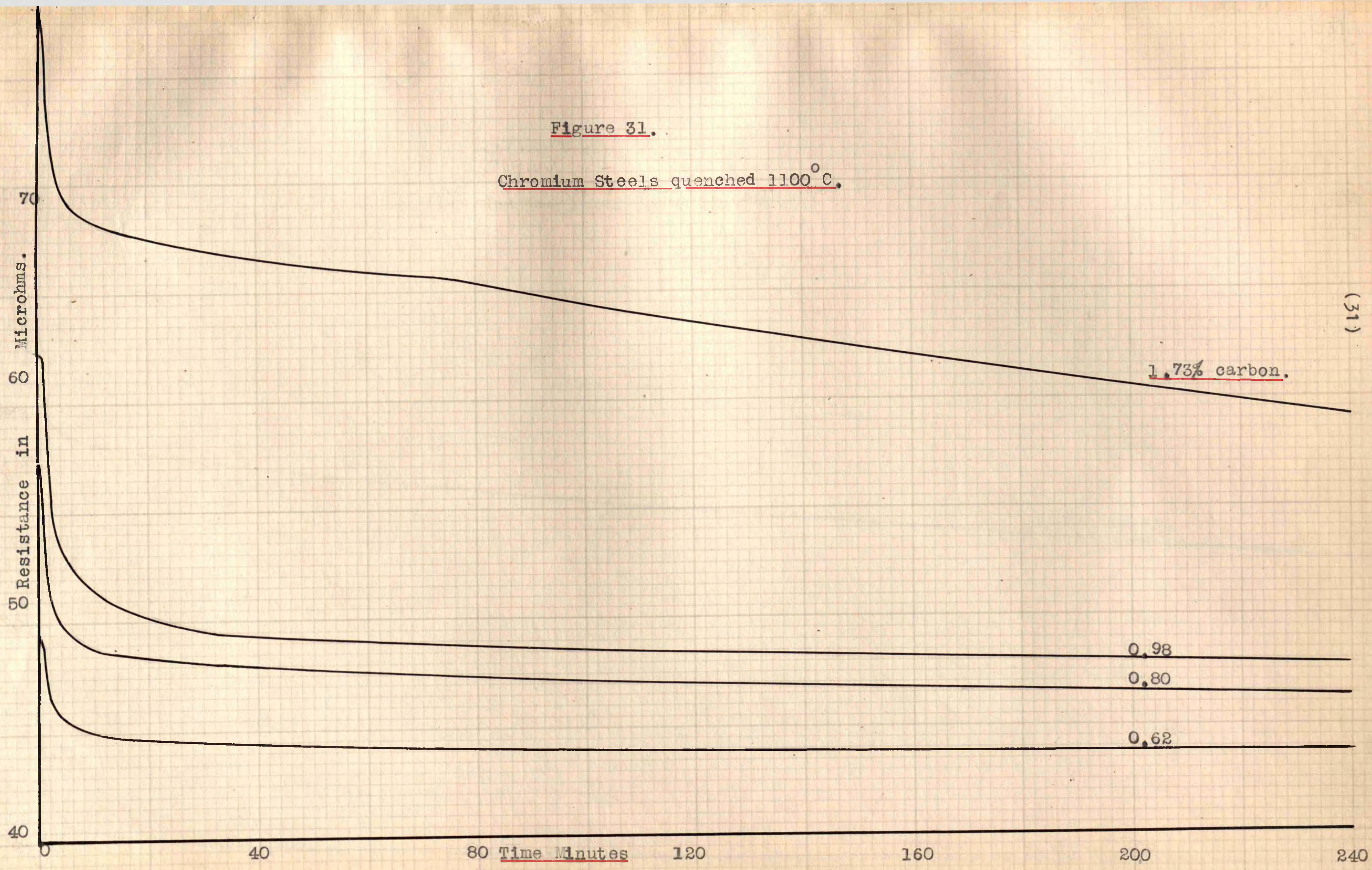


Figure 31.

Chromium Steels quenched 1100°C.

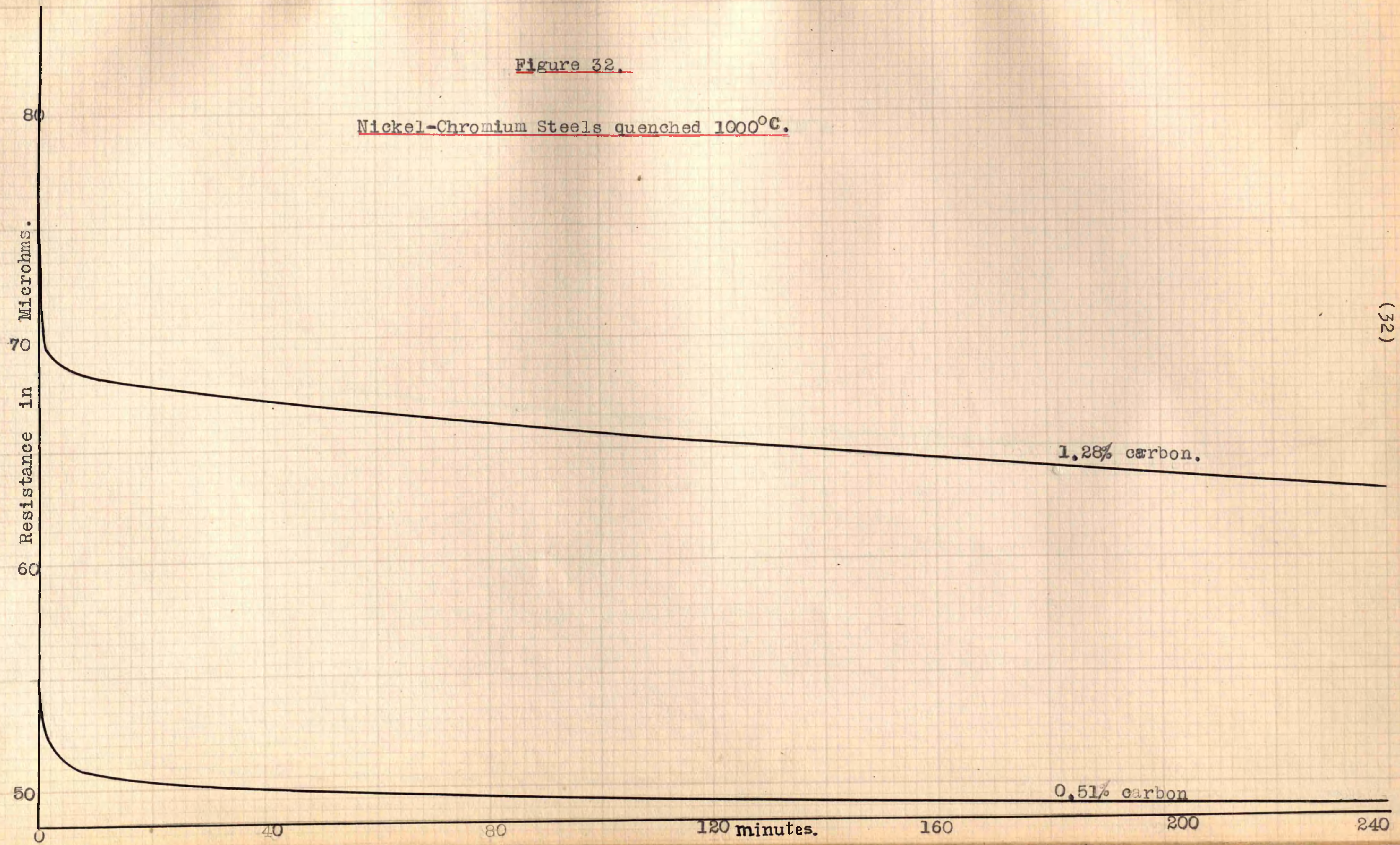


Figure 32.

Nickel-Chromium Steels quenched 1000°C.

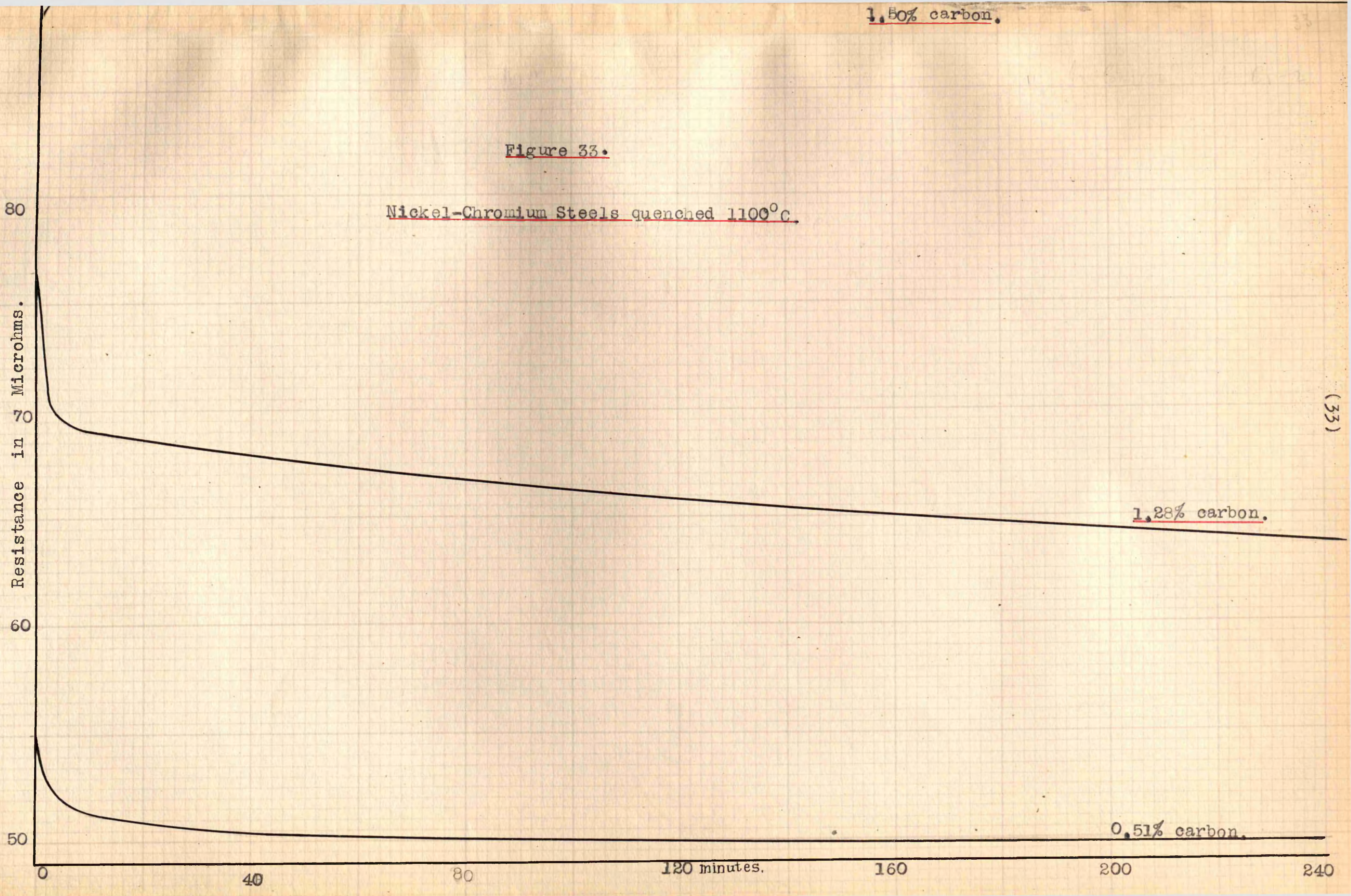


Figure 33.

