

MAGNETIC PROPERTIES OF IRON-CARBON AND IRON-SILICON ALLOYS.

Professor Dr. E. Gumlich, Member of the Physikalisch-Technische Reichsanstalt, Charlottenburg, then read his Paper on "Magnetic Properties of Iron-Carbon and Iron-Silicon Alloys," with Micrographic Investigation and Reproductions by **Professor D. Ing. P. Goerens**, Docent of the Technische Hochschule, Aachen.

(Communication from the Physikalisch-Technische Reichsanstalt and the Eisenhüttenmännisches Institut of the Technische Hochschule, Aachen.)

At the suggestion of German technologists the Physikalisch-Technische Reichsanstalt (P.T.R.) began some years ago to examine the connection between the magnetic properties of iron alloys and their chemical compositions and thermal treatments. The Verband Deutscher Elektrotechniker supported these extensive and expensive researches by the grant of a considerable sum of money, as did also some large ironworks by supplying the necessary samples and chemical analyses. Especially grateful is the P.T.R. to Professor Wüst, Director of the Eisenhüttenmännisches Institut of the Technische Hochschule at Aachen, and to Professor Goerens, for much skilful advice and for microphotographic reproductions of the samples magnetically examined at the P.T.R. The P.T.R. is much obliged for this varied and valuable assistance, without which the accomplishment of the work would hardly have been possible. The results of the experiments have not yet been published ; it was thought best to wait until a definite conclusion had been reached, and that may extend over some years. But because the discussion arranged by the Faraday Society is partly intended to elucidate the same questions as the researches of the P.T.R., the P.T.R. decided at once to place at the disposal of the public those researches which could be of importance to the actual discussion. But we must add that those experiments have not as yet been carried to a conclusion, and that later on some results may have to be corrected more or less, principally on account of the foreign substances included in the samples, the effect of which only then can be taken into exact consideration, when the influence of larger quantities of those substances on the magnetism of the iron will be examined.

I will first refer to the examination of a series of iron-carbon alloys.

In the first analytical table the samples are arranged according to the increasing percentages of carbon. The chemical analysis was made at first by the works which supplied the samples, but was afterwards repeated by the Hoesch Ironworks at Dortmund. The differences between the two analyses were not great in general ; average values are given in Table I. The diagrams, which show the connection between the magnetic properties and the percentage of carbon, are partly attached to the analyses of the works supplying the samples only, but that is of no consequence.

In order to distinguish the samples, those supplied by the Electro-steel-works R. Lindenberg at Remscheid-Hasten are indicated by a point (.) ; the samples of the Steelworks Phönix, Ruhrort, by a St. Andrew's cross (×) ; those of the Steelworks Phönix, Hörde, by a St. George's cross (+). The two first specimens are taken from commercial products, therefore the admixtures of silicon and chiefly of manganese are undesirably large. Only the samples (+) were especially made for the purpose in question of much greater purity. The quantity of phosphorus and sulphur amounts in all samples to a few hundredths per cent. only ; therefore they have not been especially noticed in the analysis.

The magnetic measurements were, in field up to 300 gauss, conducted on rods, 6 mm. in thickness and 18 cm. in length, with the Hopkinson yoke and the ballistic galvanometer. The necessary shearing was approximately known from earlier researches, and was in each case more exactly determined by measuring not only the apparent coercive force yielded by the yoke, but

TABLE I.

Chemical Analysis.

Notation..	Sign.	C.	Mn.	Si.
		Per Cent.	Per Cent.	Per Cent.
C 6	×	0·07	0·48	0·10
C 11	.	0·11	0·25	0·16
C 16	.	0·16	0·35	0·19
C 21	×	0·21	0·52	0·11
C 23	+	0·23	0·18	0·04
C 44	+	0·44	0·13	0·06
C 51	×	0·48	0·52	0·12
C 69	+	0·69	0·13	0·16
C 71	.	0·71	0·29	0·21
C 73	×	0·71	0·51	0·14
C 107	.	0·77	0·26	0·16
C 99	.	0·99	0·25	0·24
C 114	+	1·11	0·13	0·10
C 156	.	1·57	0·37	0·23
C 180	+	1·78	0·17	0·10

also the true coercive force by the magnetometer. That can likewise be quickly and surely done for rods of any shape, and thus a constant is obtained which is extremely important for the characterisation of the magnetic material.

Measurements in strong magnetic fields were made by the isthmus method, which we owe to your ingenious countryman Ewing. In the original arrangement there is placed between the poles of a strong electromagnet a double cone of the material to be examined ; the cone can be turned, and the points are joined by a short and thin rod, the isthmus. The induction is high in the isthmus, and its intensity depends upon the dimensions of the magnet, the form of the cone, the length and the diameter of the isthmus, and so forth. The isthmus is wound with two layers of a fine wire of the same number of turns, the inner of which closely surrounds the isthmus, while the outer is slightly separated from it. If the *inner* coil be joined to a ballistic galvanometer, and the cone be turned through 180 degrees, the deflection of the galvanometer is proportional to the induction in the isthmus. If on the

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other hand the two coils be joined in opposition to one another, the deflection of the galvanometer will be proportional to the flux of lines of magnetic force in the annular space between the two coils, which answers to the magnetic field there existing. The two throws of the galvanometer are therefore measuring at the same time the induction and the field belonging to it, if the latter may be regarded as identical for the isthmus and for the annular space.

For practical purposes, this method, with which Ewing made some very interesting measurements in magnetic fields of 24,000 gauss maximum, has certain inconveniences. The principal of these consist in the fact that for each new measurement a complete cone must be prepared; further, the area of the turns of the coils, which considerably influence the results of the measurements, can only be deduced from the dimensions and therefore not very correctly be determined. Lastly, the method in that form can only be applied to relatively strong fields; because, owing to the small number of

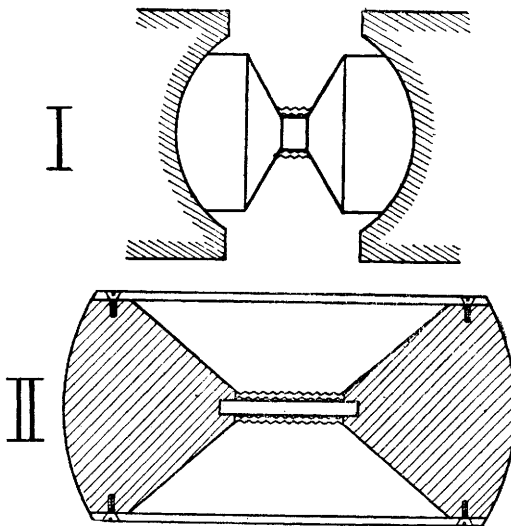


FIG. 1.

