TRANSFORMATION PLASTICITY DURING MARTENSITE FORMATION IN CERTAIN FERROUS ALLOYS

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by

O. Alpay Ankara

Department of Metallurgy,
Royal School of Mines,
Imperial College of Science
and Technology,
London, S.W.7.

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ABSTRACT

An investigation has been made to study the plastic deformation (transformation plasticity) occurring as a result of the application of stress during martensite formation in the En 30A (HD) steel and in the iron-20 wt% nickel alloy. The results were obtained by using an apparatus in which wire specimens could be austenitized and cooled under stress in vacuum and the resulting extension could be measured. Austenite yield stress measurements were also made with this apparatus.

The extension during martensite formation was determined for a range of stresses from about 1,000-53,000 lb/in^2, applied either above or below M_s or at M_s. It has been found that the plastic yielding which takes place only within a range of about 40°C below M_s occurs for stresses much below the normal yield stress of austenite. In the lower range of stresses, the total extension obtained during transformation is found to be directly proportional to the applied stress. Transformation plasticity was also studied for a range of structural conditions of austenite, obtained by different austenitizing treatments. It has been found that increasing the austenitizing time at low austenitizing temperatures or increasing the austenitizing temperature increases the transformation plasticity. The austenite yield stress was also found to vary with austenitizing time and temperature;

the

highest strengths were associated with short time treatments at relatively low austenitizing temperatures, while lower strengths were associated with high temperature treatments.

These observations have been discussed with the predictions of a macroplasticity approach to the phenomenon of transformation plasticity.

Microscopical observations showed that martensite plates formed with preferred orientation under certain conditions of stress.

Transformation plasticity during tempering of the En 30A steel was also studied and it was found that the amount of deformation increased linearly with increase of applied stress.

CONTENTS.

CONTENTS (Contn.)

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CHAPTER I.

GENERAL INTRODUCTION

1. GENERAL INTRODUCTION

The term 'transformation plasticity' is used to denote the phenomenon of deformation occurring on the application of load, during phase transformations. It has also been referred to as 'superplasticity'. This phenomenon may be of considerable importance in the case of thermal cycling of metals and alloys through a phase transformation as well as in creep and in the cooling of steel weldments. Information on transformation plasticity might also be applied to the development of low temperature forging processes.

Most of the ductility studies during phase transformations have been made at elevated temperatures(1-9) where diffusion becomes an important factor to be considered. However, little work seems to have been done to study the ductility during martensitic (diffusionless) transformations.(3,35) The investigations on transformation plasticity have shown that the mechanical weakness of various crystalline materials during phase transformation can be correlated with the difference in the specific volume of parent and product phase at its transformation temperature. Owing to the difference in densities between parent and product phase, some internal stresses are set up during transformation and the material yields at very low external stress because of the superimposition of the external stress on the high pre-existing internal ones. Thus, yielding is expected to occur when the combined external and internal stresses exceed the yield stress of the weaker phase.

The investigation reported here concerns transformation plasticity occurring during martensite formation in a medium alloy nickel-chromium-molybdenum steel and on an iron-nickel alloy with M_s temperatures of about 300°C and 250°C respectively. The medium alloy steel was selected on the basis of its high hardenability, which enabled martensite to form during free cooling in vacuum. The iron-nickel alloy was selected as a carbon free alloy with an M_s temperature approximately the same as that of the steel, so that comparison of the transformation plasticity effects could be made.

The present investigation has been made to study the influence of various important factors which may throw light on the underlying mechanism of transformation plasticity. Experiments were therefore made for a range of stresses applied at various temperatures and for a range of austenitizing conditions which produced various structural conditions of austenite. Yield stress measurements of austenite were also made and this provided the data useful for relating the extent of transformation plasticity to the yield stress of austenite.

CHAPTER 2.

LITERATURE SURVEY

2. LITERATURE SURVEY

I. Introduction.

While undergoing phase transformation, metals and alloys are mechanically weak, and plastic deformation can occur under stresses much below the normal yield stress of the parent phase; the term 'transformation plasticity' is sometimes used to denote this phenomenon, and it has been known for many years.

