"THE TRANSFORMATION BEHAVIOUR
OF CERTAIN ALLOY
CASE HARDENING STEELS."

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Abstract.

The isothermal transformation diagrams have been determined for the case and core compositions of a series of four carburising steels, and the physical properties resulting from various alternative heat-treatment procedures suggested by the diagrams determined. The dimensional changes occurring during all stages of isothermal heat-treatment have been determined, and the effects of these dimensional changes on the stress developed between the case and core of carburised specimens demonstrated. The most suitable heat-treatment procedures for optimum core properties with adequate case hardness and minimum case/core stress are given.

It was noticed that during isothermal transformation of certain of the case compositions, changes in as-quenched hardness occurred before the commencement of transformation. This behaviour appeared to be associated with the presence of carbon and carbide forming elements. The investigations were therefore extended to include a series of chrome molybdenum hot-work steels having varying tungsten and vanadium contents. By plotting the hardness values obtained on the isothermal transformation specimens on a perspective summary diagram for each steel composition, three types of behaviour were shown, and an explanation is proposed for these alternative behaviour patterns.
1. Introduction.

Isothermal techniques of heat-treatment continue to be adopted with advantage for the processing of alloy steels. It was, therefore, felt to be of interest to investigate the isothermal behaviour of the carburising steels, including the high carbon case compositions. It was envisaged that with a fuller understanding of the isothermal behaviour of both the case and core compositions for a given steel, it might well be possible to devise suitable improved heat-treatment techniques, whereby optimum core and case properties could be obtained with the minimum of manipulation.

It was considered important to obtain an assessment of the dimensional changes occurring during the various alternative heat-treatment techniques. Such dimensional changes are likely to be of particular importance when carburised steels are heat-treated. Any difference in dimensional change between the carburised surface and core composition of such steels, will show itself as a stress between the surface layer and the core, both during and after heat-treatment, and could well affect the properties of the surface. In extreme conditions, cracking or spalling of the carburised case could result. A further effect with very thin section components could be distortion of the component by the core/case stresses.

It was therefore proposed to commence the investigation by determining the isothermal transformation diagrams for the normal and carburised compositions of a series of case hardening steels. The steels chosen were, EN 35, 36, 37, and 39 of B.S. 970. This selection, in addition to including the case hardening steels in common use, gave a series having increasing nickel and chromium content, and therefore offered the opportunity of assessing the effects of these elements on transformation.

When the transformation diagrams had been obtained, it would be possible to compare them with a view to devising possible heat-treatment techniques compatible with the response of both the case and core for a given steel. Test pieces could then be processed in accordance with the heat-treatment procedures that appeared feasible, and the physical properties resulting determined. This part of the investigation would also serve to confirm the accuracy of the transformation diagrams.

In order to complete the investigations, it was intended that the dimensional changes occurring during all the stages of heat-treatment would be quantitatively determined, and a qualitative assessment made of the stresses developed between the case and core as a result of such heat-treatments.
During the determination of the transformation diagrams by the metallographic procedure careful hardness measurements were made on the as-quenched specimens. It was noticed that with the carburised composition, the expected pattern of change of hardness with the onset and progress of transformation was not always shown. Some of the steels, with certain conditions of transformation, showed an initial lowering of the as-quenched hardness. As the onset of transformation was approached, however, the hardness was found to increase to a maximum corresponding to the commencement of transformation, reducing again in the normal manner as transformation proceeded. It was felt desirable to study this phenomenon further, as it had not been reported previously.

It appeared to be a form of "secondary hardening", which had previously been observed under certain conditions of tempering hot work tool steels of the 5% chromium, 1% molybdenum type. The investigations were, therefore, extended to include a series of steels of this type having varying tungsten and vanadium contents. In addition to exploring the transformation behaviour of the normal composition with a carbon level of approximately 0.3%, the carburised compositions were also included. Under appropriate conditions, steels in this series also exhibited the anomalous behaviour noted with some of the carburising steels, and although a completely satisfactory explanation of the phenomenon was not arrived at, certain tentative suggestions have been made.
2. Historical Survey

The iron carbon equilibrium diagram has of course been known for a very considerable time. In accordance with this diagram, annealed carbon or alloy steels consist of ferrite or alpha iron solid solution and carbide at room temperature. On heating, these two phases react together to form a single solid solution of carbon and alloying elements in austenite or gamma iron. This reaction normally begins at the eutectoid temperature, the first ferrite and carbide reacting together in a definite eutectoid proportion. On raising the temperature further, the remaining ferrite of hypoeutectoid steels or carbide of hypereutectoid steels dissolves in the austenite. These changes are reversed on cooling. Due to the slow rate of diffusion in the system, however, considerable time is required for equilibrium conditions to be reached during both heating and cooling. This has necessitated the introduction of the designation Ar and Ac to represent the temperatures of phase transformation observed on cooling and heating respectively, in addition to the designation Ae for the equilibrium temperature. For a given steel composition, the values observed for Ar and Ac depend largely upon the rate of heating or cooling, and their difference from Ae represents the apparent displacement of the phase transformation temperature as a result of an inadequate time allowance for equilibrium conditions to be obtained.

The original conception of hardening techniques for steels were all based on the attainment of a sufficient cooling velocity to cause the Ar temperature to be displaced to a temperature, below that at which the austenite formed by the original heating of the steel would transform to martensite. In this way more or less complete transformation to the martensite phase was achieved.

Some indication of the time required for the attainment of equilibrium conditions at various temperatures was obtained by Portevin & Jarvin (1) in 1919, by Chapin (2) in 1922 and by Lewis (3) in 1929.

The first systematic work to add the time factor to the iron-carbon diagram, however, was by Davenport and Bain Kearny (4) in 1930. After heating small specimens of the steel under investigation to the austenising temperature, these authors quenched them into a bath of high cooling power automatically controlled at a constant temperature. In this way the specimens were rapidly cooled as austenite to a
predetermined temperature, and then allowed to transform at that constant temperature. By repeating the procedure at a series of quenching temperatures from the lower critical point down to room temperature, the general trend of the effect of transformation temperature upon the time of transformation and upon the nature of the product was worked out. The results were plotted on a summary diagram called by the authors a transformation time graph, or S curve, from the approximate S shape of the curves.

Two methods were used to follow the progress of transformation. For temperatures from the lower critical down to about 300°C a metallographic technique was used. This technique consisted of quenching in water, specimens which had been previously austenised, quenched to the transforming temperature, and held for varying times, the untransformed austenite remaining being fixed as martensite by the water quench. This martensite may be easily distinguished metallographically from the products of decomposition of austenite when this decomposition occurs in the temperature range down to 300°C.

The second method used for the lower temperatures depended upon the increase in volume that occurs when steel transforms from austenite to either martensite or troostite. This volume change was followed as a change of length, using a dilatometer similar to that previously used by Bain & Waring (5) and by Grossman (6). This dilatometer was essentially of all metal construction and was, therefore, subject to inaccuracies due to temperature and, therefore, length variations in the side rods. It is illustrated diagramatically in Fig. 1a. In an attempt to reduce the length changes in the side rods, Davenport and Bain's technique was to preheat the dilatometer in the transforming bath prior to the insertion of the specimen to be transformed. For inserting the specimen, the dilatometer was raised in the bath and the specimens quickly inserted, by means of heated tongs. The accuracy of the procedure depended upon how nearly the specimen retained the austenising furnace temperature and how nearly the dilatometer retained the transforming bath temperature.

No doubt due to the inaccuracies inherent in a dilatometer of this type, particularly when used for following transformation at low temperature ranges, these authors did not draw any distinction between the transformation behaviour when martensite was formed at the lower temperature as compared with the formation of pearlite and troostite at the higher temperatures. In their paper it is inferred that both types of transformation are time dependent.
Further important contributions to an understanding of the isothermal behaviour of steel were made in 1939 by Allen, Pfeil and Griffiths (7). For their investigation these authors used a dilatometer, of much greater inherent accuracy. The portion of the dilatometer immersed in the transforming bath was constructed of silica, as shown in the diagram Fig. 1b. The specimens were mounted in the dilatometer prior to austenising. After the correct austenising conditions had been achieved, the dilatometer and specimen were transferred rapidly to the transforming bath.

Allen, Pfeil and Griffiths were able to investigate in a quantitative manner the three transformation characteristics: the period of induction, the velocity of transformation, and the completeness of transformation. They showed that at the higher transformation temperatures (above 550°C) the rate of decomposition of the residual austenite increases continuously as the change proceeds. This behaviour is largely consistent with the hypothesis put forward by Mehl (8) for the pearlitic change. Mehl postulated that this type of change spreads out at a uniform linear rate from a number of nuclei, the rate of change being proportional to the ratio of the surface area of the transformed product to the volume of untransformed material. As this ratio increases continuously with the growth of the transformed product, so the rate of transformation progressively increases. The results obtained by Allen, Pfeil and Griffiths, however, did not completely agree with Mehl's hypothesis in one respect. Mehl assumed the production of fresh nuclei at a uniform rate during the pearlitic change. This is not compatible with the behaviour found by Allen, Pfeil and Griffiths for alloy steels, some of which, for example, show no transformation at all for a considerable time, and then begin to transform in a way which does not indicate a uniform rate of production of nuclei.

The intermediate change occurring at lower temperatures was shown to proceed at the uniform rate characteristic of a first order reaction, once its full velocity had been reached during the induction period.

Transformations occurring in the range of 500°C were shown to be more complex, but separable into two overlapping reactions, having the characteristics of the intermediate and pearlitic changes.
In addition to the very thorough quantitative treatment of the progress of transformation by these authors, they also showed how the martensite change appeared to be distinct from the pearlitic and intermediate changes in being almost entirely independent of time.

