CHAPTER IX

HEAT TREATMENT AND ANNEALING

Frequently, steel castings are sold without annealing, especially when the sections are not so unequal as to cause shrinkage stresses. The impression prevails very generally that the relief of these stresses is the chief function of annealing. Though this is far from the truth, yet it is an important function.

The magnitude of the stresses set up in the shrinkage of steel can be roughly measured by means of a test bar of the shape shown in Fig. 18. By measuring the distance between two punch marks, located as shown, on the thin portion of this casting, before and after cutting through this section with a hack saw, it will be found that this distance after sawing in two will be about .013 in. less than before.
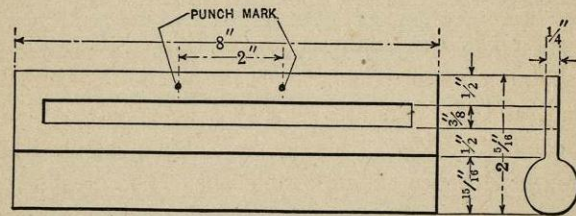

FIG. 18.—Stress-test casting.

This amount, however, is the sum of the shortening of the large section, and the lengthening of that portion of the small section lying outside the punch marks. In this case the punch marks were 2 in. apart, and the length of the casting 8 in. Therefore this sum of .013 in. equals the extension of three-fourths of the small section, plus the contraction of the large section. If we call the total extension of the small section X, then the contraction of the large section equals $.013 - \dfrac{3X}{4}$.

Now the stresses in each section in pounds are equal. The cross-sectional areas of the sections are .6902 sq. in. and .125 sq. in.

If the modulus of elasticity in compression is equal to that in tension, and amounts to 29,000,000 lb. per square inch, then:

$A = \dfrac{29,000,000}{8} =$ the force in pounds per square inch needed to stretch or compress the bar 1 in.

A times $.6902 =$ the force in pounds needed to stretch the large section 1 in.

A times $.6902$ times $\left(.013 - \dfrac{3X}{4}\right) =$ the force in pounds needed to stretch the large section $\left(.013 - \dfrac{3X}{4}\right)$ parts of an inch.

Similarly,

A times $.125$ times $X =$ the force in pounds needed to compress the small section X parts of an inch.

Then,

$$A\,(.6902)\left(.013 - \frac{3X}{4}\right) = A\,(.125)(X)$$

Whence $X = 0.014$ in.

Then,

$\dfrac{29,000,000}{8} \times (.014) = 50,750$, the stress in pounds per square inch

that was set up in the small section of this casting. This figure is of course only an approximation, as the accuracy of measurement is not great, and the bars are somewhat deflected by the stresses set up in cooling. Indeed, this figure is above the elastic limit in compression, and hence an impossible one. The great severity of this stress, however, will be easily believed when it is stated that only by cutting such a casting from its sprue can it be kept from breaking; if it is pounded with a hammer, a few blows will cause the light parts connecting the two legs to snap in two like pipe stems, and many of the castings will break apart even in cooling in the air.

These residual stresses can be relieved almost entirely by allowing the castings to cool slowly and evenly from the initial heat in a furnace—the castings either being put in at a high temperature, or got in at such temperature as can be managed, and heated up. Were the relief of stress the only, or even the chief function of annealing, therefore, it would be quite sufficient to cool the castings in pits, such for instance as are used for chilled iron car wheels, or in annealing furnaces.

We know, however, that in order to secure the greatest toughness and strength, we must heat our castings for a certain length of time to proper temperatures, and cool them from those temperatures at

an appropriate rate. This annealing could be carried out without allowing the castings to grow cold after solidifying; commonly, however, they are allowed to cool and are cleaned of adhering sand, and the sink heads and sprues removed, before annealing. The difficulty and expense of properly heating them up when heavily coated with sand, and the extra expense involved in heating sink heads, make this practice essential for proper annealing.

It is very commonly assumed that the fracture of a bar of cast steel, as viewed by the eye, is sufficient examination of the "structure" to indicate whether a given piece of steel has been annealed or not, and if the annealing has been properly carried out. It may be stated without hesitation, however, that the fracture is a most insufficient indication of the thoroughness of the annealing, since in any piece of cast steel that has been reheated above a certain temperature, the characteristic coarse crystallization of the raw steel will be found to have been replaced by a practically uniform fine grained fracture, quite regardless of the rate of cooling from the annealing temperature. The eye, therefore, is unable to distinguish in the fracture of the steel any considerable differences in internal structure, due to variations in their rate of cooling. When we come to study the internal structure of the steel, as revealed by examining under the microscope the polished surface of a test piece etched with suitable acids, we find that in two steels whose fractures as judged by the eye are identical, there may be enormous differences in structure; and tests will soon show us that these differences in *microstructure* are accompanied by, and can be used as indications of, great differences in physical properties.