Since most of the phase changes occur at elevated temperatures, the increased ductility obtained during an allotropic transformation has been regarded primarily as a high temperature phenomenon. Numerous reports of mechanical weakness in metals and alloys undergoing allotropic transformation are available in the literature.(1-9) Many methods have been used to tackle the problem from different angles by measuring the change in hardness or by performing torsion tests, tensile tests and compressive tests. Furthermore the occurrence of superplasticity is not confined to metallic materials, but it seems that in principle it should also occur in other crystalline structures, e.g. ceramic oxides.

In the earliest ductility studies, the ductility of iron and iron base alloys ~~were~~ was investigated by torsion or tension methods. For example, Clark and Russ(10) and Ihring(11,12) studied the ductility characteristics of iron base alloys by a torsion method. In these experiments there was a temperature gradient along the specimen, and undoubtedly this temperature gradient resulted in the occurrence of the $\alpha \rightleftarrows \gamma$ transformation in a certain portion of a specimen and this was in fact in the regions where maximum weakening occurred. However, the authors at that time did not take this most important factor into account, and no attempt was made to consider the effect of the phase transformation.

Later, Guessier and Castro studied the hot ductility of ferrous materials in torsion(14) and attributed the high value of plasticity to the continuous recrystallization during deformation. There are many references to the anomalous increase in ductility during high temperature creep in the literature(15-20). In this line the work of Rohnhanau(21) on pure iron, pure nickel and on nickel alloys with iron, chromium and molybdenum has also shown that the metals were very weak during recrystallization, and cold deforming the metals and alloys reduced their strength at the temperatures where recrystallization was taking place. However, after an annealing treatment above the recrystallization temperature, if the specimens were reheated to the same recrystallization temperatures, the strength of the metals and alloys was much higher than in the previous cases.

Again by applying torsion, Brandsma and Lips(15) tested 20% cold-deformed pure aluminium specimens and found that the angle of twisting suddenly increased at 300°C with the start of recrystallization and at 455°C, the value of twisting decreased as the metal was fully recrystallized. The weakening of the aluminium which occurred during recrystallization was decreased greatly by annealing the metal above the recrystallization temperature. Brandsma and Lips also showed that cold deforming the materials increased their "amorphous" plasticity, i.e. the ductility occurring during recrystallization.

A very similar experiment on amorphous plasticity was also done by Wassermann(22-23) on annealed and cold drawn copper wires. The experiments showed that, under the same condition and amount of stress, the amount of amorphous plasticity was much higher for cold drawn wire than for annealed wire. It is noteworthy that in simple recrystallization there is no question of an allotropic transformation being involved.

II. Transformation plasticity during the alpha to gamma transformation in iron and iron alloys:

The work of Albert Sauveur in 1924 on iron and steel was one of the earliest studies in the field of transformation plasticity(24). Tensile and twisting stresses were applied to iron and steel bars of 1/4" square section, heated at the middle to a predetermined temperature (between 600 and 1200°C), the temperature falling gradually towards both ends of the bars. Results were obtained indicating that as specimens were heated in the middle to ~~the~~ temperatures of above the transformation point, they underwent twisting or elongation and final rupture at two places situated symmetrically on either side of the most intensely heated middle region where the temperature was that of the A_{c3} point, e.g. 910°C for carbonless iron and 770°C for 0.3% carbon steel. However, if the specimen was heated in the middle to temperatures below ~~inferior to~~ the A_{c3} point, it underwent twisting or contraction and final rupture at or near the centre where the temperature was at a maximum, i.e. where the α-phase has been heated to the highest temperature. A similar observation was also made by Michael and Izina(25). Using high temperature metallography to find the influence of cyclic fluctuations of temperature on the character of the process of deformation in Armco iron containing 0.03%C, they observed the superplasticity. The specimens which they used were heated by an electric current, and a steep temperature gradient was developed along the specimen, with the highest temperature occurring at the middle portion of the specimen. By cyclically heating the middle zone of the specimen to 1000°C and cooling to 800°C, two necks appeared on the specimen instead of one. These two necks were situated symmetrically on either side of the most intensely heated middle region. The temperature fluctuation, in the zones where the necks are formed, was between 720 and 850°C respectively. The formation of these two necks shows the development of super-plasticity which takes place under these temperature conditions. The appearance of these two necks in the zones of lower temperature and the significance of this temperature range has been explained as follows: The carbon concentration in some parts of the grains, especially near the grain boundaries, might be higher than the average carbon content. This has been confirmed by the method of auto-radiography, by introducing radioactive carbon into iron(26). Again, Cuff and Grant showed by electron microscopic analysis, the enrichment of impurities at the grain boundaries in pure aluminium(27). The irregular distribution of carbon inside the grains and its concentration near the boundaries of grains and blocks (under cyclical changes of temperature by repeated heating and cooling) leads to the process of polymorphous $\alpha \rightleftharpoons \gamma$ transformation ~~to occur~~ occurring at temperatures of 720-850°C only in those microregions which are carbon rich. This may be explained by the fact that with increase of carbon content the temperature of transition of the α-phase into the γ-phase decreases. This temperature, which is