Further progress was made in the understanding of the martensite reaction as a result of the work done by Carpenter and Robertson (9) from 1929 onwards and summarised by them in 1939. They concluded that the progress of the martensite reaction was dependent entirely upon the temperature to which the metal was cooled, and was independent of the rate of cooling within the martensite range.

The work of Carpenter and Robertson was confirmed by Greninger and Troiano in a paper published in 1940 (10). These authors used a metallographic procedure based on the fact that tempered and untempered martensite contained in the same specimen are easily differentiated microscopically after nital etching, the former etching dark brown while the latter is almost unattacked and remains white. This was not a new technique, having been used by previous investigators, particularly by Hanemann (11). The procedure of Greninger and Troiano was to rapidly quench small specimens from the austenising temperature to the temperature under investigation. After holding at this temperature for an appropriate time, the specimens were then rapidly transferred to a tempering bath of liquid tin maintained at 280°C and held there for 5 seconds, after which they were finally quenched to room temperature in 10% caustic soda solution. By means of this technique, Greninger and Troiano were able to show that, within the martensite transformation range, the formation of martensite was independent of time, but dependent upon temperature. In addition, they indicated that at any given temperature within the martensite transformation range, the austenite untransformed to martensite, if held at constant temperature for a sufficient time, will eventually decompose according to some complex function of time and temperature. The further indicated that the rate of this constant temperature decomposition is at least initially greater at temperatures just below the upper limits of the martensite range than at temperatures just above this range.
1948 saw the publication of a paper by Grange and Stewart (13) which dealt quantitatively with the effect of composition on the martensite transformation range. These authors, after investigating the martensite transformation range for a number of alloy steels, using the metallographic technique of Greninger and Troiano, combined the data obtained with these steels with other pertinent published data, for the purpose of developing an empirical relationship between chemical composition and the proportion of martensite formed on quenching to any given temperature. While the formula proposed by these authors for the calculation of the martensite start temperature has been subsequently revised by other authors, it indicated a predictable behaviour for the martensite transformation, and proved to be a useful guide in planning the heat-treatments of alloy steels.

Several methods have been used for following the transformation of austenite in steel. The original method of plotting inverse-rate curves from time/temperature data obtained during heating and cooling of the steel, was superseded by the metallographic and dilatometric procedures of Davenport and Bain (4) already mentioned. The dilatometric procedure was improved upon by Allen, Pfeil and Griffiths (7) also previously mentioned and has since been the subject of further modifications designed to improve its accuracy and, in some cases allow for automatic recording of the dimensional changes taking place. One combination of such improvements has been the "Leitz recording dilatometer", the operation and construction of which was described by Borchels and Otto (13) in 1942.

In addition the measurement of changes in magnetic flux (Russell and Mavrocordatos (14)), and changes in electrical conductivity (McReynolds (15)), changes in hardness (Blanter (16)), x-ray diffraction methods, and acoustic methods (Forster and Scheil (17)) have all been used to follow the austenite transformation.

A detailed examination of the dimensional changes observed during the progress of transformation was published by Russell and Mavrocordatos (14) in 1949. These authors, particularly noted that with high carbon contents, the expansion due to transformation was appreciably lower than that observed with low carbon compositions. It was not indicated, however, how these dimensional changes related to those occurring during heating from room temperature to the austenising temperature and subsequently cooling, after transformation, from the transforming temperature back to room temperature.
Subsequent publications dealing with the isothermal transformation of steels have been largely confined to further detailed investigation of the mechanism of the Ferrite (Hickley and Woodhead (18)), Bainite (Ko and Cottrell (19), Ko (20), Hehemann and Troiano (21), and Steven and Haynes (22)), and Martensite (Philibert and Cussard (23), and Stevens and Haynes (22)) reactions. The transforming behaviour of carburising steels, including the high carbon case, has been limited mainly to the publication of certain transformation diagrams (24).

The four carburising steels to be investigated were obtained in the form of $\frac{1}{4}$ inch diameter drawn rods and 1 inch diameter rolled bars. A suitable number of micro tensile test specimens were machined from a portion of the $\frac{1}{4}$ inch diameter rod material and these specimens used to determine the mechanical properties obtained with different heat-treatment procedures. The remainder of the $\frac{1}{4}$ inch diameter rod material was cold rolled down to 0.025 inch thick strip as a preliminary in the preparation of dilatometer and metallographic test specimens. The 1 inch diameter rolled bar material was used for investigating possible heat-treatment procedures based on the isothermal diagrams obtained, as well as for the production of special test pieces for estimating the level of stress likely to be present between the core and carburised case.

The hot work tool steels of the 5% chromium, 1% molybdenum type used to further investigate the changes occurring prior to the onset of transformation, were not available in small section sizes. The dilatometer and metallographic specimens required were therefore obtained by machining $\frac{1}{4} \times 0.024$ inch strips from forged billets.

The chemical analysis of the steels used is given in Table 1.

The methods used for the determination of the transformation diagrams consisted of a modification of the usual dilatometric and metallographic procedures now in use. Two dilatometers were used; the one, a special high speed dilatometer, and the other, a more conventional hand dilatometer. The majority of the work was done using the metallographic procedures of Davenport and Bain (4).

While the high speed dilatometer was used in order to more accurately determine the commencement of transformation in the more rapidly transforming steels, the hand dilatometer was found of use later in the investigations in order to explore the amount of expansion occurring on transformation, as well as other dimensional changes.

The High Speed Dilatometer.

The principle of the high speed dilatometer was based on a fixed specimen suspension with moving austenising and transforming furnaces. The general arrangement of the equipment was as shown in Fig. 2a and diagramatically in Fig. 2b. The dimensional changes of the specimen were indicated by a dial micrometer graduated to read to $10^{-3}$ inches. A stop watch was mounted on the dilatometer adjacent to the dial micrometer so that recordings of the time and dimension could be made by means of a cine camera. The specimens used were in the form of strip, 0.025 inches thick and $\frac{1}{4}$ inches wide, suspended at $1\frac{1}{2}$ inch centres.

For normal operation the austenising and transformation furnaces were locked together in the lowered position as shown in Fig. 3a and diagramatically in Fig. 3b, position 1. After austenising the specimen for the required time, the position of the furnaces was rapidly changed, on operation of the release mechanism, by means of the counterpoise weights and decelerating springs. In this way the specimen was rapidly quenched into the lead bath used as the transforming medium. The position of the furnaces at this stage is illustrated in Fig. 3b and diagramatically in Fig. 3b, position 2.

Prior to operation of the furnace transfer mechanism, the recording camera motor was started. This ensured that the camera was running at full speed when quenching of the specimen occurred. Photographic recording was commenced.
by the upward movement of the furnaces operating a
switch which caused the shutter solenoid to become operative.

By making the furnaces as light as possible and, at
the same time, using very heavy counterpoise weights in
conjunction with decelerating springs, very rapid furnace
movement was achieved. Close inspection of Fig. 2 will show
that the stop watch and dilatometer proper were mounted
completely independently of the furnace transfer mechanism,
in order to reduce vibration to a minimum. With the rapid
furnace movement achieved, no difficulty was experienced
in transferring the specimen from the austenising furnace
to the transforming bath without any measurable change in
length and therefore, presumably, any measurable change
in temperature, occurring.

The upper portion of the specimen suspension consisted
of a silica rod moving inside a fixed silica tube. This
inner moving rod terminated in a hook portion which engaged
the lower end of the specimen. The outer silica tube was
continued also as a silica rod which engaged by means
of a hook, with the top portion of the specimen. The details
of the specimen mounting are shown in Fig. 4a and
diagramatically in Fig. 4b. The use of small diameter silica
rods for the portion of the suspension in the austenising
furnace and the transforming bath, besides facilitating
the easy attachment of a strip specimen, also gave a light
and therefore reduced thermal capacity suspension.

The specimens used with this dilatometer, as already
mentioned, were prepared by cold rolling the \( \frac{1}{2} \) inch diameter
rod material down to a thickness of 0.025 inches. The
resulting strip was then sheared to a width of \( \frac{1}{2} \) inch and
cropped into \( \frac{13}{16} \) inch lengths. The material was then close
annealed. Each piece was then drilled with two \( \frac{3}{16} \) inch
diameter holes at \( \frac{13}{16} \) inch centres along the centre line,
giving an effective working length when suspended in the
dilatometer of 1.63 inches. Those specimens which were
required to represent the carburised case, were then pack
carburised at 900°C for 2 hours in activated charcoal.
This resulted in carburising completely through the thickness
of the material, giving an average carbon content of 0.9%.
All the specimens were then nickel plated to a minimum
thickness of 0.0007 inches. While this plating did not
completely eliminate de-carburising during subsequent thermal
treatment, it limited it to approximately 0.0005 inches
in depth.

The method of operating the high speed dilatometer
was to manually release the furnace transfer mechanism
after the specimen had been held for a suitable time at the
austenising temperature, and after having started the recording camera motor. The action of furnace transfer automatically energised the camera shutter solenoid and so commenced automatic recording. Arrangements were provided on the apparatus whereby, after the first few minutes in the transforming bath, the continuous recording could be switched off and the camera control transferred over to intermittent automatic recording. This was arranged by the adaptation of an austenising furnace temperature recorder. This adaptation consisted of the suitable positioning of a limit switch so as to be engaged at regular intervals by the recording drum sprockets as shown in Fig. 5. This arrangement exposed a few frames every 15 minutes in order to cover the longer time portions of transformation. During the initial continuous recording, 20 frames per second were used.

The furnace temperatures were controlled by means of conventional type controllers in conjunction with adjustable rheostats. The furnace temperatures were in this way controlled sufficiently close to produce dimensional changes in the specimens limited to 0.0002 inches.