Nickel-Chromium Steels quenched 1100°C

FIGURE 34.

Interrupted Tempering-Resistance Curve
of Chromium Steel B.

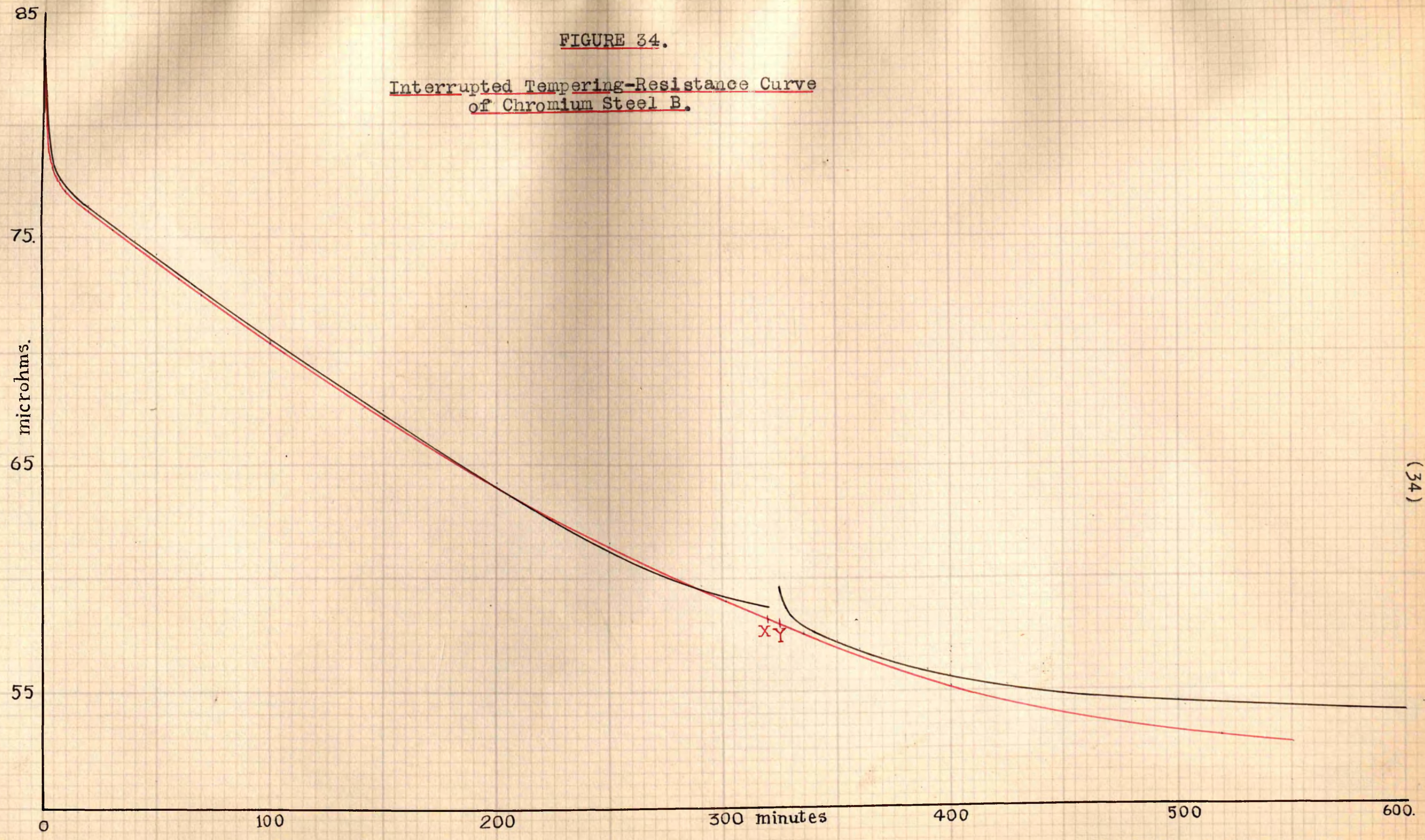


FIGURE 35.
Interrupted Tempering-Resistance Curve
of Nickel-Chromium Steel G1.

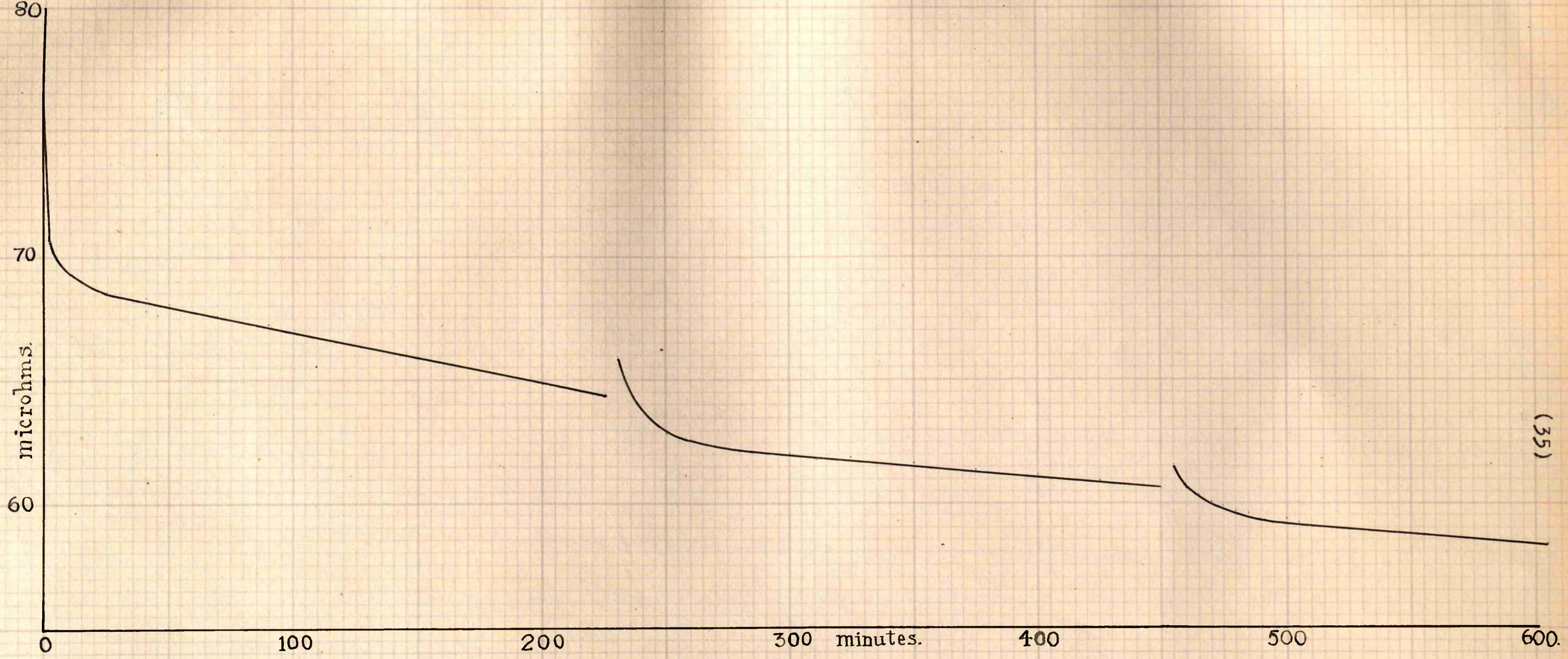


FIGURE 36.

Effect of Tempering-Temperature on the Rate of Tempering of Carbon Steel No.6 (1.19 per cent. of carbon).

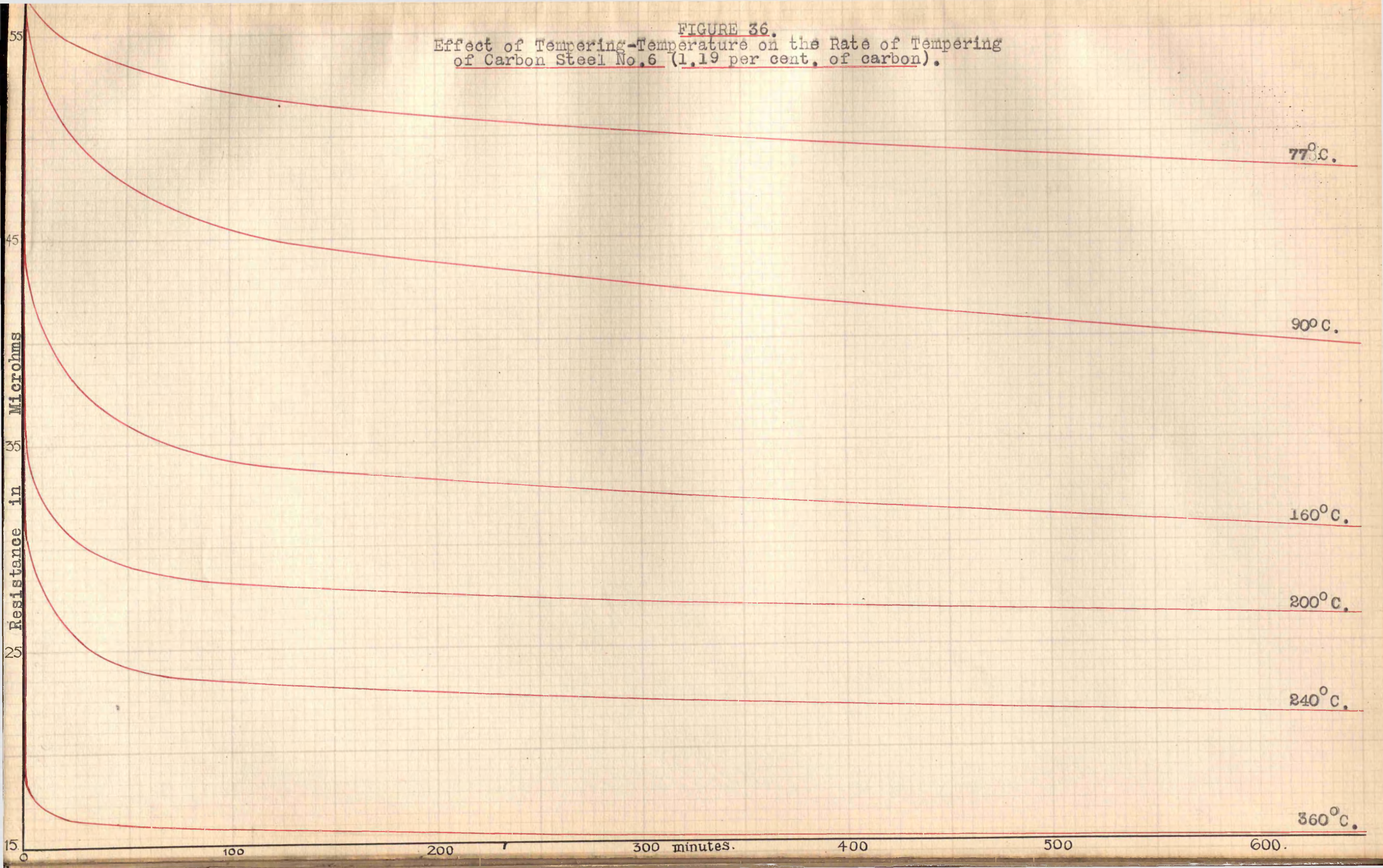


FIGURE 37

Curve showing the Rate of Tempering of a Martensitic Steel containing 1·19 per cent. of Carbon at any Temperature up to 360°C., assuming the Rate at 240°C. to be unity.