turns, when measuring in weak fields the sensibility of the ballistic galvanometers is not sufficient; on that account the isthmus measurements in strong fields cannot be made to follow directly on the yoke measurements in weak fields, although it is desirable that the measurements should be continuous. These difficulties were overcome by applying always the same cone of soft iron, and by making only the isthmus piece out of the material to be examined. The latter was made 3 mm. thick and 28 mm. long, and was inserted on both sides 4 mm. into the ends of the hollowed-out cone. The two coils, occupying the whole free space of 20 mm., were each wound in two layers on a separate tube which slipped over the isthmus, and they were cemented together; they can therefore also always be exchanged. In consequence of the large number of the turns their area can very well be determined by means of a magnetic field of known strength; lastly, for the same reason, the measurements can be begun in weak fields of 100 or 150 gauss, so that direct continuation with the yoke measurements is secured. The apparatus described allows us to attain with the small model of the

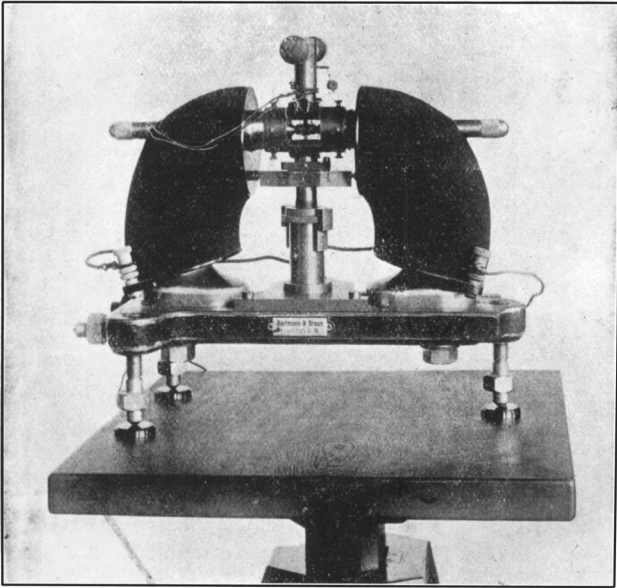


FIG. 2.

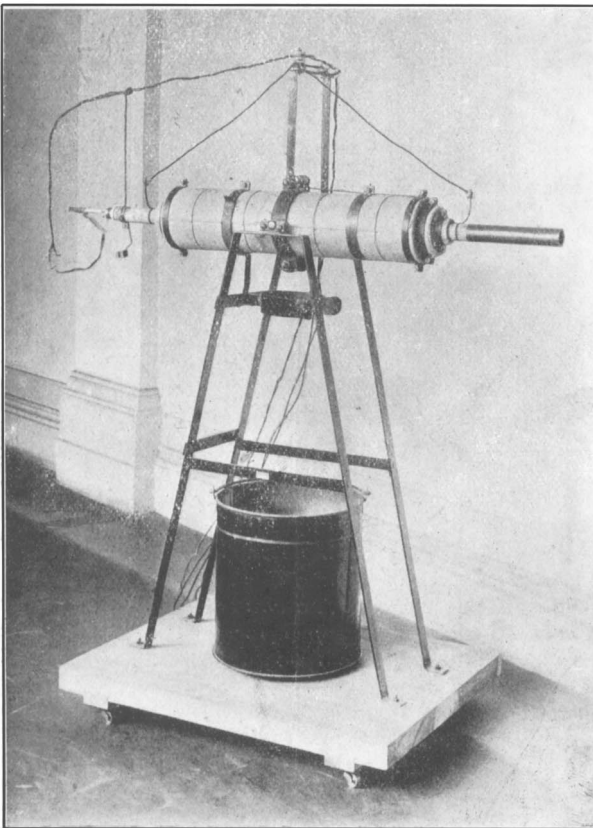


FIG. 3.

du Bois electromagnet a field of about 6,500 gauss, which is mostly quite sufficient for the measurement of the saturation; only in some extremely hard steels the limit of the saturation has not yet been reached. Fig. 2 is a diagram of the whole apparatus.

Besides the magnetic properties, the electric resistance w of the samples was also measured by sending an electrical current i through the rod and determining the potential difference iw between two edges placed on the rod by the compensation method.

As regards the treatment of the samples, one series of the rods was heated for a short time in a vacuum to 930° and slowly cooled, so that the carbon present would be surely secreted in the form of pearlite or of cementite. Other series were heated to 750° , 800° , 850° , 900° , 950° , $1,000^\circ$, and $1,100^\circ$, and

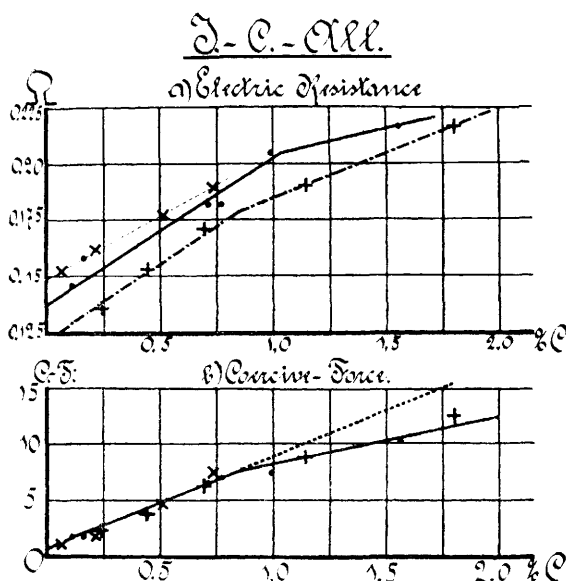


FIG. 4.

hardened in ice-water. Not to be obliged to make a whole new series of rods for every hardening temperature, however, some series were hardened successively at different temperatures. That process is indeed not quite free from objections, because it is not certain that the content of carbon is not diminished a little by the repeated processes of hardening. In that manner, for example, the fact would be explained that by hardening a rod three times at the same temperature the coercive force sank continually from 18.8 to 14.1 gauss.

For hardening the rods an electrically heated furnace was used, a view of which is given in Fig. 3. It is composed of several concentric tubes, on one of which is wound the heating bobbin of platinum foil, while the outer tubes provide for a good distribution of the heat. The rod in the middle of the furnace, the temperature of which is measured by a thermocouple, falls, when the furnace is tilted about its horizontal axis, in a fraction of a second into a vessel filled with ice-water, where it is immediately caught by a stirring apparatus and is quickly cooled.

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THE RESULTS OF THE MEASUREMENTS OF FERRO-CARBON ALLOYS.