As a result of the large surface area volume relationship of the specimens used, together with the low mass of the suspension, very rapid quenching was achieved. The time required for the specimen to cool to within 10°C of the temperature of the transforming bath varied slightly according to this temperature, but was never in excess of 0.3 seconds, and was usually about 0.2 seconds, whilst to reach equilibrium about 0.5 seconds appeared to be needed. It was noticed that during quenching some over-shooting of the thermal contraction occurred which, it was assumed, was due to the mechanical inertia of the system. Typical contraction curves obtained are shown in Fig. 17.

The Hand Dilatometer.

The hand dilatometer used is illustrated in Fig. 6. It utilised the same type of specimen and suspension as the high speed dilatometer. Provision was made on this dilatometer for thermocouple leads to be attached to the specimen itself in order to record accurately the changes in temperature of the specimen when plotting heating and cooling curves during the investigations on dimensional changes. The dilatometer head itself was protected from the heat of the austenising furnace and transforming bath by means of a water jacket. This is illustrated in Fig. 7, which shows the dilatometer set up on the furnace bench over one of the austenising furnaces.
Metallographic Procedure.

The specimens used for the metallographic determination of the transformation were prepared from part of the 0.025 inch thick strip used for the dilatometer specimens. This ½ inch wide material was cropped into ½ inch lengths, each piece having a small hole drilled on the centre line of the original strip. These test pieces were then nickel plated to a minimum thickness of 0.0007 inches to prevent decarburisation. In this way a small specimen was produced of a shape that ensured very rapid cooling to the transforming temperature. The time intervals used for transformation were in the progression of 2, 5, 10, 20, 50, 100, 200, etc., seconds, up to a maximum of 100,000 seconds. After treatment, the series of specimens representing one transformation temperature were bolted together in two groups to facilitate mounting and subsequent polishing for examination. The specimens and method of mounting are illustrated in Fig. 8.

Similarly to the dilatometric specimens, a suitable number of these metallographic specimens were pack carburised at 900°C for two hours to produce a composition representative of the carburised case. After mounting and before polishing, ⅛ inch was ground off the edges of the specimens in order to remove the nickel plate and any slightly decarburised surface layer that might be present.

The progress of transformation was followed both by the metallographic appearance of the specimens and by hardness measurements. The procedure used for hardness checking was very precise, a V.P.N. hardness being done at the centre of each specimen edge as presented in the mounted block, after metallographic polishing but before etching. A 1 kilogramme load was used, agreement being obtained to within 1 to 2 ocular readings.

The Martensite Transformation Range.

For the diagrams of the core compositions, the martensite start temperatures were calculated using the formula of Steven and Haynes (24), the martensite finishing temperatures being also calculated, using the diagram proposed by Grange and Stewart (12).

The martensite transformation ranges for the carburised compositions were determined directly, using the procedure outlined by Greninger and Troiano (10). Test pieces were austenised for ½ hour at 900°C and quenched to different temperatures (175, 150, 135, 120, 105, 90, 80, 10, 0, -7, -10°C), held in the quenching
bath for three minutes, reheated to 200°C in an oil bath for five minutes, and finally quenched to room temperature. The specimens were then mounted using a cold setting resin, polished, and etched in nital. The percentage of tempered martensite was then estimated metallographically, and represented the amount of martensite formed by the initial quench.
5. The Transformation Diagrams Obtained.

Figs. 9 to 16 are the isothermal transformation diagrams obtained. The diagrams covering transformation of the cores are based on data obtained by means of the high speed dilatometer, confirmed by metallography and hardness measurements. The diagrams for transformation of the cases are based on information obtained with the hand dilatometer, again with metallographic confirmation.

The high speed dilatometer curves obtained for the core composition of EN.33, are shown in Fig. 17. In Fig. 18 the contractions per inch observed on quenching the specimens to the transformation temperature are plotted against the various temperatures used. The graph obtained is essentially a straight line, indicating that no transformation occurred during the cooling period. Any transformation which did occur during cooling from austenising would have resulted in an expansion, which, when superimposed on the thermal contraction, would have given a reduced total contraction. The results obtained give a coefficient of thermal contraction of austenite of $0.026 \times 10^{-5}$ inch/inch/$^\circ$C. Very similar figures were obtained for the other steels investigated with the high speed dilatometer. The contraction observed agrees very closely with that obtained by Allen, Pfeil, and Griffiths (7).

Fig. 17 also demonstrates the rate of quenching of the specimen, indicating that a temperature within $10^\circ$C of the transforming bath temperature is reached within 0.2 seconds, the bath temperature being approximated to within 0.5 seconds. The apparent small expansion of the specimen at about 0.4 seconds, prior to the final attainment of the quenched dimension, shown in Fig. 17, is believed to be the result of mechanical inertia in the dilatometer suspension. Typical expansion curves for complete transformation are plotted, again for the core composition of EN.33, in Fig. 19.

Fairly close agreement was obtained between the dimensional changes indicative of transformation and changes in hardness of the metallographic specimens. It was noticed in many instances that the hardness observed before transformation commenced did not remain constant as expected. There was a tendency for a progressive rise in hardness to occur up to the commencement of transformation. No obvious change in the metallographic structure could be observed, however, to account for this change in hardness. This rise in hardness was most noticeable in the carburised structures, particularly with EN.39 in the transformation
temperature range of 400/500°C. Here a hardness change of up to 200/250 VHN was observed. A possible explanation of the phenomenon would be a change in the composition of the carbides present in solid solution in the austenite. This change in hardness before the breakdown of austenite has also been observed during the investigation of the isothermal transformation of more complex alloy steels. The changes occurring before the isothermal breakdown of austenite in both the case hardening steels and the higher alloy steels investigated, are reported and discussed in detail later.

The Ferrite Transformation Range.

The ferrite transformation range of the core compositions was investigated with respect to the rate of growth of ferrite and the rate of nucleation. Counts were made of the number of nuclei present as transformation proceeds, together with an estimate of the maximum and average ferrite patch diameter. It was considered that the smallest size patch visible for counting was 1 x 1 mm at a magnification of x 1400. This approximates to an actual diameter of 7 x 10^-4 mm, and compares with the smallest visible size assumed by Moore (25) of 10^-5 cm.

As recorded by Hickley and Woodhead (13), a quantitative evaluation of the size of ferrite particles is limited to transformation in the upper temperature ranges where the ferrite takes the form of irregular polygonal grains or "blocky ferrite". In the lower ranges of transformation the ferrite takes on an acicular form, extending to envelopes around the austenite grains, with plates growing into the grains. The morphology is illustrated in Micros Nos. 1 to 5. It is not practical to make an assessment of the nucleation and growth rates by the method of field counts when the ferrite assumes this acicular form, both because of the difficulty of assessing the area occupied and because of the early interference and coalescence with each other of particles of this shape.

Fig. 20 illustrates the typical curves obtained for the rate of nucleation with EN 33 core between 600 and 750°C. The results confirm the findings of Hickley and Woodhead (13) for plain carbon hypoeutectoid steels, to the effect that the rate of nucleation at constant temperature decreases with time, and increases with decreasing transformation temperature. The apparent reduction in the number of nuclei present eventually shown, is the result of coalescence as growth proceeds. An estimate of the rate of growth was made by plotting the average ferrite
patch diameter against log time. Fig. 21 confirms the finding of Hickley and Woodhead (18), a straight line relationship being obtained up to the beginning of coalescence. From the slope of the graphs for different transformation temperatures it is seen that the rate of growth increases with fall in transformation temperature. In comparing the transformation at upper temperatures with that at lower temperatures, it is seen that in the higher temperature ranges where "blocky ferrite" if formed, nucleation and growth tend to continue side by side over an appreciable part of the total transformation time. At the lower temperatures, when the ferrite is of the acicular type, nucleation appears to cease at an early stage when the ferrite envelopes have been completed. Further growth is then as plates extending from these envelopes into the austenite grains. This is as would be expected, considering that less nucleation energy will be required at the grain boundaries than within the grains, because of the lattice discontinuity at the grain boundaries. It would appear that the favourable grain boundary sites are all occupied in the early stages of transformation. While the conditions are then such as to allow transformation to proceed by grain growth, there is insufficient energy for further nucleation within the austenite grains.

The Bainite Transformation Range.

The morphology of the structures resulting from transformation in the bainite range is particularly interesting as some indication of the mechanism of the reaction. All the structures are essentially acicular, transformation proceeding by coherent growth of the plates within the austenite grains. Micros Nos. 6 to 8 illustrate the structures obtained with the low carbon cores at various temperatures of transformation. Micro No. 6 shows the results of partial transformation at 550°C in EN. 39. Grain boundary acicular ferrite is present as well as the bainite plates. It should be particularly noted that these bainite plates form within the austenite grains rather than at the grain boundaries. This indicates that while ferrite forms from nuclei not coherent with the original austenite grains, bainite originates from a distinct type of nucleation which is within and coherent with the austenite.

At the transformation temperature illustrated in Micro No. 6, both processes are occurring together. The simultaneous formation of alpha iron from austenite by both coherent and incoherent transformation has already been observed and noted by Ko (20) during the transformation of EN. 21 steel. This author also notes that the formation of bainite occurred earlier at the surface of his specimens than elsewhere. This was also noticed during our investigations, particularly when investigating the "hot work" steels. It would be expected that less constraint would be imposed
on a coherent transformation at the surface, as the new lattice structure would not be surrounded on all sides by the original austenite lattice.

It should be noticed also that there is no obvious carbide precipitation in Micro No. 6. It is reasonable to assume that at this high temperature the diffusion rate of carbon is sufficiently high for it to diffuse out of the alpha iron into the surrounding austenite. This is also facilitated by the relatively wide spacing of the bainite plates. That diffusion of carbon from the alpha to the gamma phase does occur during the formation of upper bainite has already been shown by Wever, Rose, and Peter (26, 27, 28).