The science of metallography is a study in itself, has its own most extensive literature, and can be covered here only in the barest outline, as applied to cast steels. First and foremost, it must be understood that by this term is meant not the mere examination of the steel under the microscope; the science of metallography in its broadest sense includes the physical, and all other testing of steel, and the study of the effect of chemical analysis, methods of manufacture, methods of heat treatment, even methods of moulding and pouring, upon it. The microscope is but one tool, although an important one, in the hands of the metallographist.

In preparing steel for microscopic examination, one face of the specimen is first brought to a mirror surface by polishing on emery papers or cloths of increasing fineness, and on buffing wheels coated with tripoli powder or jeweler's rouge. The polished surface is next

etched with acids, iodine, or other corrosive agents, that attack the microscopic constituents unequally, then washed and dried. The microscopic constituents of ordinary steel thus revealed are as follows:

Ferrite.—Pure iron, free from carbon; slightly colored by acids; the junctions of the ferrite crystals are brought out by etching agents, which eat into the junctions and reveal them as black lines. Carbonless steel or wrought iron is composed almost entirely of ferrite, but in wrought iron the slag fibers make up a considerable part of the section.

Cementite.—The chemical compound, Fe_3C, of iron and carbon. Very hard and brittle; not colored by acids, appears clear white; but colored yellow by sodium picrate.

Pearlite.—The eutectoid mixture of cementite and ferrite found in steel containing carbon, when cooled slowly from above the recalescence point. It is built up of thin parallel plates of ferrite and cementite alternately. Colored black by acids, when viewed under low power. Under high power, the banded appearance is seen—cementite white and ferrite black, because the ferrite is eaten away below the level of the cementite plates and does not reflect the light.

At from .80 to .90 per cent. (generally given at .89 per cent.), carbon, the entire mass of ordinary steel in the slowly cooled state consists of pearlite.

Steels between o per cent. carbon and .89 per cent. carbon are composed of ferrite and pearlite in proportion varying with the carbon. Low carbon steels are composed of ferrite with small areas of pearlite. In steel of .25 per cent. carbon the pearlite occupies about one-third of the cross-sectional area. As carbon increases, pearlite increases, the ferrite occupying less and less of the area; at about .60 per cent. carbon the ferrite appears chiefly as a net-work surrounding meshes of pearlite; as .89 per cent. carbon is approached the net-work of ferrite becomes discontinuous and the nets thin, finally disappearing.

Above .89 per cent. carbon, free cementite appears as a net-work surrounding pearlite meshes, the net-work growing more continuous and the nets thicker, as carbon increases. Many needle-shaped cementite areas intersect the pearlite meshes at various angles.

In general, the strength of the steel rises, and the brittleness increases, in direct proportion to the carbon content.

Steels below .89 per cent. carbon are called hypo-eutectoid steels.

Steels of .89 per cent. carbon are called eutectoid steels.

Steels above .89 per cent. carbon are called hyper-eutectoid steels.

Sorbite.—Pearlite that cannot be resolved under the microscope. The alternate plates of ferrite and cementite are too thin and discontinuous to be visible. Sometimes called "emulsified pearlite"—the pearlite having visible plates being called "lamellar pearlite." Typical of carbon steel that has been cooled in air from above the recalescence point, so that the austenite has been transformed too rapidly to produce lamellar pearlite.

Austenite.—The solid solution of Fe_3C, in iron; above the critical range, the entire mass of a steel exists as austenite. Cooled with great rapidity from above the critical range, as for instance, by quenching a thin piece of steel in iced brine (particularly if much carbon, nickel or manganese is present), steel consists wholly or partly of austenite.

Colored yellowish by acids; appears as uniform grains of a single substance, resembling ferrite; the crystal junctions outlined as black lines.

Martensite.—Austenite partly transformed; a transition product between austenite and sorbite. Very hard and brittle. Etched with iodine exhibits markings in acicular or arrow-head arrangement, sometimes heavily outlined in black. Characteristic of carbon steel that possesses marked hardening power, and that has been quenched from above the recalescence point.