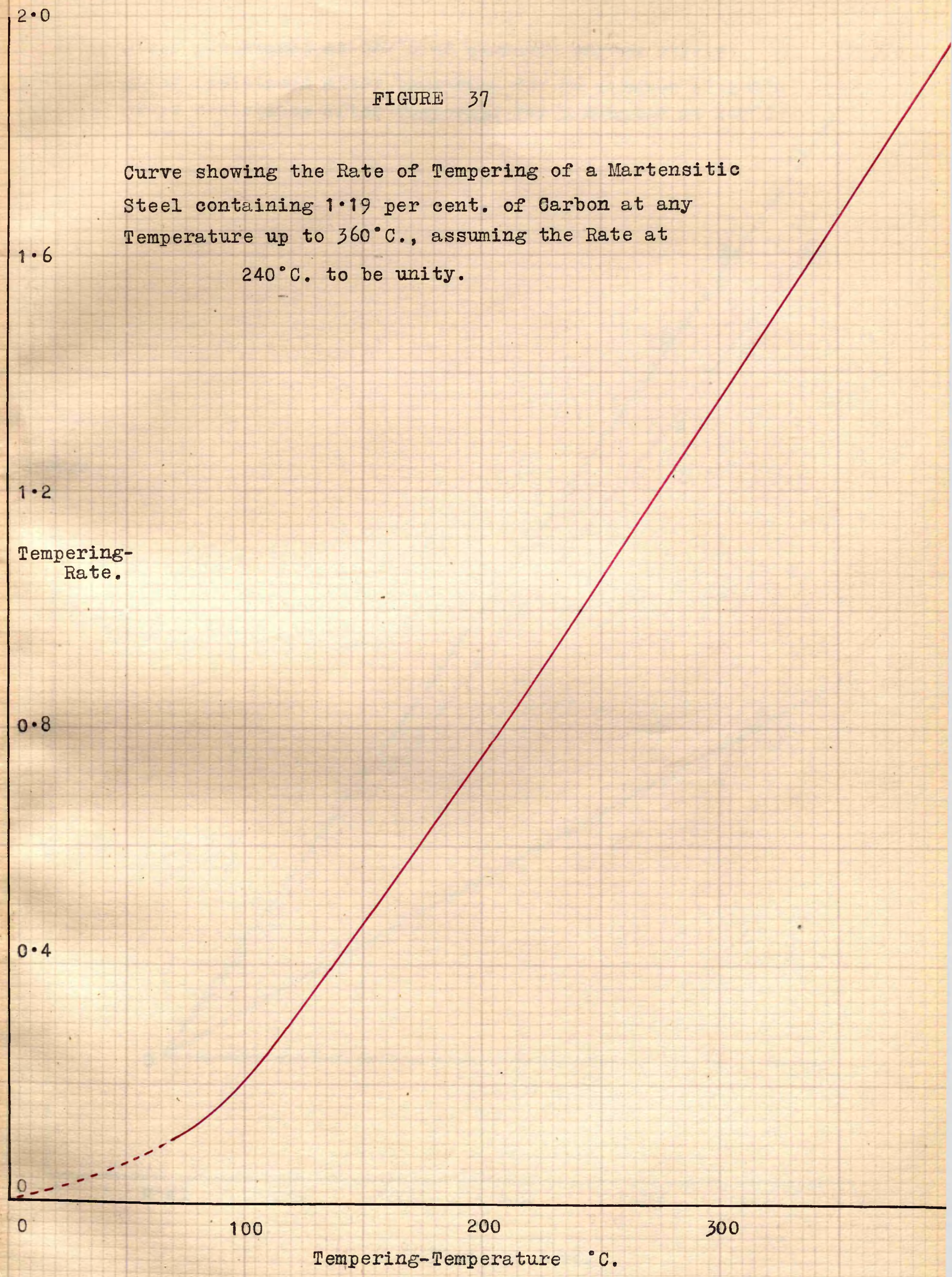
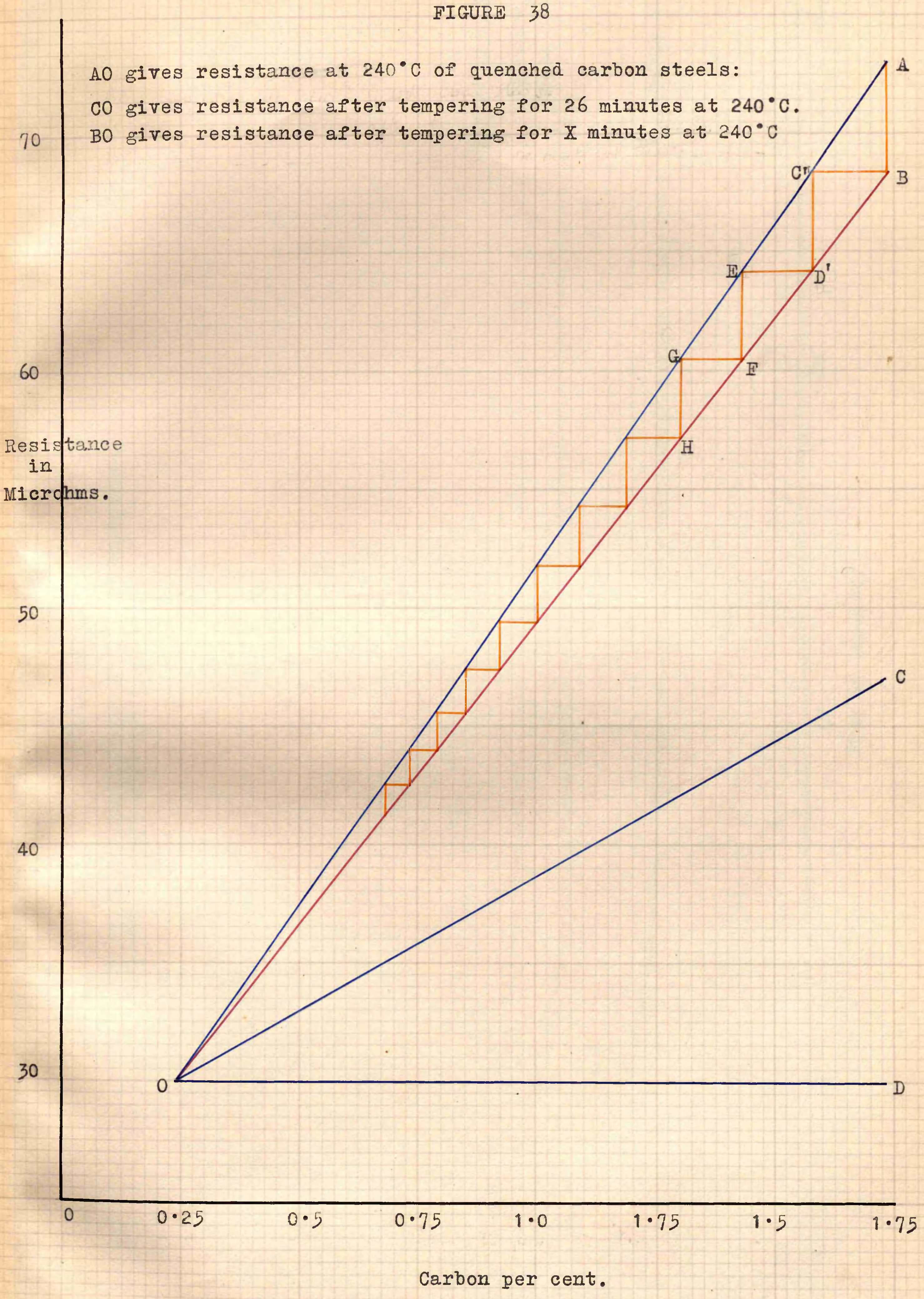


FIGURE 38

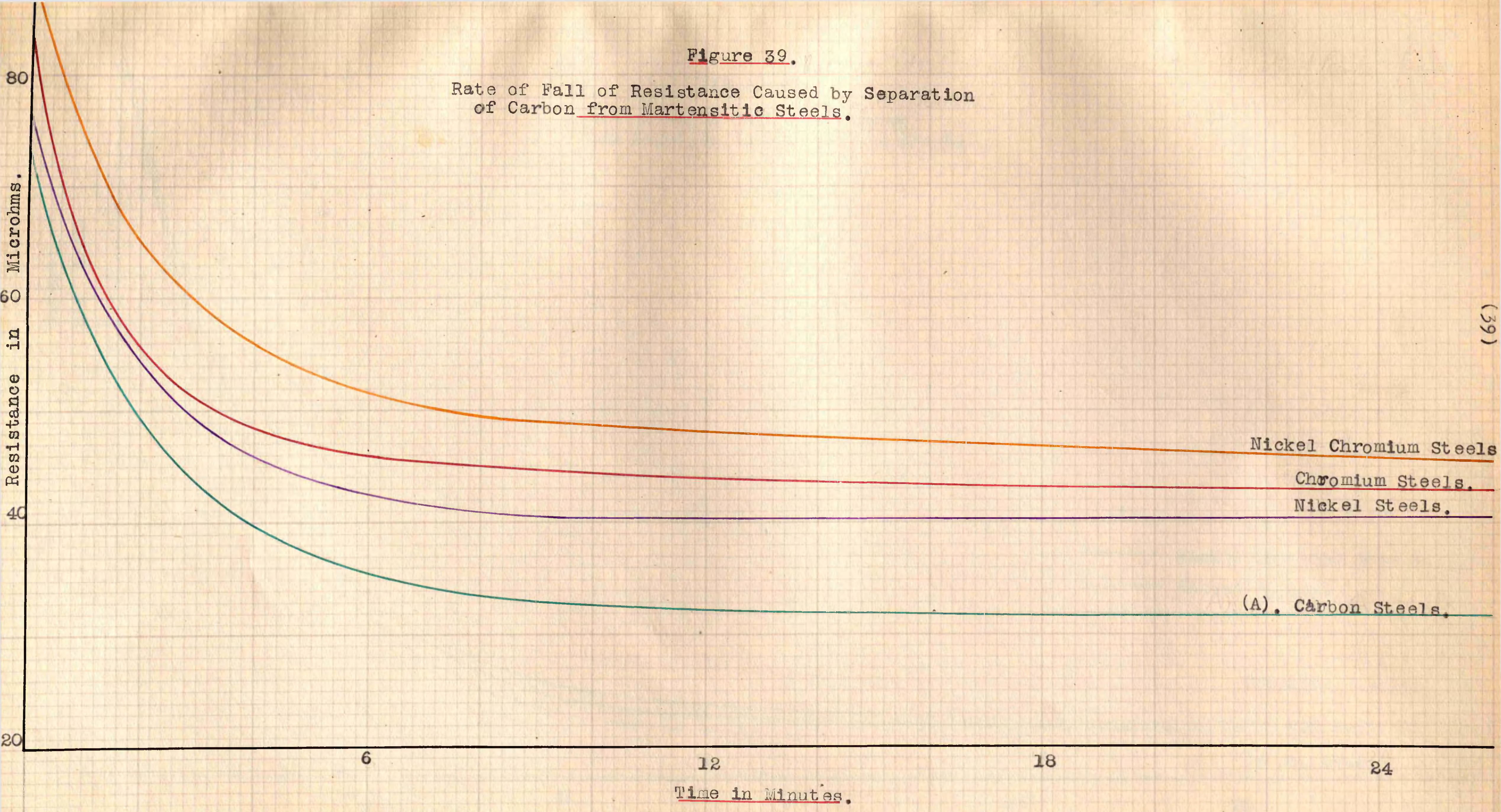


AO gives resistance at 240°C of quenched carbon steels:

CO gives resistance after tempering for 26 minutes at 240°C.