Considering the formation of bainite at lower temperatures, this is illustrated in Micros Nos. 7 and 8. Micro No. 7 is of EN. 57, the early stages of transformation at 500°C. Here the individual bainite plates do not appear so distinct, being closer together and showing obvious carbide precipitation between them. Micro No. 8 is of EN. 39 transformed at 400°C. While an acicular structure is still in evidence, the carbide precipitation is even more marked.
6. Comparison of the Transformation Diagrams for the Case and Core.

Comparison of the four diagrams covering the core structure shows little difference in the rate of transformation, with the exception of EN.39. In this steel the combination of increased nickel and chromium produces an appreciable shift of the diagram to the right. In addition a pronounced area of austenite stability appears in the temperature range of 550/580°C.

In comparing the diagrams for the case and core of each steel, it is found that, with EN.36 and EN.39, there is sufficient time available between the finish of transformation in the lower bainite region for the core and the corresponding time for the commencement of transformation of the case, to permit, with a carburised specimen, isothermal transformation of the core, followed by either isothermal transformation of the case or quenching of the case to form martensite. From a practical point of view, the isothermal transformation of the case is of little interest. The length of time required for such a transformation is excessive, and a relatively low hardness results. With EN.33 there is too much overlap of the case and core diagrams to permit isothermal transformation of the core without transformation of the case. A similar picture is found for EN.37 except that a sufficient difference in transformation rate occurs between case and core from 550 to 650°C to permit of the completion of core transformation before the commencement of case transformation. The hardness of the core transformation products produced in this temperature range, however, is low and good mechanical properties cannot therefore be expected.

Referring to the high temperature parts of the diagrams, it will be noticed that except for EN.33 and EN.36, the transformation times are too long to permit of the isothermal annealing of the case structures in practice.

Finally the martensite start temperature for the case is always well below the martensite finish temperature for the core of the corresponding steel.

From a general comparison of the diagrams, the most interesting possibility would appear to be the isothermal transformation of the core to a lower bainitic structure, having a relatively high tensile strength, immediately followed by martempering of the case to a high hardness. It was considered that such a composite treatment should be possible with the chromium containing EN.36 and EN.39.
7. Possible Heat-Treatment Procedures.

A comparison was made of a combined core austempering, case marquenching treatment with the more conventional hardening techniques. The comparison was made both with and without refining of the core. At the same time the effect of varying austenising temperatures was checked. This was in order to explore the possibility of direct quenching from a high carburising temperature; the advantage of a high carburising temperature being, of course, rapid carburising rate. Similarly the effect of oil quenching and water quenching was investigated. Oil quenching is normally used for such steels in practice, while water quenching was used in the metallographic investigations of isothermal transformations.

The test pieces used to explore these alternative heat-treatment procedures metallographically were taken from the 1 inch diameter bar material. A number of ½ inch long pieces were parted off. These were then carburised at 900°C for 2 hours; one group was pack-carburised in activated charcoal, while a further group was carburised in 50/50 cyanide salt. Mechanical properties were checked using micro physical test pieces machined from the ¼ inch diameter material. These were blank carburised for 2 hours at 900°C. The mechanical and metallographic test pieces so prepared were then subjected to the heat-treatment procedures given in Table 2. The resulting metallographic structure and hardness of the ½ x 1 inch diameter pieces was then checked together with the yield, ultimate tensile, elongation, and hardness of the micro physical test specimens.
S. Mechanical Properties.

The mechanical properties obtained for the micro test pieces are recorded in Table 3. The various heat-treatments are designated by numbers 1 to 9, corresponding to Table 2. The 1 inch diameter by $\frac{1}{4}$ inch test pieces were ground on one flat face in order to completely remove the case. Hardness checks on the core of these specimens gave very similar results to those obtained on the tensile test specimens. The hardness of the case of the 1 inch diameter by $\frac{1}{4}$ inch test pieces are included in Table 3.

The effect of austenising temperatures can be assessed by comparing the results for heat-treatments 1 and 3, 5 and 6, and 7 and 8. Inspection of Table 3 shows that very similar properties are obtained by austenising at either 900°C or 865°C. It should therefore be possible to quench direct from a carburising temperature of 900°C. A comparison of oil versus water quenching is given in the results for heat-treatments 1 and 2. No significant differences were found even with EH.58, the fastest transforming steel. This is due, no doubt, to the shift to the right of the transformation diagram with slower rates of cooling.

There are several points of interest to note with regard to the properties obtained by isothermal transformation of the core. While with EN.35, transformation of the core cannot be achieved without also almost complete transformation of the case at the same time, this is not so with the other steels, in all of which transformation of the core can be completed before transformation of the case begins. Comparing the behaviour of EN.35 with EN.57, an increase in the nickel from 5.4% to 4.7% is thus sufficient to retard the transformation of the case between 450°C and 500°C until the core transformation is complete. The initial properties obtained for EN.36 and EN.39 transformed at 500°C were found to be due to incomplete transformation. Further test pieces held in the transforming bath for 15 minutes instead of the original 7 minutes, were found to be fully transformed. The properties obtained are given in brackets in Table 3. This behaviour is not completely in accordance with the diagram obtained. The increase in transformation time required at 500°C is explicable by the usual shift of the diagram to the right with slower cooling rates. This does not explain, however, the behaviour at 450°C where a comparable shift was not found. Comparison of the properties obtained by transformation at 500°C and at 450°C shows an improvement with the lower transformation temperature.
The difference is, however, of little practical significance.

With the three steels, EN.36, 37, and 39, in which isothermal transformation of the core, followed by marquenching of the case has been achieved, the introduction of extra alloying elements has a marked effect. The ultimate tensile of approximately 50 tons per sq., inch achieved with EN.33 and EN.37 is significantly improved by the addition of chromium. The addition of around 1% chromium to the 3% of nickel of EN.33 gives an ultimate tensile of approximately 67 tons per sq., inch for EN.36. The addition of 1% chromium to the 4% nickel of EN.37, gives an ultimate tensile of around 73 tons per sq., inch for EN.39. This value for EN.39 is comparable with the properties obtained by direct hardening.

The hardness values obtained for all the specimens are plotted against the corresponding ultimate tensile in Fig. 22. It is seen that the hardness-tensile strength relationship is similar, irrespective of whether the steel has been isothermally transformed or direct hardened.
9. Dimensional Changes.

In addition to correlating physical properties with the various heat-treatment procedures suggested by the isothermal transformation diagrams, it was felt to be important to obtain a better understanding of the dimensional changes occurring throughout the heat-treatment cycles. Apart from the effects of dimensional changes on general distortion of components, heat-treatments resulting in large differences in dimensional change between case and core in a carburised component, could lead to high stresses developing between the case and core. The result of such a condition would be an increased tendency to spalling or cracking of the case.

Three separate procedures were used for this part of the investigation: (a) Dilatometric, (b) Direct measurement on rods, (c) The use of a special combination test piece.

Dilatometric Data.

The hand dilatometer was used with the standard strip specimens. First the changes occurring during the slow heating up to the austenising temperature, the contraction occurring on quenching to the transformation temperature, and the expansion on transformation, were determined. It was not possible to continue the determination to ascertain directly the contractions of the transformation products on cooling to room temperature from the transformation temperature. Inaccuracies were found due to the solidification of the adhering lead from the transforming bath. The transformed specimens were therefore removed from the dilatometer suspension, and after cleaning, remounted, and the expansion-contraction curve for the transformed product determined in an air furnace. The composite plot of dimensional change is given in Figs. 23 to 30.

Considering this plot in detail with reference to Fig. 23, the curve AB represents the expansion on heating from room temperature up to $A_1$, i.e. the expansion of alpha iron. This curve also approximates very closely to the curve obtained on reheating the transformed dilatometer specimens. $D E$ represents the contraction occurring on quenching to the appropriate transforming temperature. This is, of course, the contraction in the austenitic condition. The expansion occurring on transformation is represented by the vertical lines such as $C H$. The dotted line $E F$ represents the expansion which occurs if the specimen is quenched to a low enough temperature for
martensite to be formed.

Comparing the diagrams obtained for both the case and core of the four steels investigated, it was found that the expansion curves from room temperature up to \( A_1 \) differ very little from one another. Similarly the amount of contraction due to the alpha gamma change is of the same order for all the steels. When comparing the further expansion which occurs up to the austenising temperature, however, it is found that the rate of expansion differs markedly depending upon whether the carbon content has been increased by carburising or not. The increased rate of expansion found in the high carbon compositions continues up to the ACM temperature, after which the rate approximates to that found for the low carbon compositions and the contraction rate in the austenitic condition. This increase in the rate is due to the superimposition of the expansion due to carbon dissolving in the austenite upon the expansion of gamma iron. The effect of carbon content on the rate of expansion above \( A_1 \) is plotted in Fig. 31. The rate obtained with a steel of composition, carbon 0.4, manganese 0.5, silicon 0.25, nickel 4.1, chromium 1.3, is included in order to check the effect of intermediate carbon content.

These results show that when austenising a carburised component there is a tendency to develop tensile stresses between the case and core, these stresses being greater the larger the difference in carbon content between the case and core, and the higher the austenising temperature used up to the ACM for the case.

It is also found that on quenching to the transforming temperature, all the steels investigated showed a common rate of contraction in the austenitic condition. On transforming, the amount of expansion found is sufficient in all cases to offset the difference in expansion/contraction rates for alpha and gamma iron and the extra expansion resulting from the solution of carbon in alpha iron. This indicates that during isothermal transformation, the whole of the carbon is precipitated as carbide. This precipitation is accompanied by a contraction that reduces the expansion due to the gamma-alpha change. Because of this, at any given transformation temperature, the higher the carbon content, the smaller the total expansion observed on transformation. (Provided all the carbon has been taken into solid solution during austenising). The net result is that after transformation both the case and the core compositions have returned to the original dimensions which they exhibited at the same temperature before austenising.