By measuring the rate of heating or cooling of a sample of steel, it will be found that at certain well-defined temperatures, marked changes occur in this rate. These are called the "Critical Points," and correspond to changes of state of one or more constituents of the steel. That which occurs at the lowest temperature is known as the "Recalescence Point," because at that temperature, steel containing enough carbon to form a fair proportion of pearlite, evolves sufficient heat in cooling to cause its temperature to remain stationary for some moments, or even to rise slightly. This phenomenon is known as recalescence.

Two of the critical points, A_1 and A_3, occur at somewhat higher temperatures in heating than in cooling. This phenomenon, which is really the raising or lowering of the critical point from the temperature at which it should occur, is due to lag, or molecular inertia —the fact that the transformations require time to complete themselves. Hence the raising or lowering of the temperature is the greater, the more rapid the rate of heating or cooling. Could that rate be made infinitely slow, the critical points in heating and in

cooling would be at the same temperature. As it is, the temperature interval is considerable, so that we have to distinguish between the two points by the symbol Ac for heating and Ar for cooling. Normally Ar_1, for instance, occurs at 680 or 690° C., and Ac_1 at 720 to 740°.

It is this dragging down of the transformation points by rapid cooling, assisted by the influence of carbon, manganese, nickel and some other alloys, that gives steel its hardening power. Very rapid cooling so far depresses the temperatures at which the transformations take place that they are wholly or partially suppressed, giving us austenite in the former case, and martensite or sorbite, in the latter. This discussion omits troostite and some other transition constituents that are not of great importance in the heat treatment of steel castings; and omits also some other means by which the transition constituents can be obtained, as for instance, quenching steel heated above the recalescence point, but below the other critical points.

The Carbon-iron Diagram.—The carbon-iron diagram (Fig. 19) shows the temperatures at which steels of varying carbon content solidify, and the temperatures at which their critical points occur, plotting temperatures against carbon per cent. With all that portion of the diagram lying to the right of the point, carbon equal 1.75 per cent. we shall not concern ourselves in this discussion, as the very high carbon steels and cast irons are of little interest to the steel foundryman.

Referring to the diagram, the line PSK is the recalescence point A_1 at which pearlite is formed on cooling, and changes to austenite on heating (called Ac_1, heating, Ar_1, cooling).

MO is the point A_2 (Ac_2 heating, Ar_2 cooling), at which alpha ferrite becomes beta ferrite on heating, and *vice versa*.

GOS is the point A_3, (Ac_3 heating, Ar_3 cooling), at which beta ferrite is liberated from austenite on cooling, or changes to gamma ferrite and dissolves in austenite on heating.

It will be seen that above about .35 per cent. carbon, A_3 falls until A_2 and A_3 unite, becoming a single point A_{2-3}.

A_{2-3} continues to fall as .89 per cent. carbon is approached, until at .89 per cent. carbon, the transformations all take place at the recalescence temperature A_1, giving us a single point A_{1-2-3}.

SE represents the temperature at which free cementite in steel above .89 per cent. carbon is dissolved in austenite on heating, and is set free on cooling.

15

In crossing the line *MOSK* in heating, iron loses the bulk of its magnetism; and, of course, recovers it on recrossing the line as it cools.

At the line *AB*, solidification of liquid steel takes place. The line *Aa* represents the temperature and composition of the solidfying metal, the line *AB*, that of the molten mother metal. The first freezing particles are of lower carbon content than the mass of the steel. Thus, a steel of .5 per cent. carbon begins to freeze at the point (*q*), and the composition of the first freezing layers is (*p*). As freezing proceeds, the molten mother-metal becomes progressively
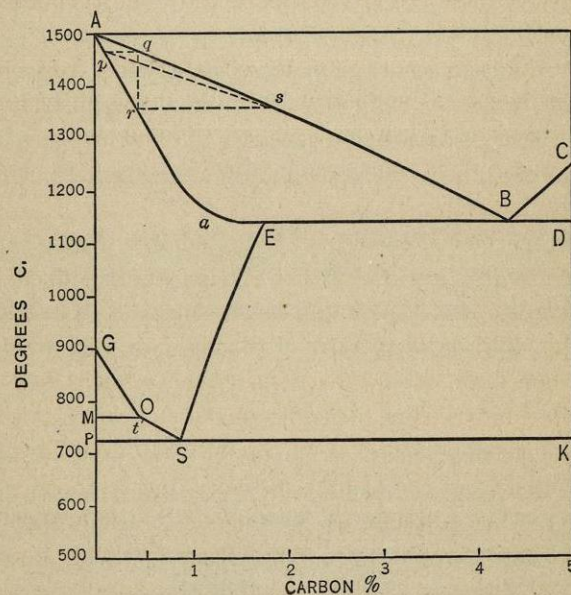
FIG. 19.—The carbon-iron diagram. From Proceedings of Sixth Congress, International Association for Testing Materials.

richer in carbon, its temperature and composition sliding along the line *AB*, and the successive particles of freezing metal also become richer in carbon, their temperature and composition sliding along the line *ps*. Meanwhile diffusion of carbon begins in the frozen portions of the mass, tending to equalize the composition of the successive layers. If this diffusion is complete, the temperature and composition of the frozen portion slide along the line *pr*.