The electric resistance of the slowly cooled alloys rises per 1 per cent. C by about 0.06 ohm per m. and sq. mm. The upper part of Fig. 4 illustrates the results; the abscissæ are the percentages of carbon, and the ordinates the resistances per m. and sq. mm. It is obvious that the percentage of Mn and Si strongly affects the absolute value of the resistance; for the series (x) with the largest percentage of Mn stands highest, the series (+) with the smallest impurities lowest. But taking the specimens of each series together, it is evident that both series with higher percentage of C show a bend at about 1 per cent. C. Since the carbon when present to about 1 per cent. exists in the form of pearlite, and at higher contents in the form of cementite, this bend seems to indicate that the same percentage of C in the form of cementite diminishes the electric conductivity less than in the form of

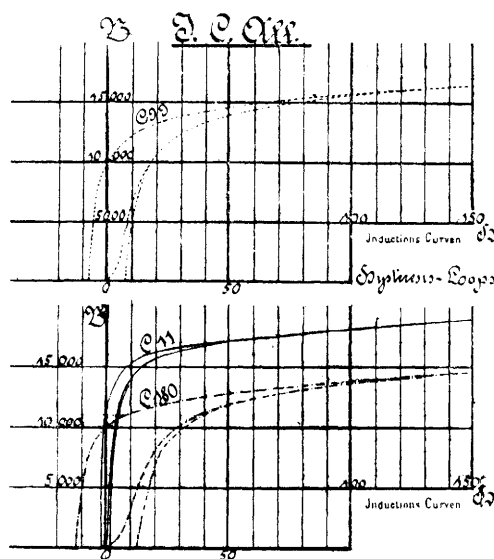


FIG. 5.

pearlite, although one constituent of the pearlite is also known to be cementite. That fact would be entirely explained by the lamellary structure of the pearlite.

On the other hand there seems to be a certain difference between the effect of under-eutectic and over-eutectic cementite. This assumption is supported by the fact that the coercive force shows the same phenomenon as the resistance, whilst it is not probable that the lamellary structure of the cementite in the pearlite would also influence the coercive force. In the lower part of Fig. 4, which represents the relation between the coercive force and the percentage of carbon, the experimental values for alloys up to about 1 per cent. C are lying on a straight line, but the points corresponding to the higher alloys are considerably lower. Thus here again we find a bend near the eutectic alloy, and it must be assumed that over-eutectic cementite causes a relatively smaller increase in the coercive force, and therefore in the magnetic hardening of the iron, than under-eutectic cementite.

The hysteresis loops of the alloys after annealing present the typical appearance ; with an increasing percentage of carbon they always become broader and lower ; thus the maximum permeability sinks from about 3,300 at 0.06 per cent. C to 450 at 1.8 per cent. C. Examples are shown in Figs. 5 and 6, illustrating the hysteresis loops and the permeability curves of the alloys with 0.11 per cent., 1 per cent., 1.8 per cent. C.

The saturation values in their dependence upon the percentage of C are represented by the upper part of Fig. 7. Disregarding the uncertainty caused by the foreign substances for which we cannot yet allow, we find nearly a straight line. The value of the saturation $4\pi I_{\max}$ diminishes by about 1,400 per 1 per cent. C ; therefore, if the value of the saturation of the pure iron is assumed to be $4\pi I_{\max} = 21,600$, that of the pearlite would be about 20,200, and that of the cementite can be calculated to be about 12,500.

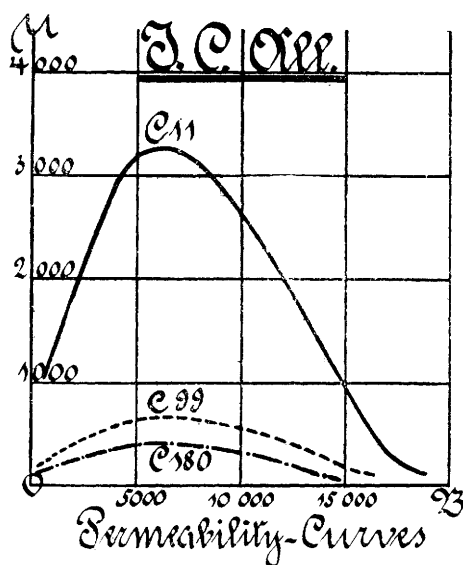


FIG. 6.

An attempt will shortly be made to obtain by direct measurements not only the values of saturation, but also the hysteresis loops of cementite.

In the lower part of Fig. 7 the saturation values of carbon alloys after hardening at 850° are indicated by the small circles. The production of good specimens is not easy, because the hardening process must be very uniformly carried out, and the conditions are very inconstant. However, it is obvious that the points observed are situated nearly on a straight line, that consequently the value of the saturation of iron diminishes approximately in proportion to the percentage of dissolved carbon, viz., by about 3,000 per 1 per cent. C ; this can, of course, be explained only by the fact that the specific gravity of the dissolved cementite cannot much differ from that of the pure iron. The last two points, for 1.57 per cent. C and 1.8 per cent. C, are situated somewhat higher, the cause of which is evidently that at 850° the contents of carbon have not entirely been dissolved yet, but a small quantity of carbon has remained in the form of cementite. The lower portion of the

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curve should have the same inclination, therefore, as the upper curve for the annealed alloy.

Corresponding measurements after hardening at $1,100^{\circ}$ have not yet given any satisfactory result.

In Fig. 8 the hysteresis loops of the same carbon alloys, examined previously after the annealing process, are shown after quenching at 850° . It is obvious how much, with increasing percentage of C, the induction diminishes in a field of 300, while the coercive force grows.

The dependence of the coercive force upon the percentage of carbon was determined for all the quenching temperatures applied. Fig. 9 gives some examples for the temperatures 800° , 900° , $1,100^{\circ}$. Very interesting is the sharp bend at about 1 per cent. C. Up to this point the coercive force rises in a straight line by about 64 gauss per 1 per cent. C; beyond that point it

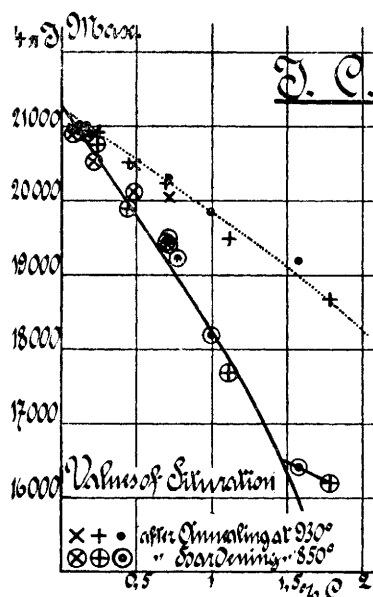


FIG. 7.

becomes nearly constant and almost independent of the contents of carbon. Thus we may say that for a hardening temperature of 800° and also 850° the coercive force rises directly in proportion to the percentage of dissolved carbon. If this conclusion be right, then the coercive force of the higher carbon alloys should also increase with rising hardening temperature, because the dissolving power of the iron for carbon grows with the increase of temperature. That is indeed the case, for in the alloy with 1.8 per cent. C the coercive force rises from 56 gauss at 750° hardening temperature to the value of 70 gauss at 950° hardening temperature, which is enormously high for a pure carbon alloy; but then it sinks again to 48 gauss after hardening at $1,100^{\circ}$. This diminution, however, appears, as the figures show, only in the higher alloys and especially in electro-steel; in the lower alloys, up to about 0.4 per cent. C, it is scarcely to be noticed. In this manner, the curve, which at a hardening temperature of 800° was composed of two straight lines, bends more and more downward. The photomicrographs of these

specimens, taken by Professor Goerens, show that this phenomenon may chiefly be traced back to the formation of austenite at high hardening temperatures and to high percentages of carbon.