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These results agree with the findings of Russell and Mavrocordatos (14) with the exception that they found that when transformation occurred in the lower ranges of temperature, an increased expansion is produced. This is presumed by these authors to be due to all the carbon not being precipitated. This effect would appear to become noticeable below a temperature of 350°C. Our investigations barely extended into the beginning of this range and this no doubt accounts for this effect not being observed.

It was found that the expansion-contraction curve for the specimens after transformation approximated to the curve obtained during the initial heating for austenising. From this it would seem in order to assume that after undergoing isothermal heat-treatment involving transformation at a temperature above 350°C, both the case and core of these steels will occupy the same volume as originally. This will result in the development of the minimum of stress between the case and core.

The dimensional changes resulting from transformation to martensite are also included in Figs. 23 to 30. Inspection of these figures shows that an appreciable increase in volume over the original volume occurs, the increase for the case being roughly double that for the core. It is therefore to be expected that with direct quenching methods of hardening, an appreciable tensile stress will be developed between the case and core. Further investigations using the two procedures following were undertaken in order to obtain confirmation of the above findings and at the same time try to obtain more quantitative measurements of the stresses involved.

Values obtained by direct measurements on tubes.

It was decided to obtain a more direct measurement of the dimensional changes resulting from heat-treatments by directly measuring the change in dimension of a specimen before and after the appropriate heat-treatment procedure. For this approximately 2 inch lengths of the ½ inch diameter stock were used. These were bored out to a diameter of 5/32 inch, leaving a tube of wall thickness approximately 3/64 inch. A suitable number of these specimens for each steel were then pack carburised right through to represent the case, while the remainder were blank carburised and used to represent the core. All the specimens were then nickel plated to a thickness of 0.0007 inch to prevent decarburisation during subsequent heat-treatment. Prior to nickel plating, the ends of each specimen were surface ground parallel to each other. After plating these
same ends were lapped in order to ensure accuracy for measurement. The lengths of the specimens were then carefully measured before and after heat-treatment. The heat-treatment consisted of austenising in a neutral salt bath for 30 minutes at 900°C followed by either quenching in oil, or quenching into a suitable transforming bath and holding for a sufficient time for complete transformation to occur at the appropriate temperature. The results are recorded in Table 4. There is a small increase in length in all the isothermally transformed specimens. In the case of the oil quenched specimens, there is an appreciable increase in length due to martensite formation. The core compositions all show approximately the same increase in length of about 0.0025 inches per inch. This is about five times the expansion of both the case and core when isothermally transformed, some of these isothermally transformed tubes were subsequently mounted in a dilatometer and a plot obtained of expansion with rise in temperature. The results are shown as a dotted curve in Figures 23 to 30. From the appearance of these curves it would seem that the small expansion found on isothermally transforming this type of specimen is in fact due to quenching stresses, these stresses being released on subsequent heating. Oil quenching of the case to martensite results in an even greater expansion. This varies, however, being progressively less with increasing alloy content of the steel, due to increasing amounts of retained austenite.

These results indicate that the amount of stress developed between the case and core will be at a minimum when the case and core are both isothermally transformed, and at a maximum when the core is isothermally transformed followed by marquenching of the case. Direct quenching of the case and core together should give intermediate values of stress.

The combination test piece.

In order to more directly assess the stress developed between the case and core, a combination test piece was developed. This consisted essentially of a bi-metal ring, the outer periphery being the carburised case, the inner portion being the core composition. The test pieces were produced as follows. A 6 inch length of each of the 1 inch diameter bars was obtained and bored out to form a tube of ½ inch wall thickness, the outside diameter having been ground to remove scale and produce a good finish. The ends of these tubes were then closed by welding on mild steel plates. In this way the bore of the tubes was protected from carburising. The samples were then packed
in carburising compound and carburised for 8 hours at 900°C. This gave a case depth of approximately 1/16 inch. When cool, the tubes were slip up on an elastic wheel to form rings a little over 1/2 inch thick. Each ring was then split at one point, giving a gap of approximately 1/16 inch. The rings were then nickel plated to a thickness of 0.0007 inch. This type of specimen is illustrated in the centre of Fig. 8 and consists of approximately equal concentric layers of case and core composition.

The procedure was then to carefully measure the diameter of the rings at right angles to the gap, to subject them to the appropriate heat-treatment, and then to repeat the measurement. Any increase in diameter would indicate a greater expansion of the core than the case, a decrease in diameter indicating a greater expansion of the case than the core. Initial experiments did not give the consistency expected. With attention to detail, however, and in particular by ensuring that the specimens were fully stress relieved before the initial measurement, and always quenched horizontally, the behaviour pattern expected was obtained. The results are given in Table 5.

These results confirmed that the maximum stress is developed between the case and core when isothermal transformation of the core is followed by marquenching of the case. This stress in these carburising steels is a tensile stress. The minimum stress is developed by isothermal transformation of both case and core.
It was noticed early in this investigation that with the core compositions, good correlation was obtained between the onset and progress of transformation, and both hardness and micro-structure of the quenched specimens. A fundamental difference was found, however, with the carburised compositions within certain ranges of transformation temperatures. On plotting the hardness of the specimens, quenched after varying times in the isothermal holding bath, it was found that, not only did the hardness depend on the isothermal holding temperature, but it did not remain constant up to the commencement of transformation. Instead a progressive increase in hardness occurred, in some cases a rise of up to 150 points VPN being observed. The maximum hardness reached always coincided very closely with the commencement of transformation. The hardness then proceeded to fall in the usual way due to transformation. In order to fully illustrate this behaviour, the hardness figures obtained on the isothermal transformation specimens have been plotted on the summary diagrams, Figure 32 to 39. It was found difficult to clearly illustrate this phenomenon graphically, and it is felt that these summary diagrams are the best approach. They are, in fact, perspective diagrams. The vertical and horizontal axes of the diagrams are VPN hardness and logarithmic time in seconds respectively. The angular or perspective axis is the transformation temperature. The commencement of transformation is shown by the full arrows. Pins denote the end of transformation, while the tailed arrows denote the commencement of the pearlite reaction. By joining the full arrows, the slightly distorted "S" curve shown dotted is obtained.

It was evident that this phenomenon was more pronounced with the chromium containing EN.36 and 39. The investigations were therefore extended to include a series of three hot work steels containing about 5% chromium and varying amounts of tungsten and vanadium, both the normal and carburised compositions being considered. The composition of these steels is given in Table 6. These particular steels were chosen because they had previously been observed to show marked secondary hardening on tempering as well as depression of hardness with decreasing cooling rates. As a preliminary the isothermal transformation diagrams were determined by the metallographic procedure, using an austenising temperature of 1000°C. The six diagrams obtained are given in Figures 40 to 45. Perspective summary diagrams were then prepared as before. These are reproduced in Figures 46 to 51. It will be seen that all but the 1% tungsten core composition exhibit a range of transformation
temperature in which a rise in the hardness of the quenched specimens precedes the breakdown of austenite.

The simplest explanation of this phenomenon would be to consider that the increasing hardness of the martensite produced on quenching reflects the decreasing stability of the austenite at the transformation temperature. It is evident that increasing carbon and carbide formers enhance the effect. In Figures 52 and 53 the quenched hardness obtained after holding for 5 seconds at the transforming temperature is plotted for the case of the carburising steels and the case and core of the hot work steels. With the chromium containing carburising steels EN 36 and 39 (case compositions) the lowering of the as-quenched hardness commences at just below 400°C, reached a maximum at about 550°C, and then decreases again. With the hot work steels the picture is similar, but the depression in hardness occurs over a much wider range, and is more marked. It is also enhanced by carburising and particularly by increasing vanadium.

This increase in hardness is what would be expected from stabilisation of the austenite resulting in retained austenite on quenching. It is now well established that thermal stabilisation of austenite can occur above Ms. According to Morgan and Ko (29) the order of time required for full stabilisation is very low, certainly less than 5 seconds at a temperature as low as 350°C, and lower still at higher temperatures. It is therefore understandable that the full stabilisation will have occurred before the quenching of the first (2 seconds) specimens.

Considering the process of transformation to be made up of three stages, incubation, nucleation, and growth or observable transformation, it is reasonable to assume that the process responsible for the decrease in stability of the austenite that eventually results in isothermal transformation, i.e. nucleation, is also responsible for counteracting the thermal stabilising effect. Two observations are put forward in support of this. Investigation of the quenched structures produced in the case of the carburising steels when the maximum rise in pre-transformation hardness was observed, showed that, not only was the martensite progressively darker etching, but also that the martensite needles became progressively much finer and more numerous. This is illustrated in Micros Nos. 9 to 11. Similar observations on the core compositions of the vanadium containing hot work steels, were initially inconclusive owing to the difficulty of resolving the martensite. A combination of relief polishing and very light etching, however, showed indications of a very fine precipitate. This was quite distinct from the normal undissolved carbides. As the start of transformation was approached, this precipitate appeared to coalesce until transformation commenced. This is illustrated in Micros 12 to 15.
In order to more clearly show the hardness changes observed prior and during isothermal transformation, models were made to illustrate the three basic types of behaviour. These are illustrated in Figures 54 to 56. The first shows the typical behaviour of a steel that does not show "pre-transformation" changes. The second shows this behaviour very markedly, while the third model shows not only "pre-transformation" changes, but also "post-transformation" changes in the lower transformation range, i.e. secondary hardening.