BO gives resistance after tempering for X minutes at 240°C

Figure 39.



Rate of Fall of Resistance Caused by Separation of Carbon from Martensitic Steels.

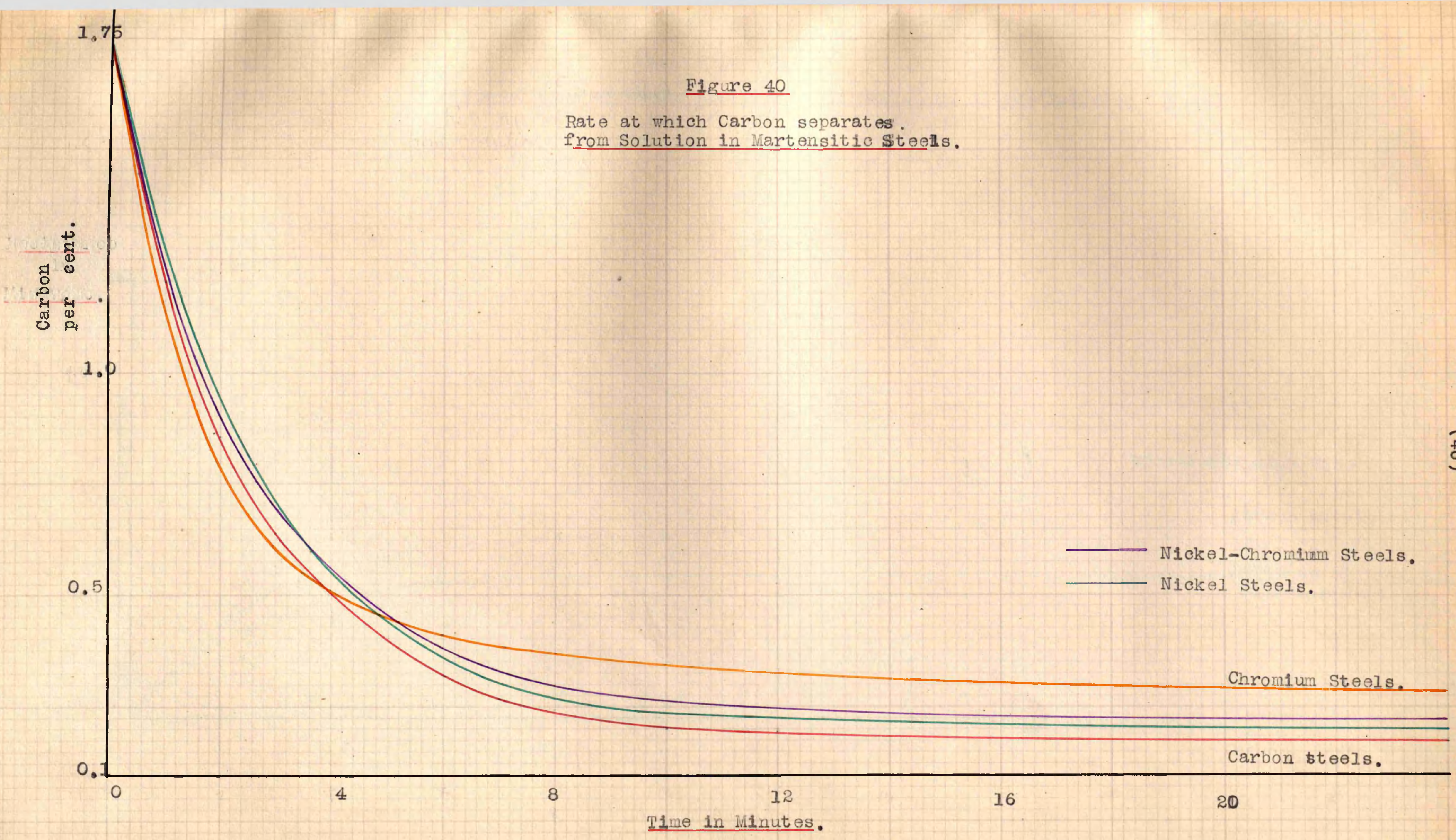


Figure 40

Rate at which Carbon separates.
from Solution in Martensitic Steels.

Figure 41.

Rate of Rise of Resistance due to Accumulation of Precipitated Carbide.

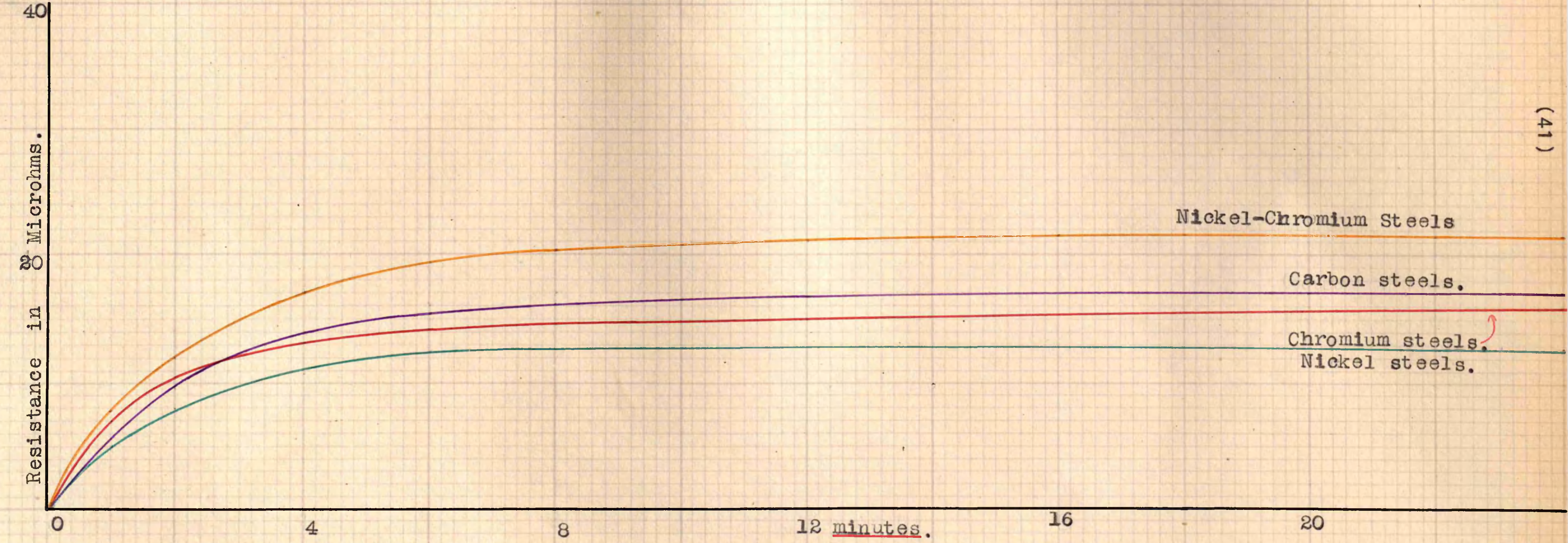
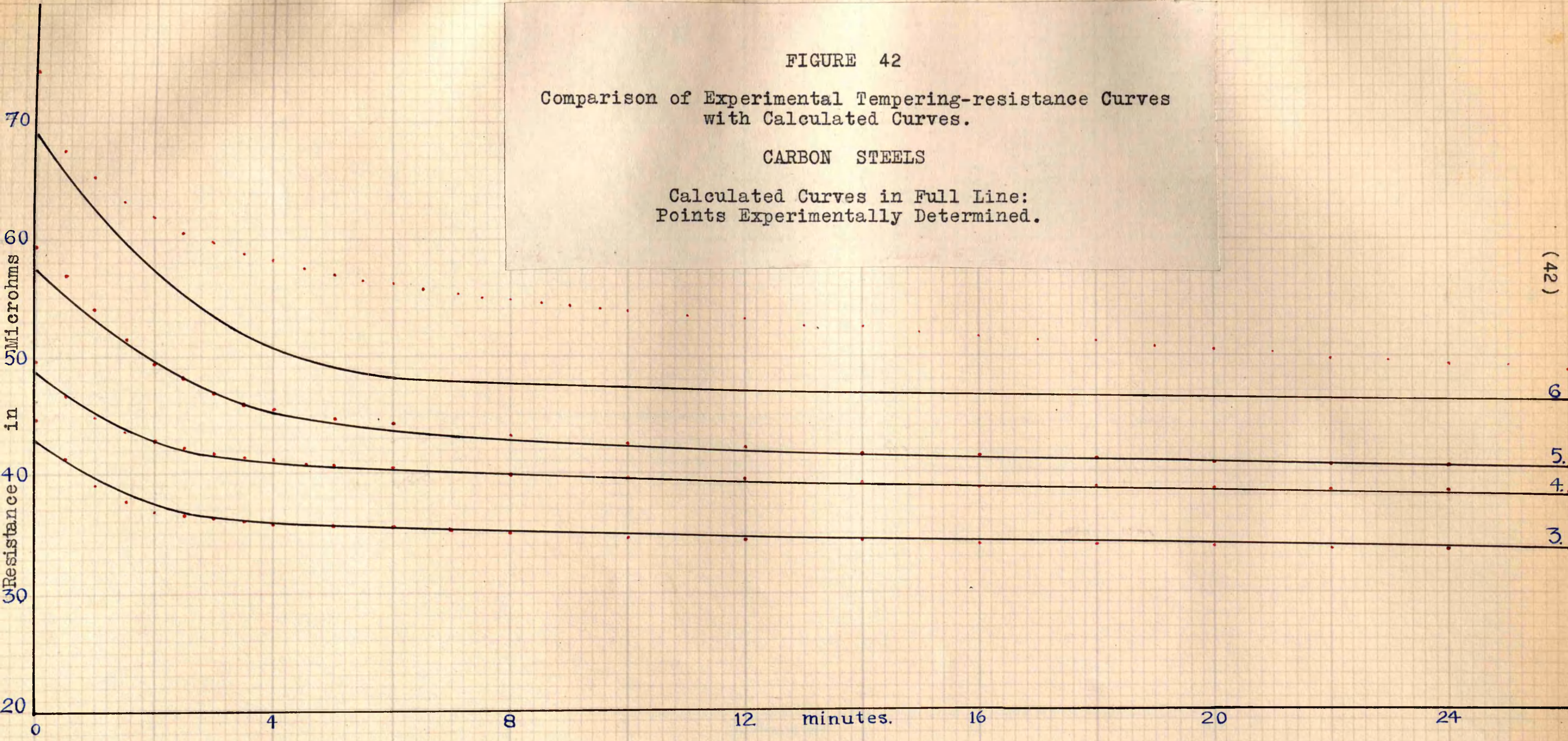


FIGURE 42

Comparison of Experimental Tempering-resistance Curves with Calculated Curves.

CARBON STEELS

Calculated Curves in Full Line:
Points Experimentally Determined.