In the same manner the electric resistance, like the coercive force,

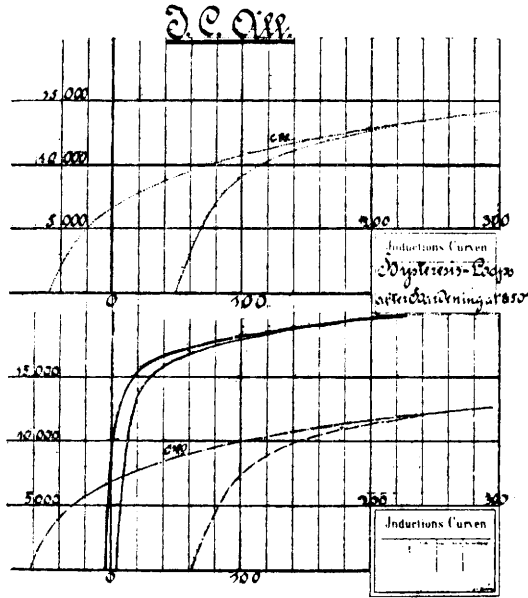


FIG. 8.

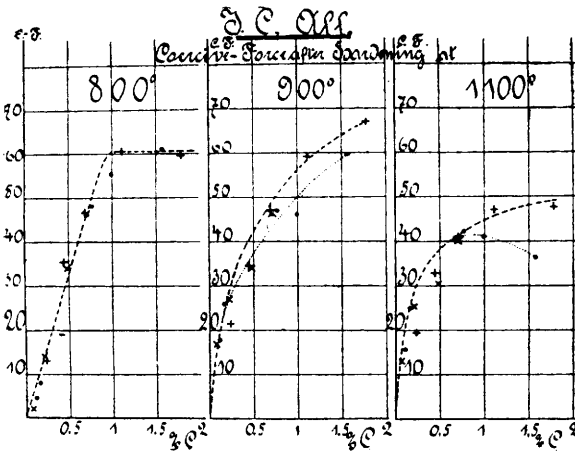


FIG. 9.

depends on the carbon *dissolved* in the iron. Fig. 10 gives the diagram for the electric resistance per m. and sq. mm. as dependent on the contents of carbon for the hardening temperature of 850°. The increase of the resistance with the growing percentage of carbon is much greater than in the case of

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annealed alloys ; that is to say, 0.35 ohm per 1 per cent. C, and is expressed by a nearly straight line. A sharp bend takes place at the same point, as in the diagram for the coercive force, and in the same manner the resistance of the highest alloys rises at higher hardening temperatures, at which more carbon is found in a state of dissolution.

Very interesting is the connection between the remanence and the contents of carbon and the hardening temperature. By remanence in this case is to be understood the *true* remanence which we observe when we examine a closed ring or a rod in the yoke ; for this remanence alone is characteristic of the material. The *apparent* remanence, on the contrary, which we observe in a magnet of the shape of a rod or a horseshoe, is much influenced by the coercive force and by the shape of the specimen, and can therefore have very different values for the same material.

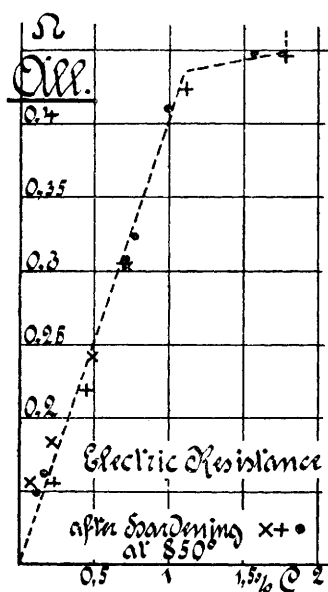


FIG. 10.

It has not so far been possible to establish a general rule for the true remanence ; we only know that the remanence can reach very high values (as high as 15,000) with pure, soft iron, while harder materials, such as steel, alloys, &c., generally have a smaller remanence. Nor has any relation between the carbon percentage and the remanence been observed in the annealed carbon alloys, the carbon of which exists in the form of cementite. But I have found it in the hardened alloys. In Fig. 11 this relation is illustrated for the hardening temperatures 800°, 900°, 1,100°, as was done before with the coercive force.

It will at once be seen that the alloys produced by the different iron-works, which also vary in their contents of Mn and Si, give separate curves ; the values of the higher alloys of electro-steel (.) especially are much lower throughout than those of the chemically purer specimens (+) ; the cause of this has not yet been found. Without counting the lowest alloys up to C 44, which appear to be in an unstable condition, it results that the remanence

diminishes nearly in proportion to the percentage of the *dissolved* carbon. The parts of the diagram between 0.44 per cent. C and 1.14 per cent. C which refer to all hardening temperatures between 850° and 1,100°, and not only those here recorded, agree indeed so closely that the conclusion seems to

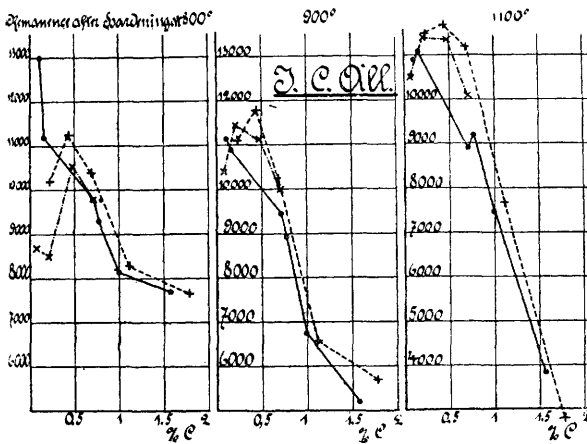


FIG. 11.

be justified that within this range there corresponds to certain contents of carbon also a certain remanence; the hardening temperature is here of no consequence. In the higher alloys, with 1.56 per cent. and 1.8 per cent. C, the more carbon will, of course, be dissolved in the iron the higher the hardening

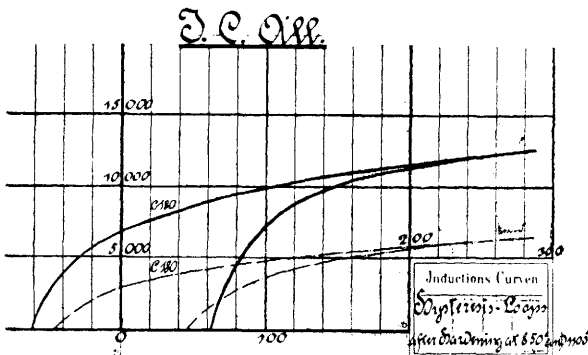


FIG. 12.

temperature, and indeed both alloys finally, at a hardening temperature of 1,100°, come down to the exceedingly low values of 3,800 and 2,800.