When the temperature of the steel has fallen to *Aa*, the final particles freeze. Thus a .50 per cent. carbon steel is completely frozen at a temperature represented by (*r*); the temperature and com

position of the last freezing particles will be (*s*); that of the last particles of the molten mother-metal to freeze, also (*s*), and if diffusion has been complete in the frozen portions, the composition and temperature of these portions (*r*).

At any temperature between (*q*) and (*r*), on the line (*qr*), the amount of solid is proportional to the horizontal distance between this point and the line *AB*, the amount of liquid by the horizontal distance from this point to the line *Aa*. Thus when we reach the temperature (*r*), no liquid remains, and the composition and temperature of the last freezing particles have reached (*s*). If diffusion were complete, these last freezing particles would then assume the composition (*r*), of the frozen portion.

Actually, diffusion is not complete, so that the first frozen particles contain less carbon than those last frozen. The austenite of the just frozen steel is thus heterogeneous, the first freezing particles forming a net-work of low carbon austenite, the axes of the nets being lowest in carbon.

During the cooling of the steel through the region *AaESOG*, further diffusion occurs; and if time enough were allowed for diffusion to complete itself, the initial heterogeneity of the steel would be effaced. With ordinary rates of cooling, however, the net-work of low carbon austenite (primary austenite), remains to form primary ferrite when the line *GOS* is crossed.

On reaching temperature (*t*) on the line *GOS*, free ferrite begins to be liberated from the (solid), austenite, the composition of the remaining mother austenite as the temperature falls sliding along the line *OS* until it reaches the composition *S*, when the remaining austenite transforms to pearlite, with the accompanying evolution of heat. In the liberation of free ferrite as the steel cools from *GOS* to *PS*, those parts of the steel lowest in carbon liberate their ferrite first, because their transformation point is highest; and thus the primary austenite net-work is perpetuated as a net-work of (primary) ferrite, which is made visible by etching. The (secondary) ferrite set free within the meshes of austenite forms finer net-works and veins.

Referring to Fig. 21, a cast steel showing typical net-work ferrite, it will be noticed that the heavy white lines are not arranged in a true net-work, but rather are roughly parallel. The inclusions of slag or sulphide of manganese are located at the axes of these ferrite lines. This suggests that in the original solidification of this steel, the austenite that froze first assumed the pine-tree or dendritic shape

generally characteristic of the first-freezing crystals. In such a method of crystallization, the austenite shoots out into the molten metal in long sheets, with parallel side branches. If there is anything in the theory of nucleus action, it is reasonable to assume that these dendrites may be formed upon slag or sulphide droplets as nuclei. According to this reasoning, the just frozen steel would be heterogeneous in the manner shown diagrammatically in Fig. 20. The lines Aa represent the low carbon axes of the first-freezing austenite

A   AXES OF LOW-CARBON DENDRITES
a „   „   BRANCHES OF   „
B   „   „   HIGH-CARBON AUSTENITE
X   SLAG OR SULPHIDE INCLUSIONS
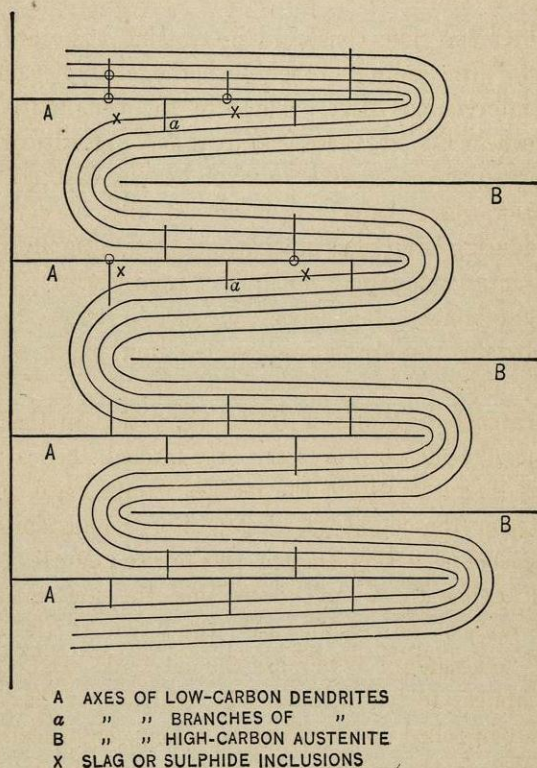
FIG. 20.—Dendritic freezing upon nuclei of slag or sulphide of manganese.