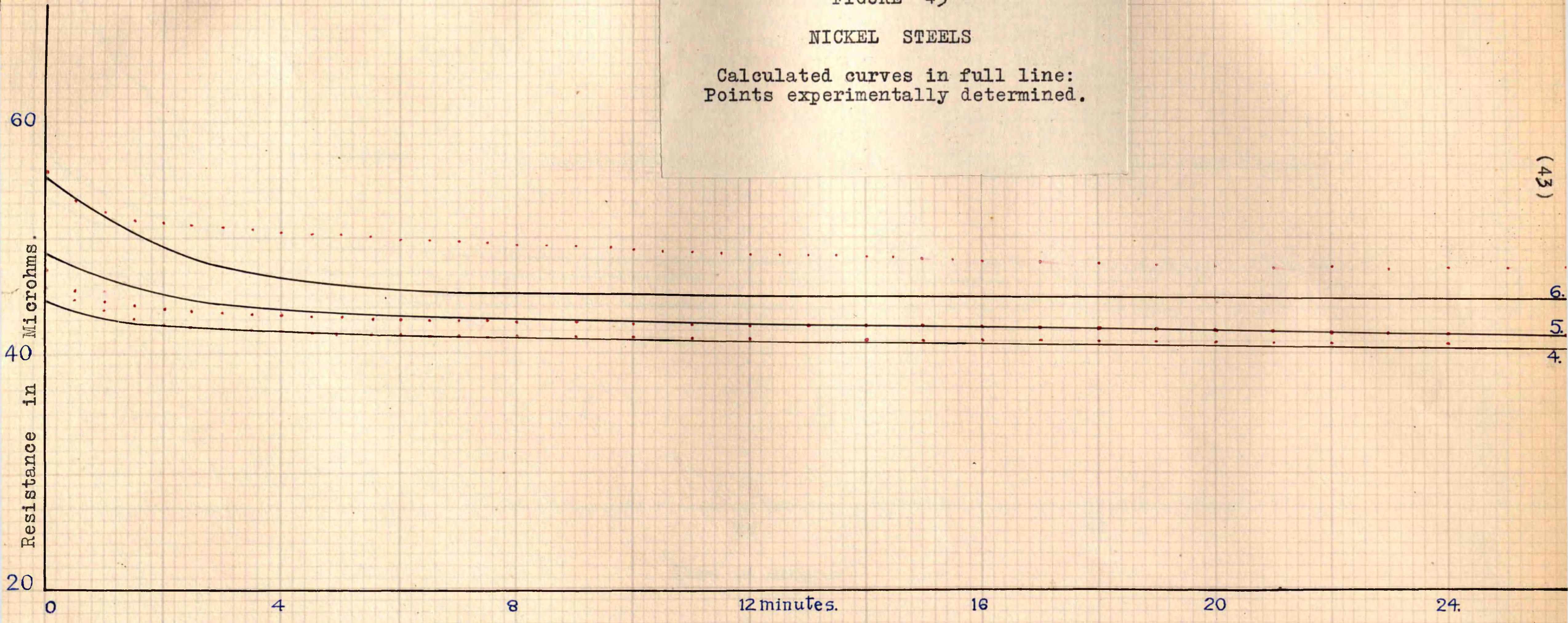


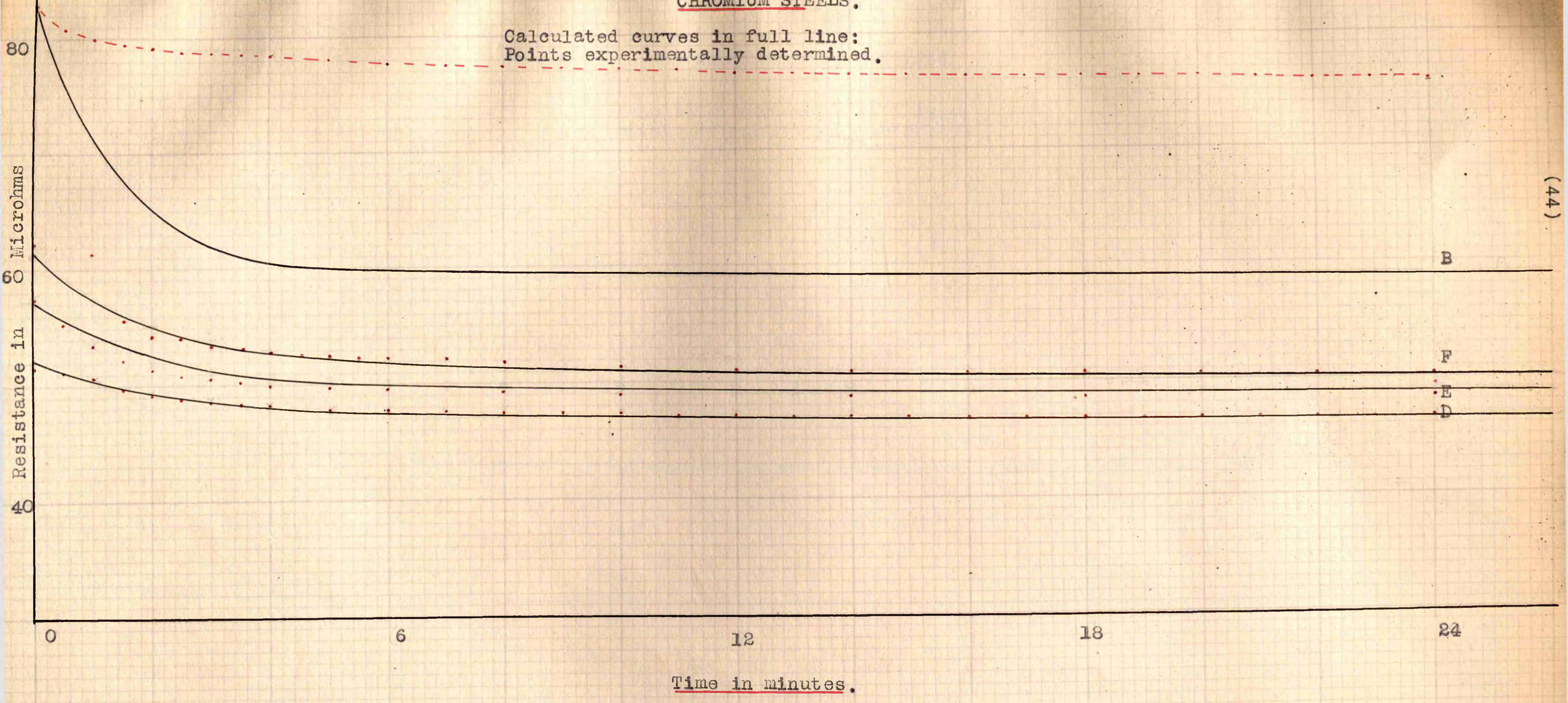
FIGURE 43

NICKEL STEELS

Calculated curves in full line:
Points experimentally determined.

FIGURE 44.

CHROMIUM STEELS.



Calculated curves in full line:
Points experimentally determined.

FIGURE 45.

NICKEL-CHROMIUM STEELS.

Calculated curves in full line:
Points experimentally determined.

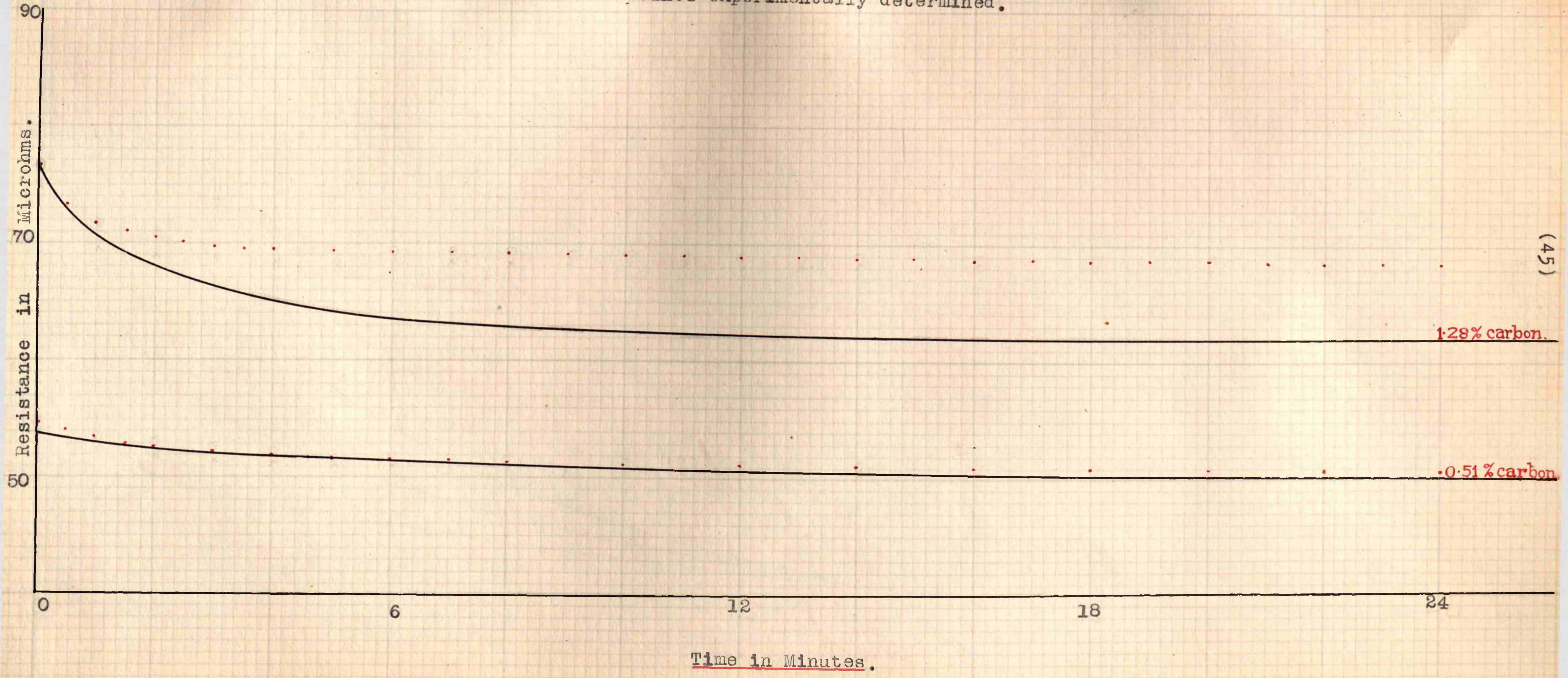


FIGURE 46.

Carbon Steel No.6 (1.61% of carbon).

Curve A: Experimental tempering-curve.

Curve B: Theoretical curve for martensite of the same composition.

Curve C: Theoretical curve for austenite of the same composition.