Here also all the hysteresis loops become extremely low, as Fig. 12 shows; it represents the hysteresis loop after hardening at 850° and 1,100° for the alloy containing 1.8 per cent. C. That again seems to be an effect of the austenite, the magnetic properties of which are still unknown or uncertain.

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If now, as has just been shown, with an increasing percentage of dissolved carbon the coercive force increases, whilst the remanence decreases, there is no possibility, at least for pure carbon alloys, of joining high remanence and high coercive force, as we desire to do in permanent magnets. An attempt must hence be made at least to unite the relatively most advantageous conditions; that is to say, as a rule the hardening temperature should neither be taken below 800° nor above 950° , and the percentage of carbon should be adapted to the shape of the magnet. If the permanent magnet to be produced is to receive the form of a straight rod or of a nearly closed ring or of a horseshoe with a very narrow gap, the best results would be given by relatively low carbon contents, about 0.5 per cent.; for in that case the most important matter is a high *real* remanence, whilst the demagnetising effect of the ends will be rather small and in consequence the amount of coercive force not very essential. If, on the contrary, it is a question of a short, stout rod or a horseshoe magnet with a broad pole-gap, &c., a high percentage of carbon is to be preferred, which will indeed give a small real remanence only, but a large coercive force, so that the demagnetising effect of the ends will be relatively small.

Whether and in how far the addition of other elements often used in the manufacturing of permanent magnets, such as tungsten, molybdenum, chromium, &c., will influence these rules, must be reserved for a later investigation.

The photomicrographs reproduced here (Figs. 13 to 19) generally confirm the deductions from the physical researches. We see first some carbon alloys after hardening at 750° (Fig. 13): alloy with 0.11 per cent. C; the light (clear) part is ferrite, the dark martensite, as will be recognised in more highly magnified reproductions. With an increasing percentage of carbon, the area of the martensite also increases, as is seen from Fig. 14 (C 23) with 0.23 per cent. C, and Fig. 15 with 0.51 per cent. C. In Fig. 16 (C 114) with 1.14 per cent. C the martensite appears already surrounded by cementite, and more so still in Fig. 17 (C 156) with 1.56 per cent. C. Here the martensite has a very delicate structure. The coercive force of that material was comparatively high, viz., 58 gauss. After hardening at $1,000^{\circ}$ this material shows a much coarser structure (Fig. 18); the coercive force had decreased to 43 gauss. Finally, after hardening at $1,100^{\circ}$ (Fig. 19), the coercive force is not more than 36 gauss; the structure presents the typical appearance of a steel with a certain quantity of austenite, which seems to be the reason of the much worse magnetic properties.

FERRO-SILICON ALLOYS.

When in 1900 Messrs. Barrett, Brown, and Hadfield published the results of their interesting researches on the magnetic properties and the electric resistance of iron-aluminium alloys and iron-silicon alloys, I had the idea to utilise the high specific resistance of the silicon alloys for the diminution of the eddy currents in transformer and dynamo sheet metal. The P.T.R. therefore requested some prominent German firms to produce transformer sheets out of silicon alloys, and began itself to make experiments with the new material. It resulted in the course of these experiments that more had been attained than had been expected; for not only were the eddy currents weakened in accordance with the higher specific resistance, but also the hysteresis loss was often smaller, and the permeability in low fields was higher than in the usual dynamo iron. Thus the so-called "legierte Blech" could not fail almost entirely to replace the usual material within

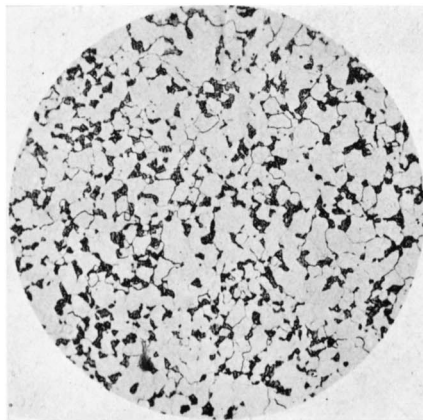


FIG. 13.

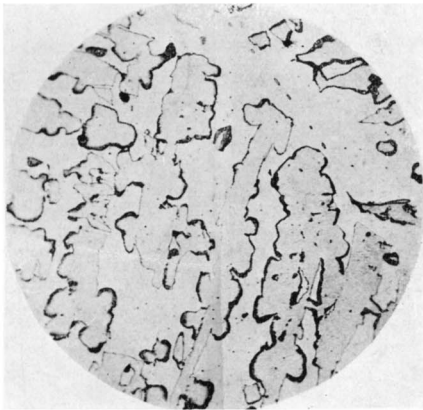


FIG. 14.

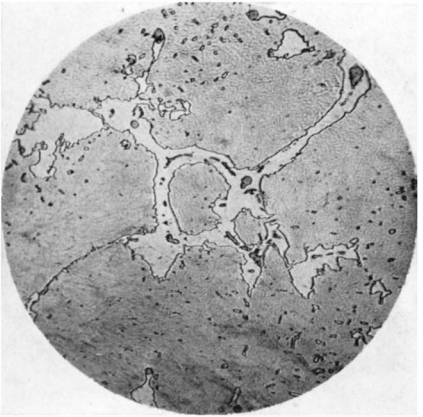


FIG. 15.

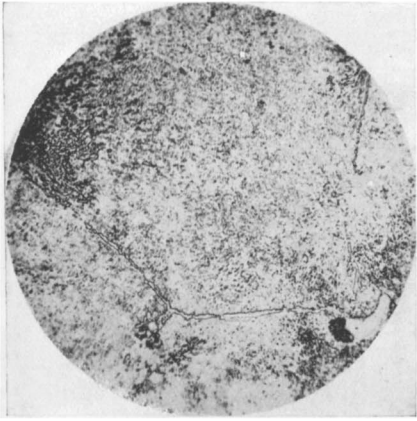


FIG. 10.

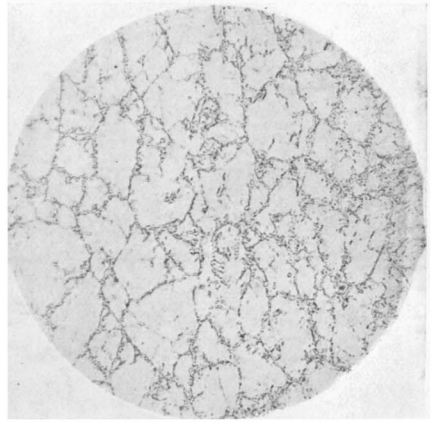


FIG. 17.