dendrites, the lines B, the corresponding axial lines of the highest carbon austenite rejected, during the process of freezing, into the spaces between the dendrites. It must be understood that the carbon content increases progressively from A to B; and that diffusion will partly eliminate the smaller side branches aa, and also, of course, lessen the difference in carbon content between A and B.

When the transformation range A_3 is reached, the low carbon axes A will liberate their ferrite first, because the liberation of

ferrite takes place at a higher temperature, and therefore sooner, in the lower carbon portions of the steel. If this ferrite tends to precipitate upon the slag or sulphide inclusions, as it is reasonable to assume it will, the ferrite will largely remain in the position of the axes of the original dendrites of primary austenite; and perhaps will concentrate strongly around the inclusions. The ferrite rejected from the space between A and B will to some extent follow the first precipitating ferrite, and no doubt will concentrate upon the inclusions.

The greater the time consumed in crossing the transformation range A_3 to A_1, the more thoroughly will the ferrite be concentrated from this intervening space. Thus, in the comparatively small casting shown in Fig. 21, there is a great deal of ferrite precipitated in this intervening space; while in much heavier castings, especially in the interior where the cooling is slowest, the concentration goes further, producing large lakes of ferrite about the slag inclusions, and leaving the intervening space much freer from ferrite. The coarseness of the original heterogeneity of such a large steel casting reinforced by this transformational segregation, makes these ferrite masses very large.

The researches of Professor Howe[1] show that in the formation of ferrite at A_3, the net-works are first formed, then a great deal of ferrite is liberated within the meshes of austenite, much of it being located in the octahedral cleavages of the austenite crystals; then, if the cooling is slow enough, this ferrite migrates to the net-works, thickening them, and decreasing the amount of ferrite within the grains. In certain cases, the ferrite in the octahedral cleavages is so abundant as to mask the net-works, and occasionally the latter appear to be lacking. The evidence at hand appears to show that in rolled or forged steels, a long, high heating (above A_3) so increases the tendency of this cleavage ferrite to persist that it is not absorbed into the net-works. Belaiew[2] obtained this structure, associated with a net-work, in a steel cooled very slowly from the solidification point. It is characteristic of meteorites; and has been found in masses of steel that are known to have been heated for a long time to an elevated temperature.

It is possible that the absence of slag or sulphide nuclei, upon which primary austenite can precipitate, may in some cases so retard the

[1] American Society for Testing Materials, 1911, p. 263. *International Journal of Metallography*, Vol. III, p. 4, and elsewhere.
[2] *Revue de Metallurgie*, 1910, p. 510; 1912, p. 321.

precipitation of that austenite in freezing that the initial heterogeneity of the steel is not as great as in the majority of cases. Should this tendency to restrain initial heterogeneity be reinforced by a cooling slow enough to allow diffusion to make the austenite quite homogeneous, the ferrite might be so largely precipitated at A_3 in the cleavages of the austenite grains, by a suitable rate of cooling through the transformation range, as to give rise to this structure to a very great extent. At any rate, the presence in cast steel of this Widmanstättian structure, as it is called, is often accompanied by a marked lack of slag or sulphide inclusions. Fig. 22 shows a typical example of this structure.

It may be that the amount of slag or sulphide inclusions may have a determining influence in fixing the size of the ferrite networks in steels that show a net-work structure. Thus the presence of but few inclusions, especially if they are large, may concentrate the primary austenite at a few points, and so give rise to a coarse net-work; and that of a great many small inclusions, to a fine net-work. If the location of the net-work ferrite is governed in part by the position of the primary austenite net-works, and if the ferrite set free at A_3 has a tendency to precipitate upon the inclusions, they may well have this effect upon the size of the structure.

The size of the casting, however, is the chief factor in determining the net-work size. This is partly, no doubt, because the austenite net-works formed in the slow freezing of the large casting are coarse; and partly because the migration of the cleavage ferrite into the net-works, and that of the smaller net-works into the larger ones, during the cooling from A_3 to A_1, has time to proceed very far, on account of the slow cooling of such large castings. The interior of heavy castings, moreover, exhibits to a marked extent the "balling up" of the net-work ferrite, and its concentration into great lakes (often surrounding slag or sulphide inclusions), which, as Professor Howe has shown in the articles referred to above, takes place to a marked extent during slow cooling.