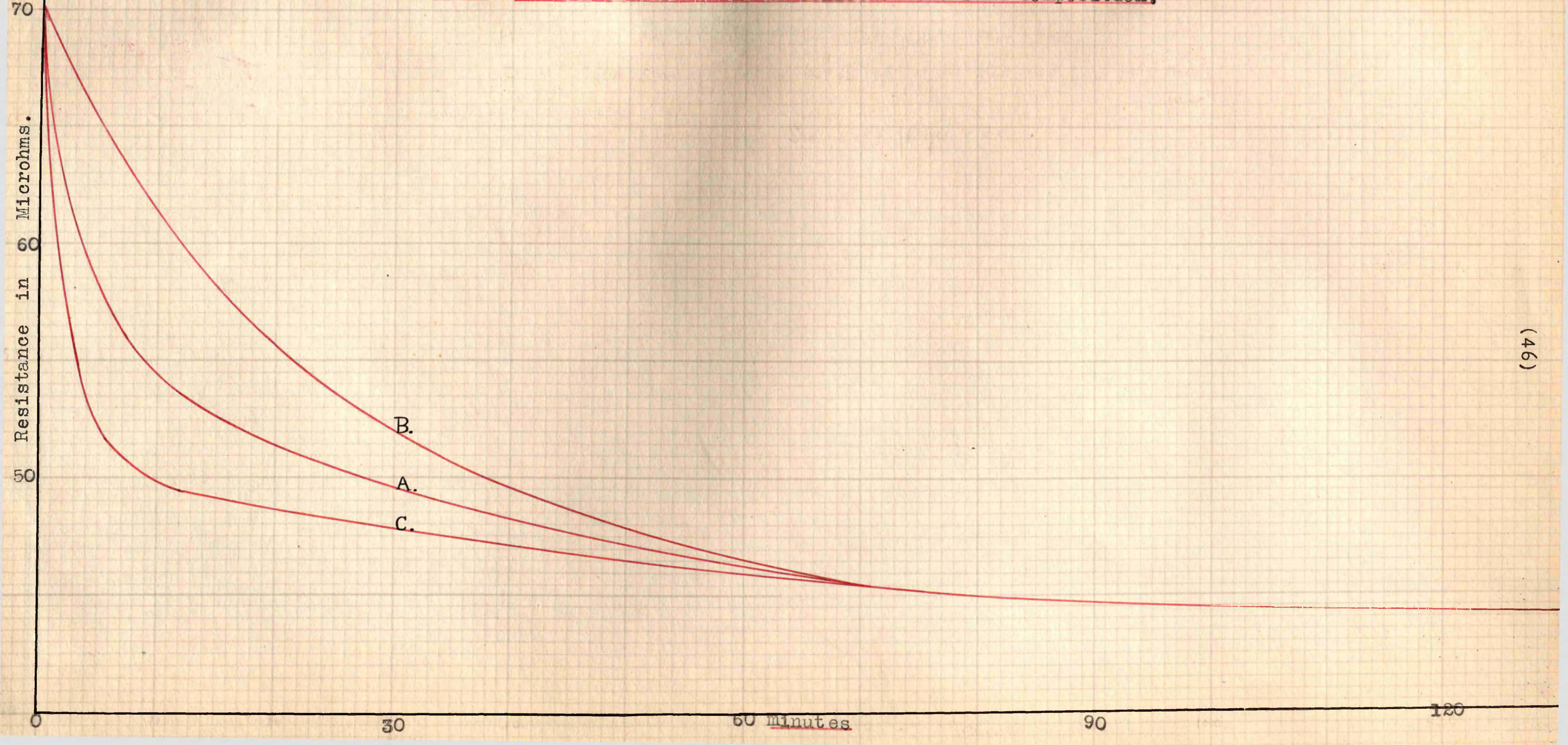


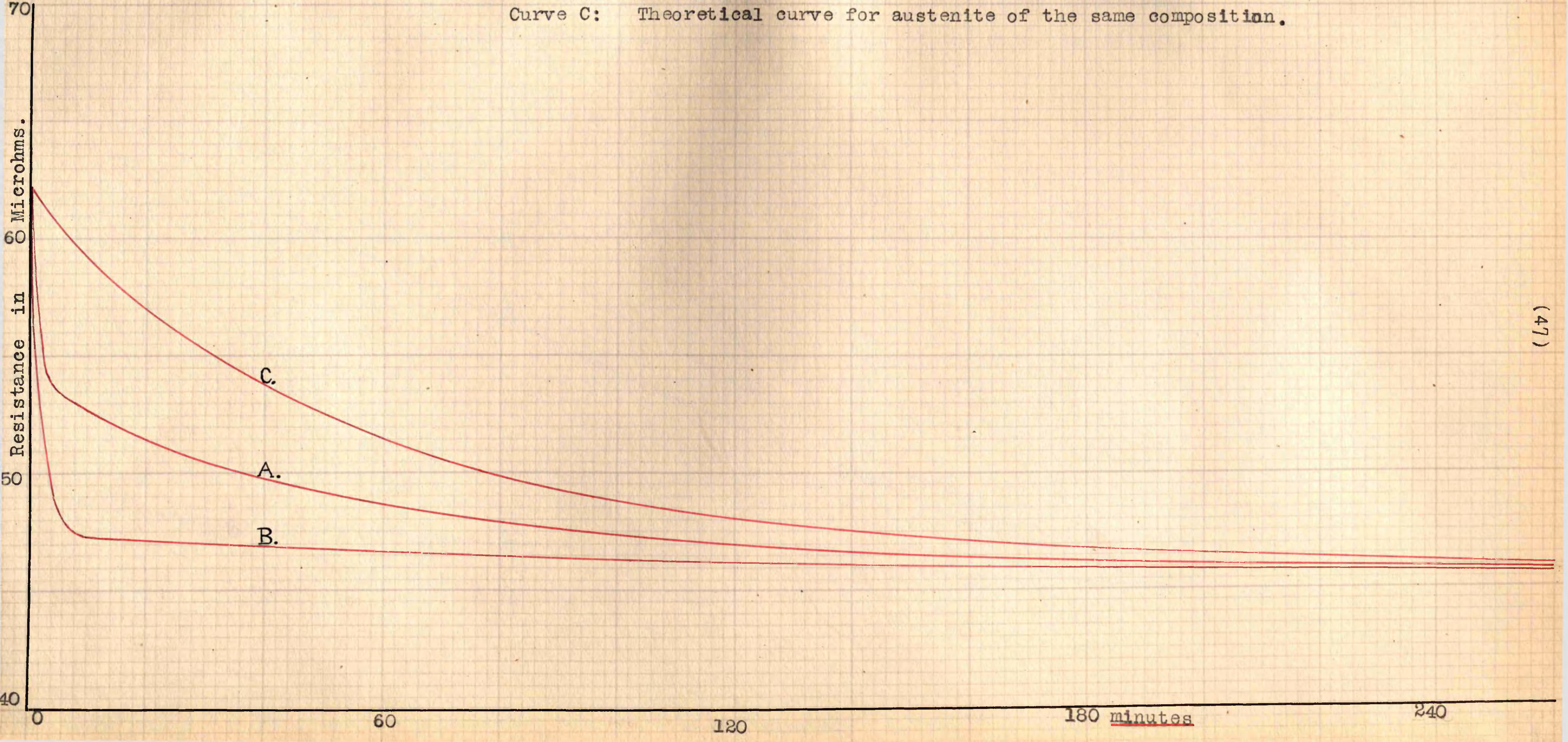
FIGURE 47.

Nickel Steel No.7 (1.06% carbon).

Curve A: Experimental curve.

Curve B: Theoretical curve for martensite of the same composition.

Curve C: Theoretical curve for austenite of the same composition.



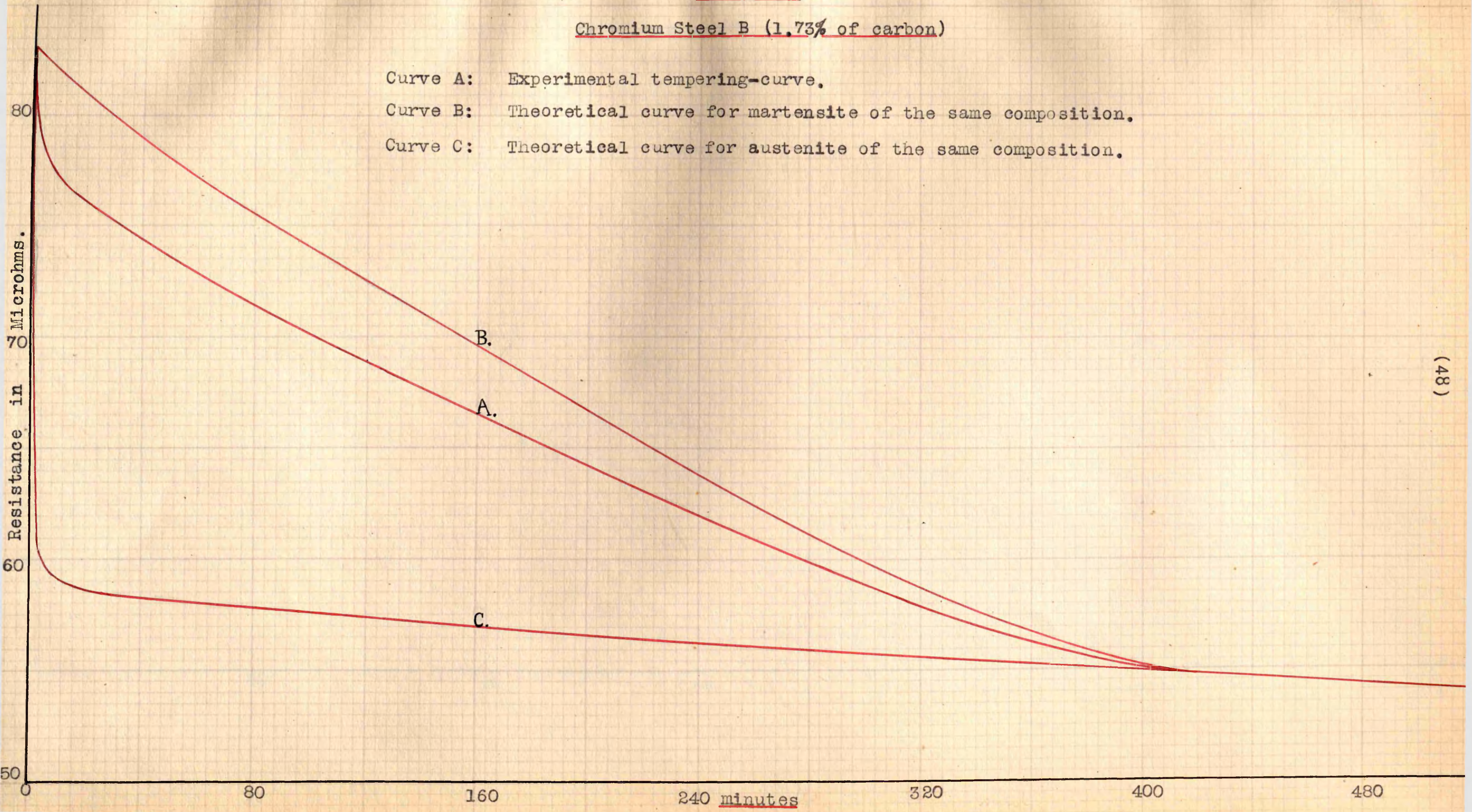


FIGURE 48.

Chromium Steel B (1.73% of carbon)

Curve A: Experimental tempering-curve.
Curve B: Theoretical curve for martensite of the same composition.
Curve C: Theoretical curve for austenite of the same composition.

FIGURE 49.

Nickel-Chromium Steel G1.