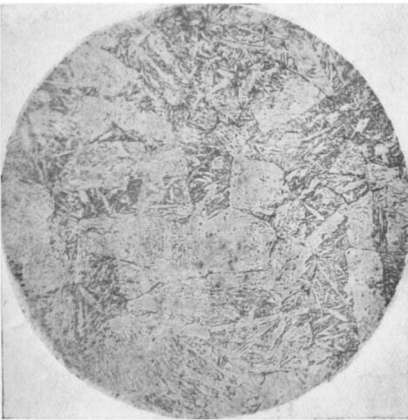


FIG. 18.

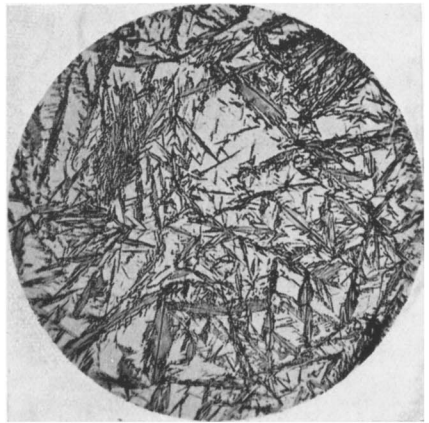


FIG. 19.

a short time, in spite of the initial difficulty of production and of the much larger price, and German electrical engineers are much indebted to Messrs. Barrett, Brown, and Hadfield for the researches which supplied the foundation for this great improvement in transformer material.

The question in which way the addition of silicon may improve the magnetic properties of iron still awaits explanation. Let me outline the opinion to which the experiments conducted in the P.T.R. have led me.

I do not believe, in the first instance, that the silicon has a *directly* improving influence on the iron. If it were so, the magnetic properties of the alloys must improve in *proportion* to the increasing contents of silicon ; that is, the coercive force must decrease, the maximum permeability and the value of saturation must rise ; but that is not so. In the following figures examples are given of the results of the measurements made in the P.T.R.

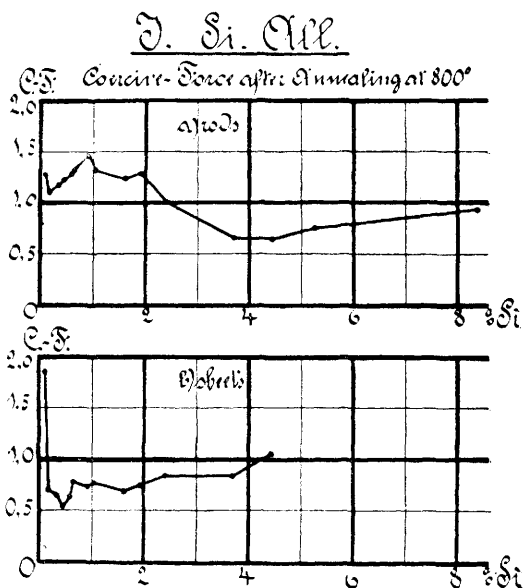


FIG. 20.

Fig. 20 gives the dependence of the coercive force upon the percentage of silicon after annealing at 800°, the upper part referring to rods containing up to 8.6 per cent. of silicon, the lower part sheets with up to 4.5 per cent. Quite analogous are the results (not here reproduced) obtained after annealing at nearly 1,000°. The observed values of the coercive force are joined by straight lines. The material of the rods and sheets is the same ; it was kindly supplied to the P.T.R. by the firm of Krupp at Essen and rolled out by the firm of Capito and Klein at Benrath.

The first point concerns the coercive force of the original material containing about 0.06 per cent. Si only ; the coercive force is in both cases rather large. In the rods as well as in the sheets a very great improvement is occasioned by the addition of about 0.1 per cent. Si. This increase continues in the sheets ; the lowest value of 0.54 gauss is reached already with 0.4 per cent. Si ; by higher percentages the material is again deteriorated.

The material of the rods from 0.15 per cent. Si upwards does not improve

at first with a rising Si percentage, but deteriorates ; with 1 per cent. an improvement sets in which continues up to about 4·5 per cent. Si ; with a larger percentage of silicon the coercive force rises again. Higher annealing temperatures are rather more advantageous for high-grade alloys, but the difference is of no consequence.

The diagrams show rather clearly that the dependence of the coercive force upon the percentage of silicon is very irregular.

This is seen still more distinctly in the maximum permeability $\mu_{\max.}$, the variation of which is represented for rods and sheets in Fig. 21, after annealing at 800°. In the rods the difference is not very great ; the values certainly are relatively high for a high percentage of silicon, but just in the richest alloy the maximum permeability decreases again. The examination of the sheets leads to very irregular results. Here the highest value of the maximum

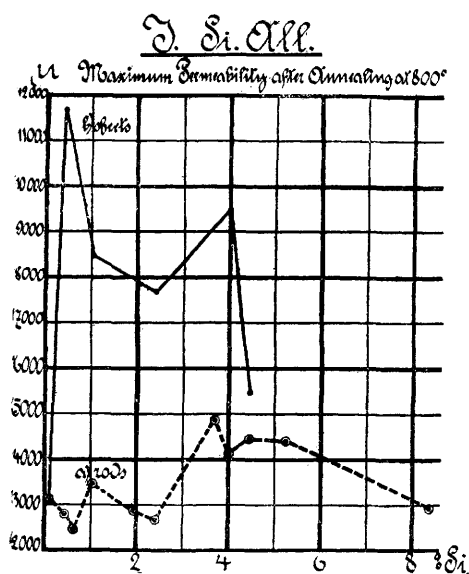


FIG. 21.

permeability occurs in the alloy with only 0·4 per cent. Si ; with an increasing percentage the values even decrease considerably.

But the saturation values, $4\pi I_{\max.}$, the variations of which are shown in diagram (Fig. 22), bear the strongest evidence against the supposition that silicon has a directly improving influence. These saturation values do not increase with the growth of the percentage of silicon, but they decrease, and the change takes place along a curve, the upper portion of which, up to about 4·5 per cent. Si, may approximately be considered as a straight line ; the decrease amounts to about 500 maxwells per 1 per cent. Si. It may therefore be said : The silicon is acting like a foreign substance, which diminishes the active cross-section of the iron, and diminishes in consequence the value of the saturation. Messrs. Hopkinson and Hadfield came to the same result in the account of their researches published last year, "The Magnetic Properties of Iron and its Alloys in Intense Fields." *

* *Journal of the Institution of Electrical Engineers*, 46, 235-309, 1911.