The prime object of annealing or heat treatment is to break up by heating the coarse crystallization of ferrite and pearlite (ingotism, or ingot structure), and replace it by a uniformly fine crystallization of the constituents, with resulting strong inter-crystalline bonding; and the abolition of the long planes of inter-crystalline weakness due to the large sheets of ferrite in the slowly cooled steel. Heating the steel above AC_3, that is, above the line GOS, causes the pearlite to transform to austenite (in crossing AC_1), and the ferrite to dissolve

in the austenite (in crossing AC_3). The absorption of the ferrite, however (especially the primary ferrite), requires time, since it necessarily consists in the diffusion of carbon from the high carbon into the low carbon portions of the steel (a microscopic cementation, as it were, of the ferrite by the pearlite portions), and cementation we know is a gradual process.

The coarseness of the microstructure of the steel, as we have seen, depends chiefly upon the size of the casting, which determines the rate at which it has frozen and cooled. The time necessary for a fairly complete diffusion of the carbon at a given temperature, therefore, depends upon the size of the castings, since the larger the ferrite areas the longer it takes to cement them. This necessary time should be counted, not from the moment that the annealing furnace reaches the desired temperature, but from the time when the casting itself is hot through. It has been found, however, that the time necessary to cement the small ferrite areas in a piece of a casting $\frac{1}{2}$ in. thick, is not a great deal shorter than the time needed to cement the large areas in a piece cut from the center of a casting 6 in. thick; so that evidently the time actually needed to eliminate the larger masses of free ferrite is not as much greater than that necessary to eliminate the small masses, as their difference in size would lead us to expect.

The carbon-iron diagram shows that .60 per cent. carbon steel can be annealed at a lower temperature than .20 per cent. carbon steel. But the rate of diffusion is so much lower at the lower temperatures, that for practical purposes it will in many cases be found convenient, for steel normally containing free ferrite, to use a uniform annealing temperature of from 850 to 900° C.

The castings should be heated to this temperature for at least two hours and preferably four hours, *after they have become hot through*—very large castings may be heated longer if it is found necessary. The total annealing time of the large castings will in any case be considerably longer than that of the small ones, as it requires so much longer to bring them up to heat throughout their mass.

By heating cast steel containing free ferrite to about 1200° C. for a few minutes, the diffusion of the carbon is made much more rapid—but a coarse structure will generally be produced by heating to this temperature, to remove which a second heating to about 900° is required. Hence, it is generally a saving of fuel to use a single heating to about 900°, for from two to four hours. The high heat (1200°) is expensive in fuel, in wear and tear on furnace brick-

work, in scaling of the steel, and in excessive warping of the castings, and partakes more of the nature of a laboratory method than of a practical one.

Now, upon cooling from the annealing temperature, these initially coarse-grained cast steels have a strong tendency to reprecipitate their ferrite in heavy masses, apparently located to a great extent along the lines of the old primary ferrite net-work. Steels which possessed the Widmanstättian arrangement of the ferrite appear on slow cooling from the annealing temperature to show a ferrite-pearlite structure of fair uniformity, though not fine grained; but steel that was of the net-work class shows ferrite in very coarse masses. Reference to the accompanying microphotographs illustrates this very clearly.

Figs. 21 and 22 show the characteristic microstructures of medium carbon steel in the cast condition, Fig. 21 being of a typical net-work structure steel, Fig. 22 of a steel in which the Widmanstättian structure predominates.

ANALYSIS OF STEEL SHOWN IN FIGS. 31–36

No	C	Si	Mn	S	P
21	.43	.38	.47	....	....
22	.37	.32	.51	....	....
23	.42	.51	.60	....	....
24	.39	.48	.47	....	....
25	.41	.45	.51	....	....
26	.43	.49	.60	....	....
27	.41	.52	.68	....	....
28	.39	.41	.55	....	....
29	.35	.42	.65	....	....
30	.35	.42	.65	....	....
31	.59	.14	.72	0.23	.017
32	.21	....	1.17	....	....
33	.18	.41	.93	.065	.052
34 35 36	.20	.45	1.02	.057	.049