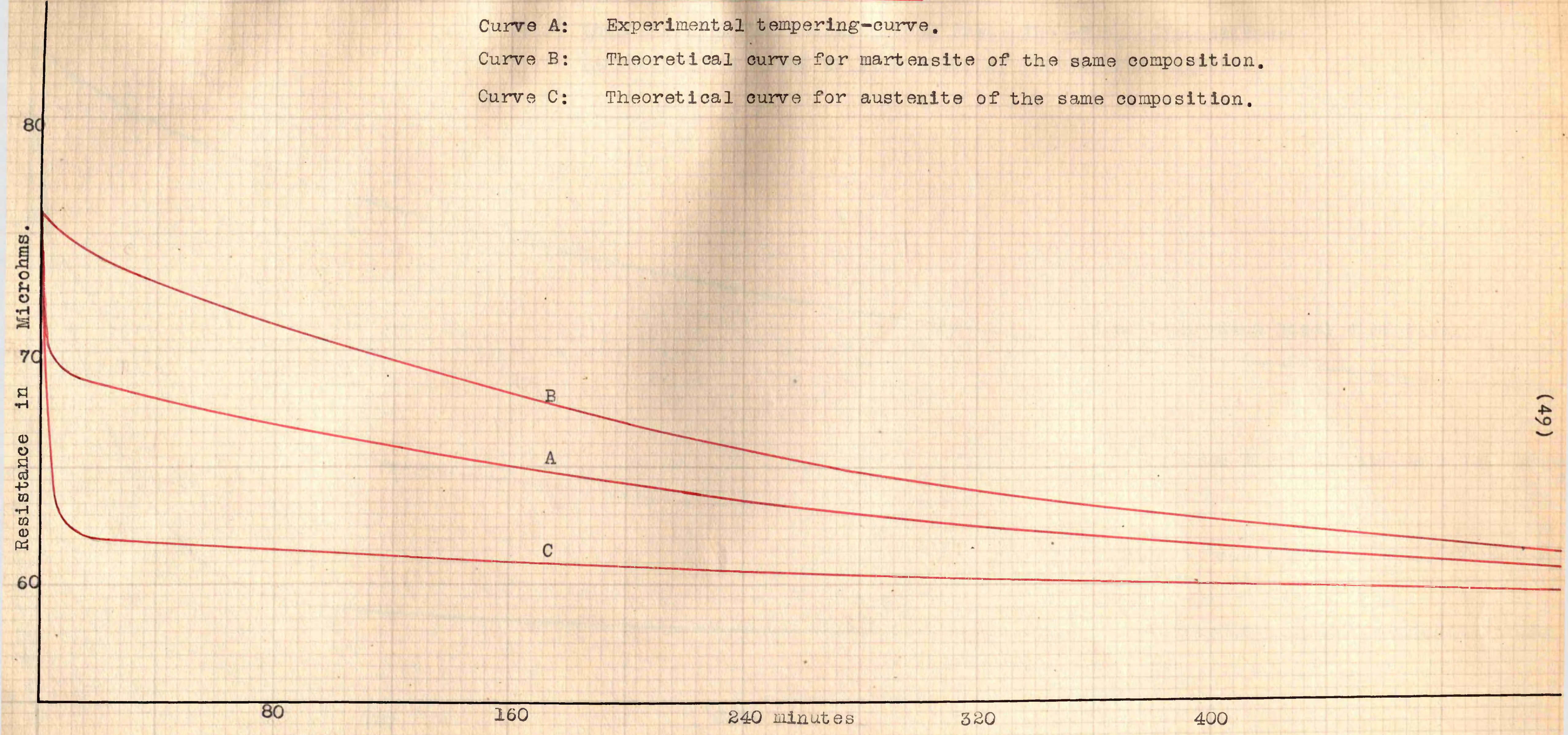


FIGURE 50.

Rate at which the Resistance of Austenite of the Compositions indicated is reduced by separation of Carbon from Solution.

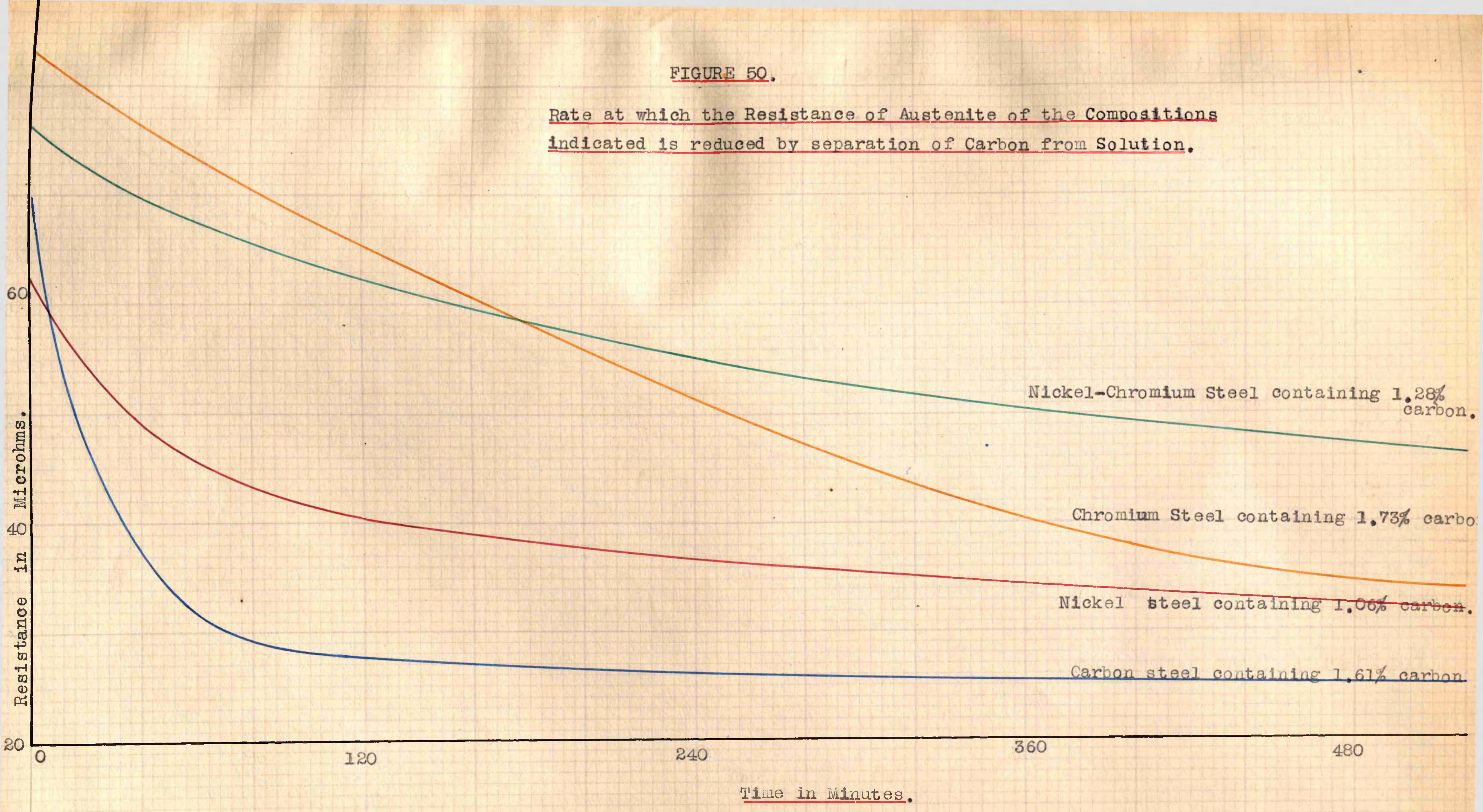


FIGURE 51.

Rate at which Carbon separates from Solution in Austenitic Steels.

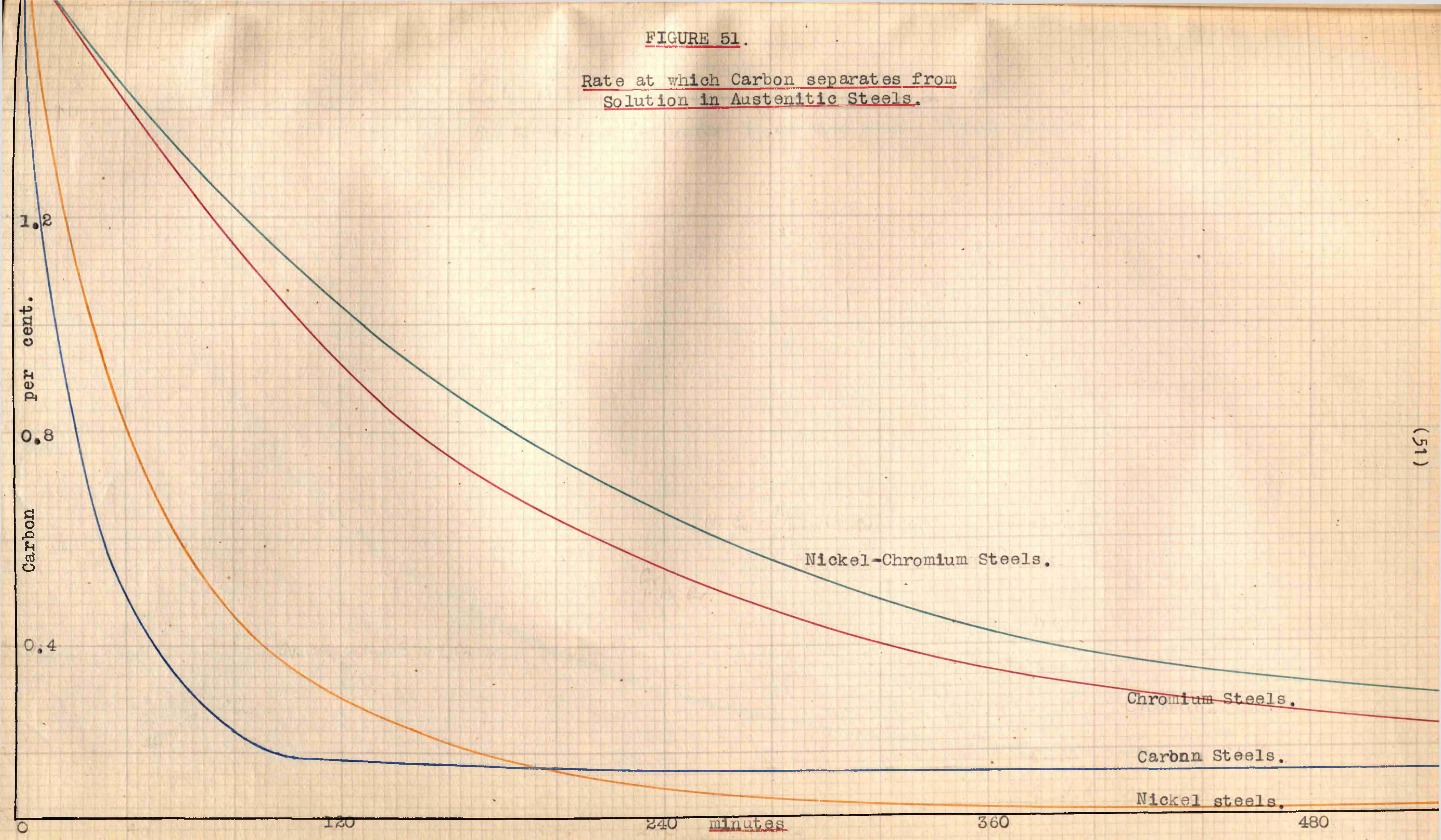


FIGURE 52.

Comparison of rate of Separation of Carbon from Martensite with the rate of separation from Austenite.

Red Curve (separation from martensite in a carbon steel) time scale: 1" = 1 minute.

Green Curve (separation from Austenite in carbon steel No.6) time scale: 1"= 10 minutes.