It would appear that the conditions of sufficient carbon and carbide forming elements necessary for the "pre-transformation" changes to occur are also conducive to secondary hardening effects. It is felt that these two processes are related, and it is suggested that complex carbides are formed in the upper ranges of temperature prior to austenite breakdown. This will result in the effective concentration of carbon and alloying elements in the austenite being reduced, thus both inducing the onset of transformation and raising the Ms temperature. This would account for the "pre-transformation" changes and absence of "post-transformation" changes or secondary hardening effects with the higher transforming temperatures. In the lower ranges of temperature it is considered that the diffusion of the alloying elements is so slow that iron carbide is the first result of transformation, a carbide rearrangement subsequently occurring with an increase in hardness due to lattice distortion. This results in the secondary hardening effects found in this temperature range.
11. Summary and Conclusion

The original aims of the investigation have been achieved, in as much as the isothermal transformation diagrams for the case and core compositions of four representative carburising steels have been determined and the possibility of satisfactory isothermal heat-treatment techniques for these steels explored. It was possible to devise suitable isothermal techniques for three of the four carburising steels, namely EN 56, 57, and 59. The dimensional changes occurring in both the case and core compositions as a result of the various alternative heat-treatment procedures, have also been determined quantitatively, and a qualitative assessment made of the case/core stresses resulting from these dimensional changes.

The isothermal transformation diagrams have been determined for 14 steel compositions. In addition to the diagrams for the case and core compositions of the four carburising steels, EN 33, 35, 37, and 39, the normal and carburised composition of a series of three hot work steels have been included. Comparison of the transformation diagrams obtained for the core of the carburising steels, shows little difference with the exception of EN 39, in which an appreciable shift to the right occurs, together with pronounced austenite stability in the temperature range 550-580°C. There is a considerable overlap of the case and core diagrams when nickel alone is present as in EN 33 and 37. The presence of chromium in EN 36 and 39, however, eliminates this overlap in the lower bainite range. From a comparison of the diagrams, the most suitable heat-treatment, apart from the normal direct hardening methods, would appear to be isothermal transformation of the core in the lower bainite region, followed by marquenching of the untransformed case. The physical properties obtained on test pieces indicate that useful core properties and reasonable case hardness are obtained by this method in section sizes up to at least ½ inch, and with all the steels except EN 33. With EN 33 adequate case hardness is not obtained.

The dimensional changes accompanying the direct quenching and isothermal hardening of the case and core of the carburising steels have been followed. With isothermal transformation above 350°C, the original room temperature dimensions are maintained after treatment. With direct quenching an increase in size occurs dependent upon carbon content and retained austenite. The differences in the dimensional changes occurring in the case and core with various treatments and resulting in stress between the
case and core have been confirmed by the heat-treatment of special split ring specimens. These specimens had an outer portion of the case composition and an inner portion of the core composition. It is concluded that if minimum distortion of the component as a whole is required, this will be best achieved by isothermal transformation of the core followed by marquenching of the case. The minimum of case/core stress, together with optimum mechanical properties will be developed by direct or marquenching of the core and case together. Isothermal transformation of the core followed by marquenching of the case results in the development of the maximum stress between the case and core.

The hardness changes occurring in the quenched structures as a result of isothermal holding and transforming have been carefully investigated for these steels. It has been found convenient to plot them on a composite diagram. These diagrams bring out clearly three main types of behaviour exhibited by the steels examined. The first type is conventional, exhibiting a constant as quenched hardness at all transforming temperatures, up to the onset of isothermal transformation, when a progressive drop in hardness occurs with the presence of increasing amounts of softer transformation product, and less martensite from the undecomposed austenite. When sufficient carbon and carbide formers are present, however, a second type of behaviour shows itself. In the higher ranges of transformation temperature the as quenched hardness is initially depressed, gradually increasing, however, and reaching a maximum coinciding with the onset of transformation. The third type of behaviour takes the form of a progressive rise in the as quenched hardness, which occurs after the completion of transformation in the lower temperature ranges, in addition to the pretransformation changes shown at the higher temperatures. This third type of behaviour is shown by certain of the hot work steels.

The results obtained for the co-efficient of thermal contraction of austenite agree with the values reported by Allen Pfeil and Griffiths (7). In addition the findings of Hickley and Woodhead (15) have been confirmed, that the rate of nucleation in the ferrite range at constant temperature decreases with time and increases with decreasing temperatures.

The simultaneous formation of alpha iron from austenite by both coherent and incoherent transformation observed by Ko (20) has been confirmed. Similarly the formation of bainite earlier at free surfaces than in the mass also observed by Ko (20) has been confirmed. This phenomenon is
particularly marked with the hot work steels.

The morphology observed for the upper bainite in the carburising steels, in particular the lack of carbide precipitate between the plates, confirms the findings of Wever, Rose, and Peter (26, 27, 28) that at high transformation temperatures, the diffusion rate of carbon is high enough for it to diffuse out of the alpha iron into the surrounding austenite. Carbide precipitation is found between the plates at lower transformation temperatures.

The findings of Russell and Mavrocordatos (14) that all the carbon is precipitated as carbide during isothermal transformation in all but the lower ranges of temperature is confirmed. The lower ranges of temperature were insufficiently explored by us to disagree with this author's findings, that carbon is at least in part retained at lower temperatures.

The findings of Morgan and Ko (29) that thermal stabilisation of austenite occurs above Ms after very short holding times, appears to have been confirmed, and is presumed to account for the depression of as-quenched hardness of certain of the steel compositions investigated. The subsequent decrease in stability of the austenite which results in the progressive rise in as quenched hardness up to the onset of transformation has not been reported before, however. Metallographic investigation of these pre-transformation changes in the austenite have indicated certain structural changes. With the case compositions of the carburising steels exhibiting this effect, a progressive increase in the fineness of the martensite needles occurs up to the onset of transformation. Careful etching of the hot work steels, has indicated a very fine precipitate which progressively agglomerates up to the start of transformation.

While it has not been possible to give a positive explanation of the hardness changes found before the onset of transformation in some of the compositions investigated, they have been shown to be associated with the presence of carbon and carbide forming elements, and the suggestion has been made that the effect is similar to "secondary hardening". It is suggested that when holding in the upper ranges of transformation temperature, prior to the onset of transformation, diffusion of alloying elements occurs, and complex carbides are formed. Because of this association of alloying elements and carbon, their effective concentration in the austenite is progressively reduced, thus inducing the onset of transformation and raising the Ms temperature.
This would result in a progressive reduction in the amount of retained austenite occurring after quenching specimens held for increasing times at the transformation temperature. It is further suggested that the "secondary hardening" shown by these steels, either on suitably tempering after direct hardening, or on prolonged holding at the lower transformation temperatures, is due to the formation of the same complex carbides. At these lower temperatures, however, it is considered that the diffusion of the alloying elements is so slow that iron carbide is initially formed on transformation, a carbide re-arrangement subsequently occurring with an increase in hardness due to lattice distortion.
#### TABLE 1.

**Composition of the materials used.**

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<th>Steel</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
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<th>Ni</th>
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Heat-treatment Procedures Used.

1. Austenise $\frac{1}{2}$ hour at 900°C. Oil quench. Temper $\frac{1}{2}$ hour at 200°C.

2. Austenise $\frac{1}{2}$ hour at 900°C. Water quench. Temper $\frac{1}{2}$ hour at 200°C.

3. Austenise $\frac{1}{2}$ hour at 865°C. Oil quench. Temper $\frac{1}{2}$ hour at 200°C.

4. Austenise $\frac{1}{2}$ hour at 865°C. Oil quench. Reheat to 780°C for 15 minutes. Oil quench. Temper $\frac{1}{2}$ hour at 200°C.

5. Austenise $\frac{1}{2}$ hour at 900°C. Transform the core at 500°C. (7 minutes) Oil quench. Temper $\frac{1}{2}$ hour at 200°C.

6. Austenise $\frac{1}{2}$ hour at 865°C. Transform the core at 500°C. (7 minutes) Oil quench. Temper $\frac{1}{2}$ hour at 200°C.

7. Austenise $\frac{1}{2}$ hour at 900°C. Transform the core at 450°C (7 minutes) Oil quench. Temper $\frac{1}{2}$ hour 200°C.

8. Austenise $\frac{1}{2}$ hour at 865°C. Transform the core at 450°C (7 minutes) Oil quench. Temper $\frac{1}{2}$ hour at 200°C.

9. Austenise $\frac{1}{2}$ hour at 900°C. Transform the core at 450°C (7 minutes) Oil quench. Reheat to 780°C for 15 minutes. Oil quench. Temper $\frac{1}{2}$ hour at 200°C.
### TABLE 5.

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</table>

2. Austenised at 900°C. Transformed at 450°C. |          |       |            |           |                |
| En 33 Core | 2.2500  | 2.2511 | +0.0011 | +0.0006 | +0.0003 | +0.0003 |
| En 33 Case | 2.2820  | 2.2835 | +0.0015 | +0.0007 | +0.0003 | +0.0003 |
| En 36 Core | 2.2695  | 2.2710 | +0.0015 | +0.0007 | +0.0003 | +0.0003 |
| En 36 Case | 2.2513  | 2.2540 | +0.0027 | +0.0018 | +0.0012 | +0.0007 |
| En 37 Core | 2.2500  | 2.2516 | +0.0016 | +0.0007 | +0.0003 | +0.0003 |
| En 37 Case | 2.2660  | 2.2671 | +0.0011 | +0.0005 | +0.0002 | +0.0003 |
| En 39 Core | 2.1930  | 2.1939 | +0.0009 | +0.0004 | +0.0003 | +0.0003 |
| En 39 Case | 2.2160  | 2.2169 | +0.0009 | +0.0004 | +0.0003 | +0.0003 |

3. Austenised at 900°C. Oil quenched. |          |       |            |           |                |
| En 33 Core | 2.2430  | 2.2533 | +0.0053 | +0.0056 | +0.0025 | +0.0020 |
| En 33 Case | 2.2705  | 2.2807 | +0.0102 | +0.0045 | +0.0020 | +0.0017 |
| En 36 Core | 2.2515  | 2.2567 | +0.0052 | +0.0025 | +0.0015 | +0.0012 |
| En 36 Case | 2.2500  | 2.2650 | +0.0050 | +0.0020 | +0.0015 | +0.0012 |
| En 37 Core | 2.2505  | 2.2557 | +0.0052 | +0.0023 | +0.0015 | +0.0012 |
| En 37 Case | 2.2500  | 2.2535 | +0.0035 | +0.0015 | +0.0012 | +0.0012 |
| En 39 Core | 2.2703  | 2.2769 | +0.0066 | +0.0029 | +0.0015 | +0.0012 |
| En 39 Case | 2.2480  | 2.2559 | +0.0079 | +0.0035 | +0.0015 | +0.0012 |
TABLE 5.