It may then be taken for granted that the effect of silicon is not a direct, but an indirect one, and the question arises, of what kind is this indirect effect? Here also Messrs. Hopkinson and Hadfield have already suggested an explanation, saying that the silicon seems to neutralise to a certain degree the effect of the carbon on the value of saturation. I arrived at the same conclusion some years ago already on the strength of numerous measurements. Indeed, I believe that this influence is much more effective still in the region of the low inductions than in that of the high ones, which Messrs. Hopkinson and Hadfield had especially in view, and that the value of the coercive force in particular and therefore of the hysteresis loss, which is approximately in proportion to the coercive force, is very much influenced by the silicon. I would like to go a little further into this question.

I mentioned before that the increase of the coercive force by 1 per cent. C is about 7.5 gauss if the carbon exists in the form of cementite but at least

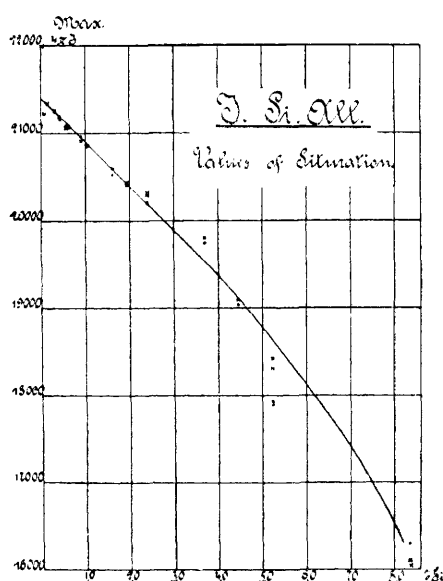


FIG. 22.

60 gauss if it is dissolved in the iron. Now there were among the silicon alloys examined in the form of rods several with relatively high contents of carbon, as may be seen from Table II.

If these amounts of carbon had been dissolved in the material, they should have produced approximately the values of the coercive force marked in the fourth column; but if in the form of cementite, the values in the fifth column, assuming that the original material (without carbon) had a coercive force of 0.7 gauss. In reality, there were found for the coercive force, before annealing at 700°, the values of column 6; after annealing at 700°, the values of column 7; after annealing at 975°, the values of column 8. And here it must be mentioned that the two first numbers, 30 and 50, were cast in the shape of stout blocks afterwards cut up; the two last, 52 and 85, on the contrary, in the form of thin rods about 1 cm. in diameter; the latter have therefore cooled rapidly, the former slowly. That is the reason why the coercive force of the two latter has been relatively high before annealing,

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especially that of number 85. Evidently a small quantity of C has remained in a state of dissolution. In the three other alloys, however, the coercive force is already less before annealing than it should be according to the contents of carbon, if this had existed in the material in the form of cementite; therefore part of the carbon must have been secreted already during the cooling process after the casting in the form of the passive temper-carbon (graphite). After annealing at 975° for twenty-four hours, the carbon seems entirely to be secreted in the form of temper-carbon, for the coercive force is, as column 8 shows, as small as that of pure iron, which scarcely contains any carbon. In the higher alloys, Nos. 50 and 52, this segregation occurs already after annealing at 700°, for the coercive force has, as column 7 shows, already gone down to very low values.

On the other hand, the coercive force of the *low-grade* Si alloys in the form of rods is still very high, about 1·2 to 1·5 gauss, and is only slightly influenced also by the height of the annealing temperature, whilst it decreases in the same alloys to about half its value, if these are rolled to sheets. The enquiry into the explanation of the fact that the magnetic properties of low-grade alloys in the shape of sheets are generally much better than those of rods, is not yet finished.

TABLE II.

1	2	3	4	5	6	7	8
Specimen Number.	Percentage of		Coercive Force.				
			C dissolved.	C in form of Cementite.	Before Annealing.	After Annealing	
	C	Si.				At 700°.	At 975°.
30	0·21	2·4	13	2·3	1·30	1·19	0·70
50	0·29	4·5	18	2·9	1·26	0·65	0·69
52	0·18	5·25	11	2·1	1·93	0·70	0·59
85	0·34	8·35	20	3·2	6·56	4·60	0·70

From the same point of view the facts can be explained which are observed in hardening silicon alloys with a considerable percentage of carbon. Even when the material had not been annealed before, the coercive force after quenching is not nearly so high as in the pure carbon steel of the same carbon percentage. For instance, an alloy with 4·4 per cent. Si and 0·29 per cent. C, after quenching at 950°, yielded a coercive force of 6 gauss, whilst a pure carbon steel of the same percentage of carbon under identical circumstances would have shown a coercive force of about 33 gauss. The effect becomes still more surprising when the material had previously been annealed for a longer time. The same material, after having been annealed for twenty-four hours at 800°, showed when hardened at 850° only a coercive force of 0·76 gauss, so that surely no hardening carbon could have been present in this material. These conclusions were confirmed by the chemical analyses made by the firm of Krupp. In all the specimens examined, which were hardened at different temperatures, the presence of a trace of hardening-carbon could only be proved in one case; in all other cases the considerable amounts of carbon consisted partly of cementite and partly of temper-carbon.

An observation made during the study of the transition-points of the iron-silicon series will finally have a certain interest in this respect. The first

transition-point was observed by the Roberts-Austen method, the second by the deflection of the magnetometer. The latter can be observed under all circumstances with sufficient exactitude, whilst the former is only then clearly to be seen when the material forms pearlite. These transition-points are shown on the diagram, Fig. 23, where the abscissæ give the percentages of silicon in the alloys and the ordinates the temperatures of the first and second transition-points. It appears striking at first that the second magnetical transition-point is lowered by an increasing percentage of silicon, whilst the first, pearlitic transition-point, the mean between Ac_1 and Ar_1 , rises even above the second. The two coincide for 2.2 per cent. Si. Up to 2.4 per cent. Si, the first pearlitic point of transition can also be well perceived, but in the higher alloys it cannot be traced at all, though the absolute percentage of carbon rises also with the increasing contents of silicon, and grows, as has already been mentioned, to about 0.3 per cent. This fact can scarcely be explained in any other way than by assuming that in these high-grade silicon alloys the possibly present pearlite is transformed into temper-carbon already during the slow heating up to about 800° .

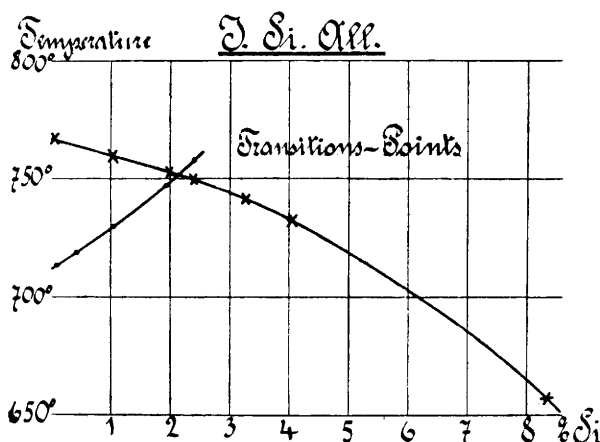


FIG. 23.