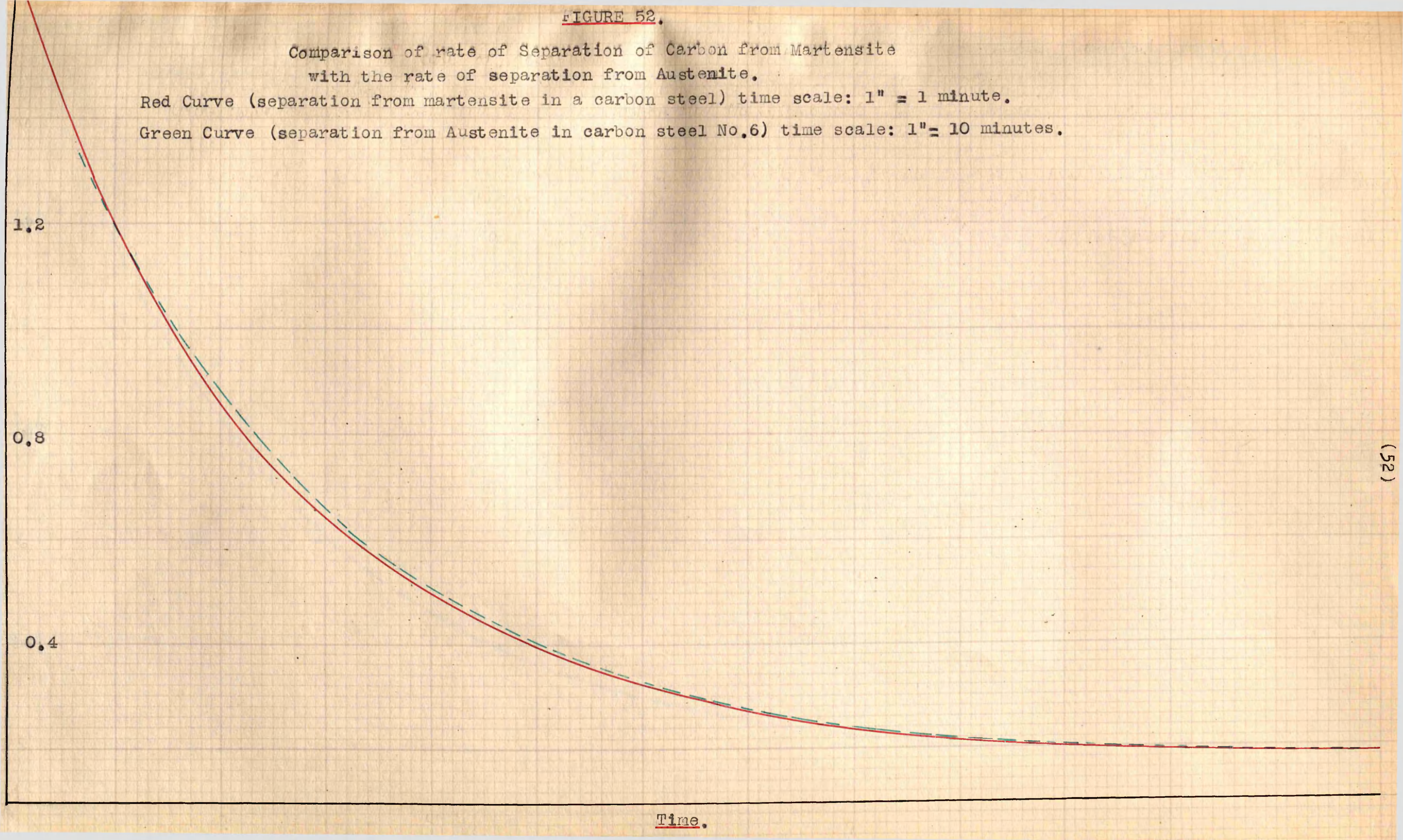
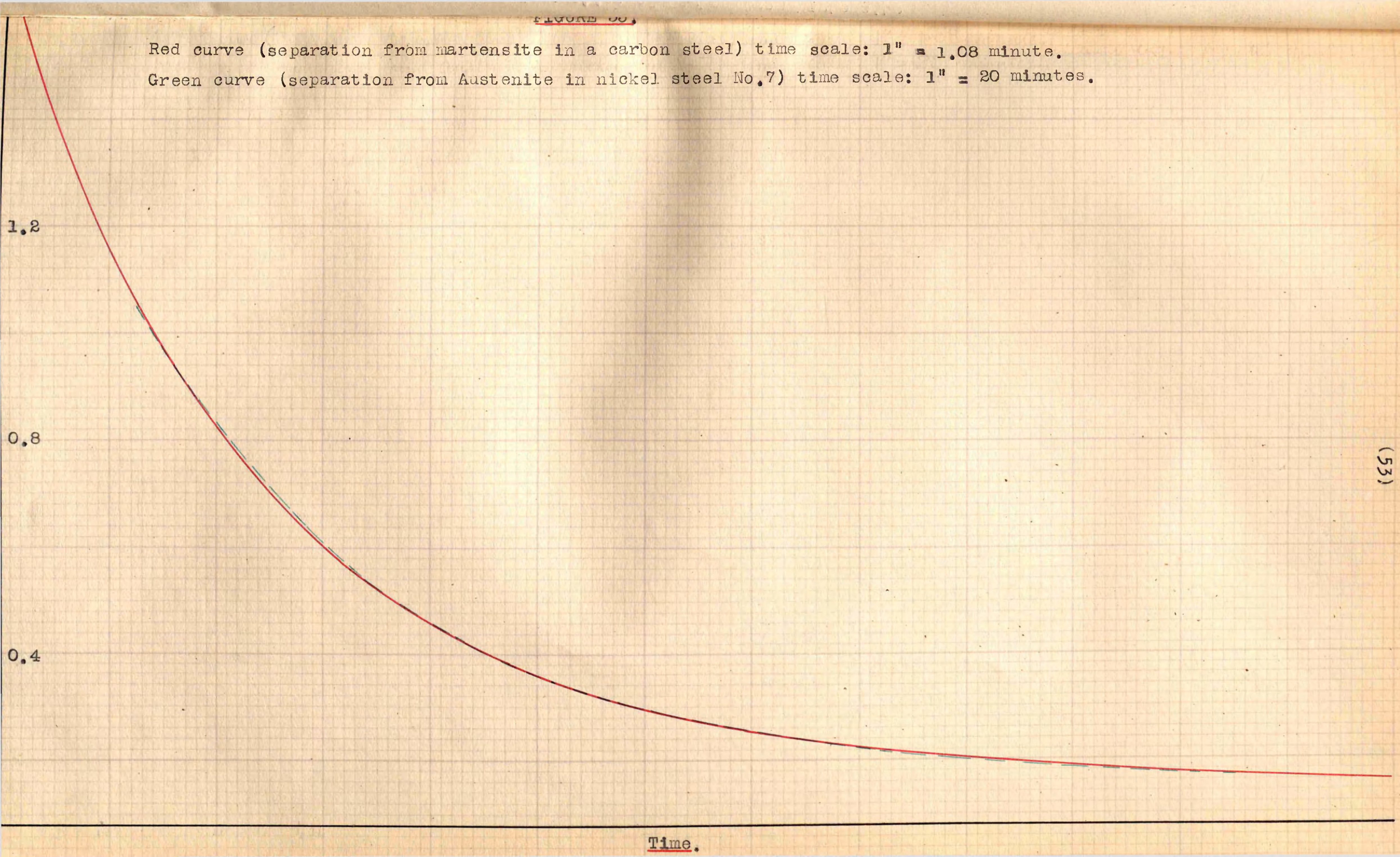


FIGURE [illegible].



Red curve (separation from martensite in a carbon steel) time scale: 1" = 1.08 minute.

Green curve (separation from Austenite in nickel steel No.7) time scale: 1" = 20 minutes.

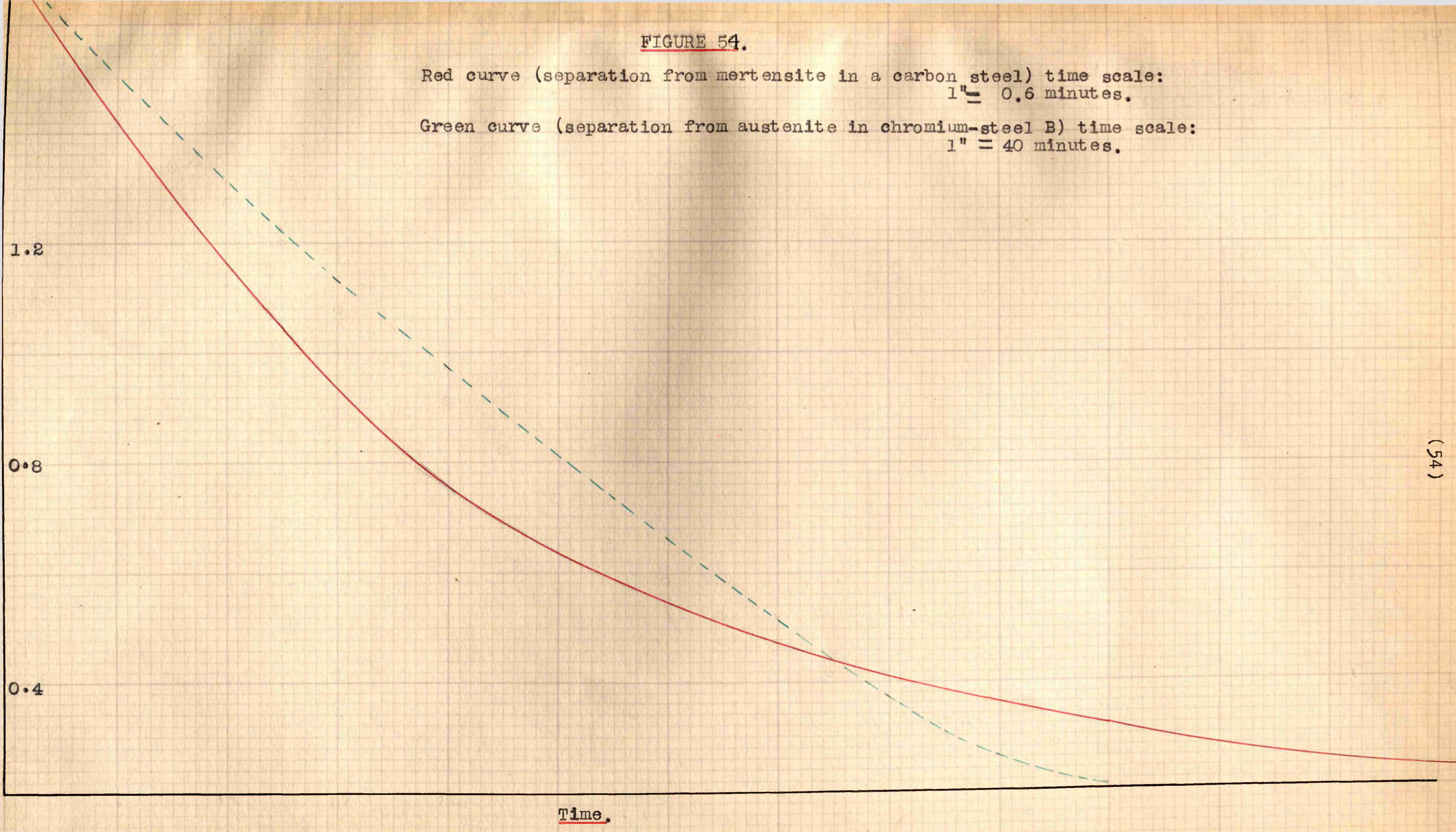


FIGURE 54.

Red curve (separation from mertensite in a carbon steel) time scale:
1" = 0.6 minutes.

Green curve (separation from austenite in chromium-steel B) time scale:
1" = 40 minutes.

FIGURE 55.

Red Curve (separation from martensite in a carbon steel) Time Scale: 1" = 0·5 minute.

Green Curve (separation from austenite in nickel-chromium steel G1): Time Scale: 1" = 40 mins.