Dimensional changes observed in the combination test pieces.

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<th>Difference</th>
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## Table 6.

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<th>P</th>
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<th>Mo</th>
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Micro 1. Start of Transformation to "Blocky Ferrite".
EN 33 transformed at 700°C.
Etch: Nital. Mag: x1000.

Micro 2. Coalescence of "Blocky Ferrite".
EN 33 core transformed at 700°C.
Etch: Nital. Mag: x1000.
Micro 3. Commencement of Acicular Ferrite Range.  
Early stage of Transformation.  
EN 33 core, transformed at 600°C.  
Etch: Nital.  Mag: x 1000.

Transformation nearing completion.  
EN 33 core transformed at 600°C.  
Etch: Nital.  Mag: x 1000.
Micro 5. Acicular Ferrite.
EN 33 core transformed at 575°C.
Etch: Nital. Mag: x 1400.

Micro 8. Lower Bainite.
Early stage of Transformation.
EN 39 core transformed at 400°C.
Etch: Ferric Chloride. Mag: x 1400.
Micro 9.  EN 39 Carburised. Quenched to 500°C.
19,900 Seconds before the Start of Transformation.
Etch: Nital. Mag: x 1400.

Micro 10.  EN 39 Carburised. Quenched to 500°C.
10,000 Seconds before the Start of Transformation.
Etch: Nital. Mag: x 1400.
Micro 12. 1% Vanadium Hot Work Steel.
Quenched to 650°C.
800 Seconds before the Start of Transformation.
Etch: Diluted Ferric Chloride. Mag: x 1400.

Micro 13. 1% Vanadium Hot Work Steel.
Quenched to 650°C.
800 Seconds before the Start of Transformation.
Etch: Diluted Ferric Chloride. Mag: x 1400.
Micro 14. 1\% Vanadium Hot Work Steel.
Quenched to 660°C.
10,000 Seconds before the Start of Transformation.
Etch: Diluted Ferric Chloride. Mag: x 1400.

Micro 15. 1\% Vanadium Hot Work Steel.
Quenched to 650°C.
Transformation Commenced at 20,000 Seconds.
Etch: Diluted Ferric Chloride. Mag: x 1400.
Fig. 1a Diagram of Dilatometer used by Davenport & Bain in 1930.
Fig. 1b Diagram of Dilatometer used by Allen, Pfeil & Griffiths
Fig. 2a. General arrangement of the high speed dilatometer.
Fig. 2b. Diagram of High Speed Dilatometer.
Fig. 3a. Position of furnaces during austenising.
Fig. 5b. Position of furnaces during transformation.
Fig. 4a. Dilatometer specimen mounting.
Fig. 4b. Diagram of Specimen Mounting for High Speed Dilatometer.

- Inner Silica Rod Actuating Dial Guage.
- Outer Silica Tube Cemented into Dial Gauge Clamp.
- Silica Rods.
- Specimen.
Fig. 5. Automatic recording controller.
Fig. 6. The hand dilatometer.

Fig. 7. General arrangement of the hand dilatometer in use.
Fig. 8. The various types of specimen used.
**Fig. 9. ISOTHERMAL CURVES**

**TYPE OF STEEL**

**En 33 Crome.**

**ANALYSIS (Weight%)**

<table>
<thead>
<tr>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
<th>P</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.12</td>
<td>0.45</td>
<td>0.21</td>
<td>0.03</td>
<td>0.02</td>
<td>3.4</td>
<td>0.20</td>
<td>-</td>
</tr>
</tbody>
</table>

**Grain Size (A.S.T.M.)**

<table>
<thead>
<tr>
<th>McQuaid-Ehn</th>
<th>As Quenched</th>
</tr>
</thead>
</table>

**PREVIOUS TREATMENT**

**AUSTENITIZING CONDITIONS**

- 900 °C for 30 MINS

**METHOD OF DETERMINATION**

Dilatometric and Metallographic.

**METHOD OF DETERMINATION**

**Degrees Cent.**

**Degrees Fahr.**

**TEMPERATURE DEGREES CENT.**

**DEGREES FAHR.**

**TIME HELD IN CONSTANT TEMPERATURE BATH FROM START OF QUENCH SECONDS**

**Ms (0 °C)**

10%

100% Transformation.

**Mf (Calc)**

340
**Fig. 16. ISOTHERMAL CURVES**

**TYPE OF STEEL**

| En 33 Case. |

**ANALYSIS (Weight%)**

<table>
<thead>
<tr>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
<th>P</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.18</td>
<td>0.45</td>
<td>0.22</td>
<td>0.03</td>
<td>0.02</td>
<td>3.4</td>
<td>0.20</td>
<td>-</td>
</tr>
</tbody>
</table>

**Grain Size (A.S.T.M.)**

McQuaid-Ehn As Quenched

**PREVIOUS TREATMENT**

900 °C FOR 30 MIN

**AUSTENITIZING CONDITIONS**

**METHOD OF DETERMINATION**

Dilatometric and Metallographic

**PROCESS CAST NO.**

**AUSTENITIZING CONDITIONS**

900 °C FOR 30 MIN

**METHOD OF DETERMINATION**

Dilatometric and Metallographic
Fig. 11. ISOThermal CURVES

<table>
<thead>
<tr>
<th>TYPE OF STEEL</th>
<th>PROCESS</th>
<th>CAST NO.</th>
</tr>
</thead>
<tbody>
<tr>
<td>En 36 Cire.</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>ANALYSIS (Weight%)</th>
<th>Grain Size (A.S.T.M.)</th>
</tr>
</thead>
<tbody>
<tr>
<td>C</td>
<td>Mn.</td>
</tr>
<tr>
<td>0.10</td>
<td>0.43</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>PREVIOUS TREATMENT</th>
<th>AUSTENITIZING CONDITIONS</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>900°C FOR 30 MIN</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>METHOD OF DETERMINATION</th>
</tr>
</thead>
<tbody>
<tr>
<td>Dilatometric and Metallographic.</td>
</tr>
</tbody>
</table>

![Isothermal Curves Diagram](image-url)

<table>
<thead>
<tr>
<th>TIME HELD IN CONSTANT TEMPERATURE BATH FROM START OF QUENCH</th>
</tr>
</thead>
<tbody>
<tr>
<td>SECONDS</td>
</tr>
</tbody>
</table>
# Isothermal Curves

## E36 Case

### Analysis (Weight %)

<table>
<thead>
<tr>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
<th>P</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.94</td>
<td>0.42</td>
<td>0.14</td>
<td>0.04</td>
<td>0.04</td>
<td>364</td>
<td>130</td>
<td>McQuaid-Ehn As Quenched</td>
</tr>
</tbody>
</table>

### Previous Treatment

- **900 °C for 30 Mins**

### Method of Determination

- Dilatometric and Metallographic

### Diagram Details

- **TEMPERATURE DEGREES FAHR.**
  - 0 to 1,600
- **DEGREES CENT.**
  - 0 to 1,600
- **TIME HELD IN CONSTANT TEMPERATURE BATH FROM START OF QUENCH SECONDS**
  - 0 to 10,000
Fig. 13. **Isothermal Curves**

**Type of Steel**  
E-37 Core

<table>
<thead>
<tr>
<th>Analysis (Weight%)</th>
<th>Grain Size (A.S.T.M.)</th>
</tr>
</thead>
<tbody>
<tr>
<td>C 0.12</td>
<td>Mn 0.37</td>
</tr>
</tbody>
</table>

**Previous Treatment**  
Austenitizing Conditions:  
900°C for 30 mins

**Method of Determination**  
Dilatometric and Metallographic

**Diagram**

- Various temperatures and transformation phases are plotted.
- Ms and Mc points are marked.
- Time held in constant temperature bath from 0 to 10000 hours is indicated.

**Table**

<table>
<thead>
<tr>
<th>Temperature (°C)</th>
<th>Time (hours)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>0</td>
</tr>
<tr>
<td>100</td>
<td>100</td>
</tr>
<tr>
<td>200</td>
<td>200</td>
</tr>
<tr>
<td>300</td>
<td>300</td>
</tr>
<tr>
<td>400</td>
<td>400</td>
</tr>
<tr>
<td>500</td>
<td>500</td>
</tr>
<tr>
<td>600</td>
<td>600</td>
</tr>
<tr>
<td>700</td>
<td>700</td>
</tr>
<tr>
<td>800</td>
<td>800</td>
</tr>
<tr>
<td>900</td>
<td>900</td>
</tr>
<tr>
<td>1000</td>
<td>1000</td>
</tr>
<tr>
<td>1100</td>
<td>1100</td>
</tr>
<tr>
<td>1200</td>
<td>1200</td>
</tr>
<tr>
<td>1300</td>
<td>1300</td>
</tr>
<tr>
<td>1400</td>
<td>1400</td>
</tr>
<tr>
<td>1500</td>
<td>1500</td>
</tr>
<tr>
<td>1600</td>
<td>1600</td>
</tr>
</tbody>
</table>
**Fig. 14. ISOTHERMAL CURVES**

<table>
<thead>
<tr>
<th>TYPE OF STEEL</th>
<th>PROCESS</th>
<th>CAST NO.</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>E, 37</strong> CASE</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>ANALYSIS (Weight%)</th>
<th>Grain Size (A.S.T.M.)</th>
</tr>
</thead>
<tbody>
<tr>
<td>C</td>
<td>Mn</td>
</tr>
<tr>
<td>0.79</td>
<td>0.27</td>
</tr>
</tbody>
</table>

**PREVIOUS TREATMENT**

**AUSTENITIZING CONDITIONS**

900 °C FOR 30 MINS

**METHOD OF DETERMINATION**

Dilatometric and Metallographic.