Fig. 23 shows a similar steel annealed for two hours at 900°, cooled slowly to 800° and quenched in water; and Fig. 24 a steel annealed for two hours at 900°, cooled slowly to 750° and quenched in water. Fig. 23 shows a very uniform sorbitic structure (the cooling being too slow, owing to the size of the test bar, and the carbon too low, to produce martensite), while in Fig. 24, as the steel has been allowed to cool slowly to a temperature below Ar_3, the ferrite has reprecipitated, and shows a marked tendency to the net-work arrangement. This tendency of the ferrite to coarse crystallization
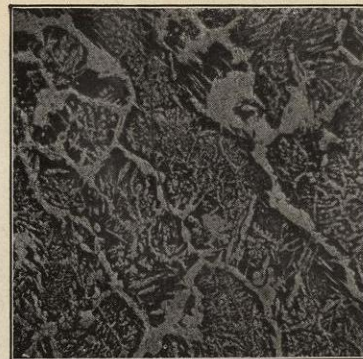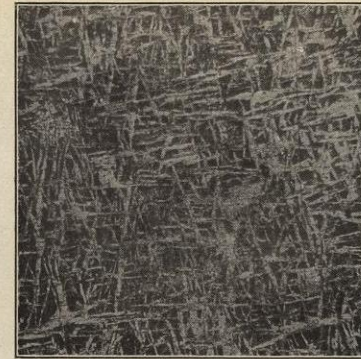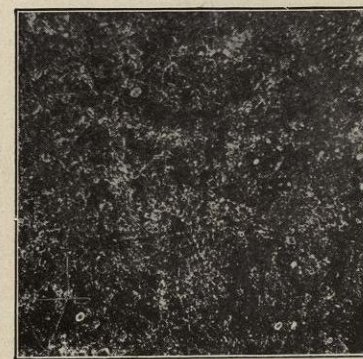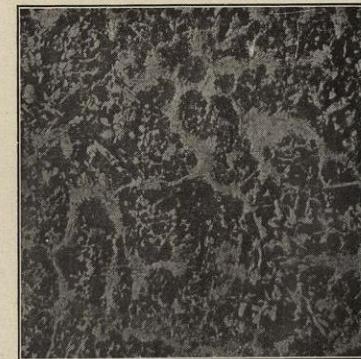

FIG. 21.   FIG. 22.

FIG. 23.   FIG. 24.
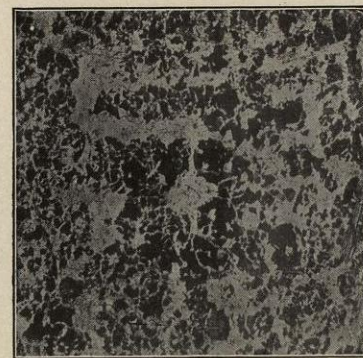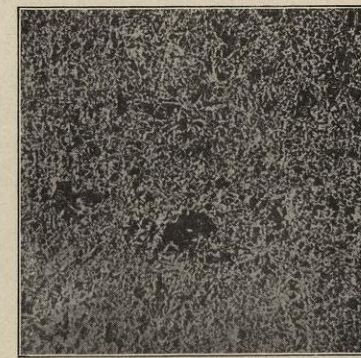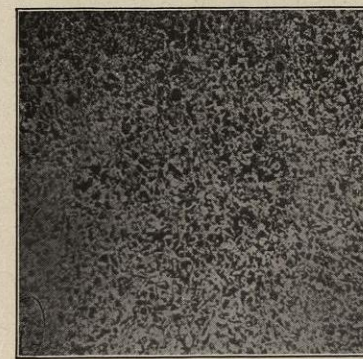
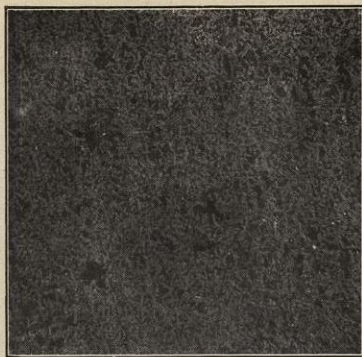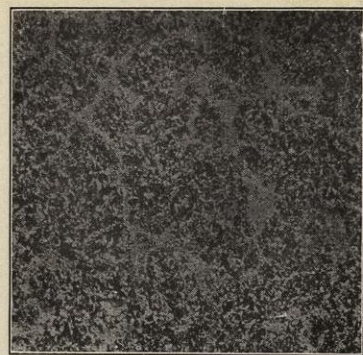FIG. 25.   FIG. 26.

FIG. 27.   FIG. 28.

Reduced about ⅔ from original magnification of 60 diameters.

FIG. 29.

FIG. 30.