It was to be hoped that these conclusions as to the influence of Si on C would be supported by the micrographic study, and this is indeed the case, as the following examples (Figs. 24 to 30) will demonstrate. The first are photographs of untreated iron-silicon alloys enlarged a hundredfold (Fig. 24), Si 10 ; 1 per cent. Si ; 0.2 per cent. C (Fig. 25), Si 30 ; 2.4 per cent. Si, 0.2 per cent. C (Fig. 26), Si 50 ; 4.5 per cent. Si ; 0.29 per cent. C. The structure of all these alloys is clearly pearlitic, as will immediately be recognised by examining the last specimen with an enlargement of 750 diameters (Fig. 27). This pearlitic structure has been preserved almost unchanged after annealing at 700° for twenty-four hours in the low alloys and up to 2.4 per cent. Si (Fig. 28), that is, up to the highest alloy in which the pearlitic transition-point could still be distinguished, but no longer in Si 50 with 4.5 per cent. Si and 0.29 per cent. C (Fig. 29). Here the pearlite has entirely disappeared, and in its stead enclosures of temper-carbon have appeared ; but at the same time the coercive force has decreased from 1.26 to 0.65 gauss.

But the reverse takes place, even after annealing at 975° , in the original

material, which contains only very little silicon (0.06 per cent.) ; the content of pearlite has entirely remained unchanged (Fig. 30) ; as the percentage of carbon is not small (0.15 per cent.), the rather high coercive force of 1.5 gauss observed in this material even after annealing needs no further explanation.

These results seem to justify the following conclusion. The presence of larger amounts of silicon prevents, even with rather quick cooling, the formation of the extremely injurious solid solution of carbon and iron ; the contaminations with carbon appear only in the shape of the much less injurious pearlite. Even the pearlite is decomposed by a prolonged process of annealing, under the influence of the silicon, into ferrite and temper-carbon, and thus becomes magnetically almost quite passive. To produce this effect with certainty, however, there must be at least from 3 to 4 per cent. of silicon in the material. The fact that thin sheet-metal containing smaller amounts of silicon may show exceedingly good magnetic properties must have other reasons.

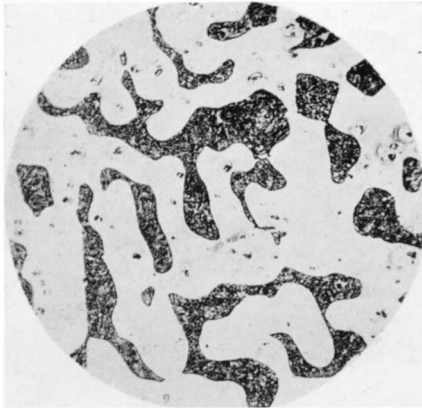


FIG. 26.

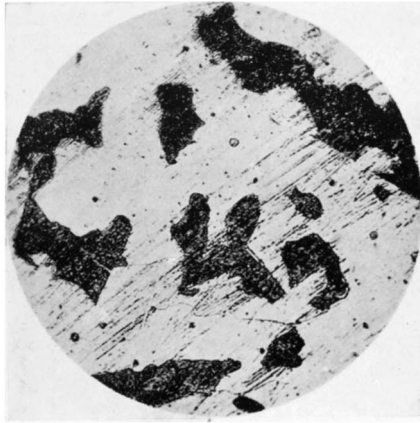


FIG. 25.

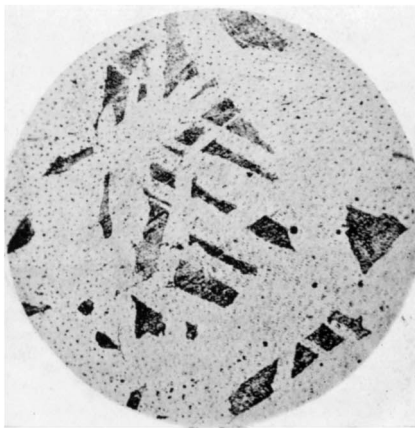


FIG. 24.

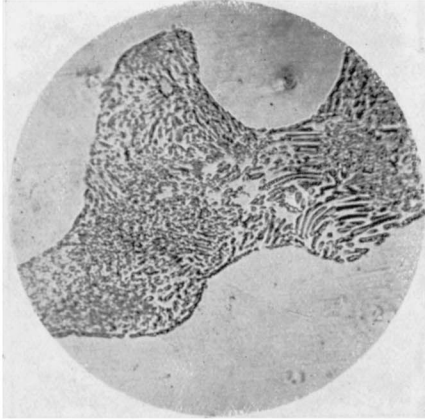


FIG. 27.

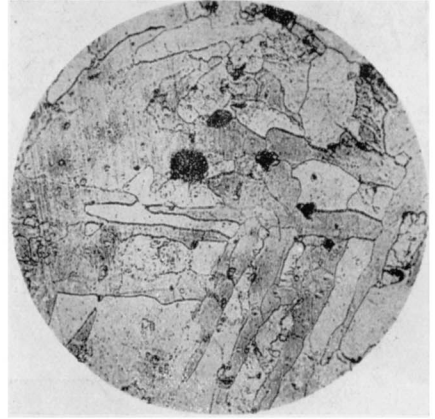


FIG. 28.

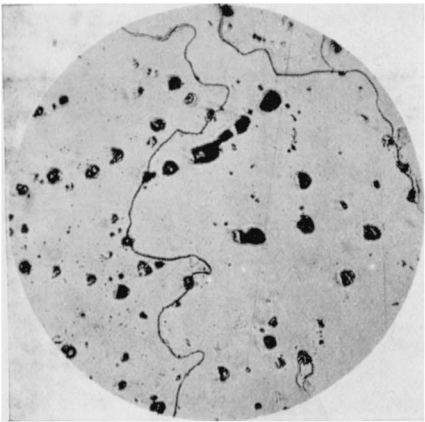


FIG. 29.

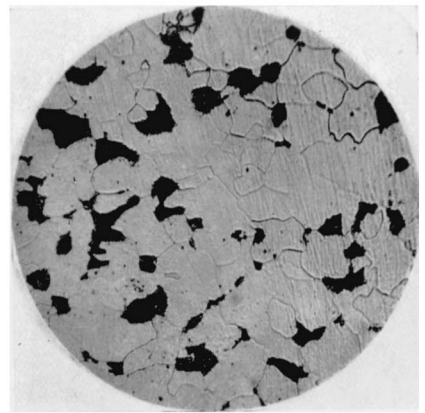


FIG. 30.