**TIME HELD IN CONSTANT TEMPERATURE BATH FROM START OF QUENCH SECONDS**

- **Ms**
- **M 10%**
- **M 50%**
- **Mf**

**TEMPERATURE**

**DEGREES FAHR.**

**TIME HELD IN CONSTANT TEMPERATURE BATH**

**FROM START OF QUENCH SECONDS**

**TEMPERATURE**

**DEGREES CENT.**

**DEGREES FAHR.**

**TRANSFORMATION %**

0 10 20 30 40 50 60 70 80 90 100
Fig. 15. **Isothermal Curves**

**Type of Steel: E734A Core**

**Analysis (Weight%):**

<table>
<thead>
<tr>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
<th>P</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.12</td>
<td>0.037</td>
<td>0.24</td>
<td>0.04</td>
<td>0.04</td>
<td>4.40</td>
<td>1.24</td>
<td></td>
</tr>
</tbody>
</table>

**Grain Size (A.S.T.M.):**

<table>
<thead>
<tr>
<th>McQuaid-Ehn</th>
<th>As Quenched</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
</tr>
</tbody>
</table>

**Previous Treatment:**

**Austenitizing Conditions:**

900 °C for 30 mins

**Method of Determination:**

Dilatometric and Metallographic.

**Temperature Degrees Cent.:**

900 800 700 600 500 400 300 200 100

**Degrees Fahn.:**

1,600 1,500 1,400 1,300 1,200 1,100 1,000 1,000

**% Transformation:**

0 10 50 90 100

**Time Held in Constant Temperature Bath:**

From start of quench to seconds.
Fig. 16. **Isothermal Curves**

**Type of Steel**

**E-79 A CASE**

<table>
<thead>
<tr>
<th>Analysis (Weight%)</th>
<th>Grain Size (A.S.T.M.)</th>
</tr>
</thead>
<tbody>
<tr>
<td>C</td>
<td>Mn</td>
</tr>
<tr>
<td>1.05</td>
<td>0.37</td>
</tr>
</tbody>
</table>

**Previous Treatment**

**Austenitizing Conditions**

900 °C for 30 mins

**Method of Determination**

Dilatometric and Metallographic.
Fig. 17: In 83 Core, High Speed Dilatometer.

(Auto Recorded at 20 frames per second)
This gives a Coefficient of contraction of

\[0.033 \times 10^{-3}\] inches per inch per °C.

**Fig. 18.** En 33 Core. High Speed Dilatometer.

Thermal contraction of Austenite on quenching into the transformation bath.
Fig. 20. Time in Seconds after Start of Transformation.
Fig. 22. Variation of Hardness with Tensile Strength.
Fig. 24  En 33 Case, Dimensional changes During Isothermal Transformation.
Fig. 87. En 37 Core. Dimensional Changes During Isothermal Transformation.
Fig. 31. The Effect of Carbon Content on the Rate of Expansion above A1.
TEMPERATURE DEGREES CENT.

Fig. 41. I SOTHERMAL CURVES

<table>
<thead>
<tr>
<th>TYPE OF STEEL</th>
<th>PROCESS</th>
<th>CAST NO.</th>
</tr>
</thead>
<tbody>
<tr>
<td>1% Tungsten Hot Work, Carburised.</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

ANALYSIS (Weight%)

<table>
<thead>
<tr>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
<th>P</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
<th>W</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.0</td>
<td>0.03</td>
<td>0.8</td>
<td>0.02</td>
<td>0.02</td>
<td>0.2</td>
<td>5.0</td>
<td>17</td>
<td>1.0</td>
</tr>
</tbody>
</table>

PREVIOUS TREATMENT

AUSTENITIZING CONDITIONS

1000 °C FOR 30 MINS

METHOD OF DETERMINATION

Metallographic.

TEMPERATURE DEGREES FAHR.

GRAIN SIZE (A.S.T.M.)

McQuaid-Ehn As Quenched
**Fig. 42. ISOTHERMAL CURVES**

**TYPE OF STEEL**

|PROCESS| CAST NO.
<table>
<thead>
<tr>
<th></th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td>1/2% Vanadium Hot Work</td>
<td>Core</td>
</tr>
</tbody>
</table>

**ANALYSIS (Weight%)**

<table>
<thead>
<tr>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
<th>P</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
<th>V</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.4</td>
<td>0.3</td>
<td>1.0</td>
<td>0.02</td>
<td>0.02</td>
<td>0.2</td>
<td>5.0</td>
<td>1.4</td>
<td>0.6</td>
</tr>
</tbody>
</table>

**Grain Size (A.S.T.M)**

<table>
<thead>
<tr>
<th>McQuaid-Ehnh</th>
<th>As Quenched</th>
</tr>
</thead>
</table>

**PREVIOUS TREATMENT**

1000 °C FOR 30 MINS

**AUSTENITIZING CONDITIONS**

**METHOD OF DETERMINATION**

Metallographic.

**TIME HELD IN CONSTANT TEMPERATURE BATH FROM START OF QUENCH SECONDS**

- 1000 °C for 30 mins
- 1200 °C for 1 hour
- 1500 °C for 10 hours
- 1600 °C for 30 hours

**TEMPERATURE DEGREES FAHRENHEIT**

- 0
- 100
- 200
- 300
- 400
- 500
- 600
- 700
- 800
- 900
- 1000
- 1100
- 1200
- 1300
- 1400
- 1500
- 1600

**DEGREES CENTIGRADE**

- 0
- 100
- 200
- 300
- 400
- 500
- 600
- 700
- 800
- 900
- 1000
- 1100
- 1200
- 1300
- 1400
- 1500
- 1600
Fig. 43. **Isothermal Curves**

**Type of Steel**

<table>
<thead>
<tr>
<th>Process</th>
<th>Cast No.</th>
</tr>
</thead>
<tbody>
<tr>
<td>1% Vanadium Hot Work. Carburised.</td>
<td></td>
</tr>
</tbody>
</table>

**Analysis (Weight%)**

<table>
<thead>
<tr>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
<th>P</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
<th>V</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.9</td>
<td>0.3</td>
<td>1.0</td>
<td>0.02</td>
<td>0.02</td>
<td>0.20</td>
<td>0.50</td>
<td>1.4</td>
<td>0.6</td>
</tr>
</tbody>
</table>

**Grain Size (A.S.T.M.)**

<table>
<thead>
<tr>
<th>McQuaid-Ehn</th>
<th>As Quenched</th>
</tr>
</thead>
</table>

**Previous Treatment**

Austenitizing Conditions

1000 °C FOR 30 MIN.

**Method of Determination**

Metallographic.

**Graph**

- Isothermal curves showing transformations at various temperatures and times.
- Time held in constant temperature bath from start of quench in seconds.

**Degrees Cent.**

- 900, 800, 700, 600, 500, 400, 300, 200, 100

**Degrees Fahr.**

- 1600, 1500, 1400, 1300, 1200, 1100, 1000, 900, 800, 700, 600, 500, 400, 300, 200, 100

**Temperature**

- 0 to 1000 °C

**Time Held**

- 0 to 10,000 seconds

**Legend**

- A: Austenite
- F: Ferrite
- C: Carbide
- F + C: Ferrite + Carbide
- H + F + C: Austenite + Ferrite + Carbide

**Notes**

- Analysis includes elements C, Mn, Si, S, P, Ni, Cr, Mo, and V.
- Grain size analysis using McQuaid-Ehn and As Quenched methods.
- Transformation of steel at various temperatures and times is illustrated.
- Temperature bath conditions for austenitizing are specified.
Fig. 44. ISOTHERMAL CURVES

TYPE OF STEEL
1% Vanadium Hot Work. Core.

ANALYSIS (Weight %)

<table>
<thead>
<tr>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
<th>P</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
<th>V</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.3</td>
<td>0.8</td>
<td>0.02</td>
<td>0.02</td>
<td>2</td>
<td>5</td>
<td>1.8</td>
<td>1.0</td>
<td></td>
</tr>
</tbody>
</table>

PREVIOUS TREATMENT

AUSTENITIZING CONDITIONS
1000 °C FOR 30 MIN

METHOD OF DETERMINATION
Metallographic.
ISOETHERMAL CURVES

TYPE OF STEEL
1% Vanadium Hot Work, Carburised.

ANALYSIS (Weight%)
C. Mn. Si. S. P. Ni. Cr. Mo. V
1.0 0.8 0.02 0.02 0.2 5.0 1.8 1.0

Grain Size (A.S.T.M.)
McQuaid-Ehn As Quenched

PREVIOUS TREATMENT
900°C
800°C
700°C
600°C
500°C
400°C
300°C
200°C
100°C

AUSTENITIZING CONDITIONS
1000 °C FOR 30 MINS

METHOD OF DETERMINATION
Metallographic.
Fig 52. Variation in Initial Quenched Hardness with Transformation Temperature. Carburising Steels. Carburised Compositions.
Fig. 54. Model of Hardness Changes in Isothermal Transformation Specimens. Behaviour Type 1. EN 33 Core.
Fig. 55. Model of Hardness Changes in Isothermal Transformation Specimens. Behaviour Type 2. EN 39 Carburised.
Fig. 56. Model of Hardness Changes in Isothermal Transformation Specimens. 
Behaviour Type 5. 1% Vanadium Hot Work Steel.
References.


6. M. A. Grossmann: Discussion of Papers by Bain & Waring.