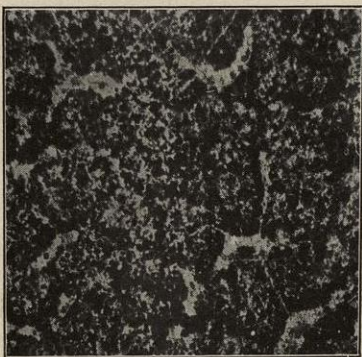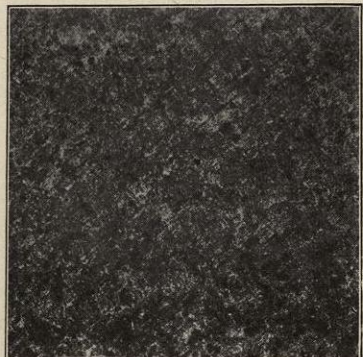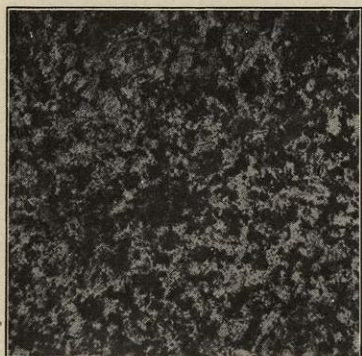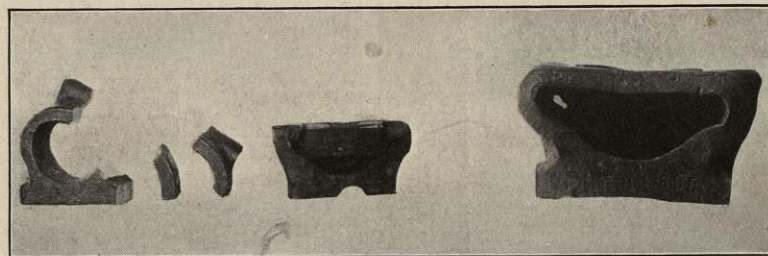FIG. 31.

FIG. 32.

FIG. 33.

FIG. 34.

Reduced about ⅔ from original magnification of 60 diameters.

FIG. 35.—About ⅙ full size.   FIG. 36.—About ⅓ full size.

is also very strongly marked in Fig. 25, a similar steel annealed at 900° for three hours and cooled very slowly; six hours were spent in cooling this bar from 900 to 600°. There is more ferrite in Fig. 25 than in Fig. 24, because in the latter the separation of ferrite that normally takes place in cooling from 750 to 680° has been partially suppressed by quenching.

The black areas in Fig. 24 are sorbite, because the rapidity of the final cooling has prevented the formation of true (lamellar) pearlite; in Fig. 25, the black areas are pearlite. Higher magnification should show this clearly.

Fig. 26 shows the structure of a similar steel annealed at 900° for three hours, and cooled in the air in about 10 minutes. Here, evidently, we have quite a different structure from any that we have yet seen. The black sorbite remains, and in two places it shows a strong tendency to dominate and give a structure like that of Fig. 23, but the cooling, although not sufficiently rapid to repress altogether the liberation and agglomeration of ferrite, yet has not been slow enough to allow the ferrite to agglomerate to such an extent as to leave large ferrite areas. Hence, we have the comparatively fine ferrite net-work, resembling somewhat a rolled steel after annealing.

Fig. 27 shows a similar steel heated to 900° for one hour, and cooled to black in about 40 minutes; the cooling through the critical range has been too rapid again to permit the formation of large ferrite patches, though, as will be seen by comparing Fig. 26 and Fig. 27, the coagulation of the ferrite has partially taken place, since the ferrite in Fig. 27 is noticeably coarser than that in Fig. 26. The white areas in Fig. 27 are ferrite, the black, sorbite or pearlite.

The formation of lakes of ferrite which has been prevented by quenching as shown in Fig. 23, or by air cooling as shown in Fig. 26, will take place again if the steel is reheated to $Ar1$, or nearly to that point. Curiously enough, the reappearance of ferrite in a coarse net-work takes place at a lower temperature in the reheating of a quenched steel than in the reheating of an air-cooled steel.

Thus, Fig. 28 shows a steel of this class annealed at 900° for two and one-half hours, cooled slowly to 800° and quenched (resulting of course in structure like that of Fig. 23), then reheated to 700° for a few minutes and again quenched. The separation of very finely divided ferrite from the sorbite background is clearly shown, and also the agglomeration of ferrite into lakes in a partially defined coarse net-work.