910°C for pure iron, decreases to 721°C if the sample contains 0.83% carbon.

As a result of the localised transformation, a sharp reduction of strength occurs which in turn results in an increase in plasticity, accounting for the formation of the necks.

Michael and Izina(25) carried out a number of other tests for the purpose of studying the influence of the upper temperature limit of cyclic heating on the disposition of the necks which were forming on either side of the middle zone. In those cases where the upper temperature limit was 850°C, only one neck appeared just in the middle zone of the specimen. As the upper temperature limit of the middle zone of the specimen was increased, the distance between the two necks also increased. This has been explained by the widening of the middle zone where the temperature remained always above A_3 value.

The experiments of Wassermann(3) in 1937 on pure iron by the application of torque have also revealed similar behaviour. An increased amount of deformation was observed during the $\alpha - \gamma$ and $\gamma - \delta$ transformations.

Also in this field the experiments of Guarniery and Kanter(28) on metastable austenite of 4-6% chromium plus 0.5% molybdenum steel were of great interest with regard to the relationship between stress, creep and austenite transformation. Experiments were made using standard 0.505" tensile test specimens on a creep testing machine. As the load was applied to the specimens in the austenitic condition, no creep was observed until transformation to pearlite started. However, during transformation, a very rapid rate of creep was observed which occurred only during a short time interval. This effect was interpreted as being due to the drastic microstructural disturbance that occurred during the austenite to pearlite transformation.

A series of investigations on the ductility of iron and iron-0.2wt% carbon alloy, deformed while undergoing allotropic transformation, was performed by de Jong and Rathenau(29). They found that the tensile elongation obtained per temperature cycle in iron while undergoing periodic transformation in the α and γ phase region, was not dependent on the rate of transformation (i.e. rate of heating or cooling) in a given temperature interval. The experiments on iron-0.2wt% carbon alloy yielded also the same result. The advantage of using this alloy over pure iron was the reproducibility of the results and the stability of the transformation front, even in the absence of a severe temperature gradient. Thus the amount of material transformed could easily be controlled by changing the temperature.

A similar result to those obtained with ~~of~~ iron and iron-0.2wt% carbon alloy has been found by Bibring and Sebilleau, on cobalt(30). These results indicate that transformation plasticity is not a time-dependent phenomenon; i.e. not a special kind of creep.

de Jong and Rathenau also found that the tensile elongation in an iron-carbon alloy, while undergoing periodic partial transformation under a constant applied tensile stress, was proportional to the amount of material transformed. Fig.1 shows the change of length of specimens of polycrystalline iron 0.2wt% carbon alloy while passing through the whole $(\alpha + \gamma)$ region both ways under a stress of 560 gr./mm^2. At 723°C the pearlite to austenite transformation was taking place whereas in the interval of 723-825°C the ferrite to austenite transformation was occurring.

The elongation $\frac{\Delta l}{l}$ accompanying the γ to α transformation, without the application of an external stress, has been found to be equal to 1/3 of $\Delta V/V$ for iron, where $\Delta V/V$ is the difference between the specific volumes of the two phases. This can only be true when the direction of the transformation front is arbitrary. If the direction of the transformation front is arbitrary, the transformation front may be approximated more or less by a closed surface(31,32). If it is a sphere $\frac{\Delta l}{l}$ would be

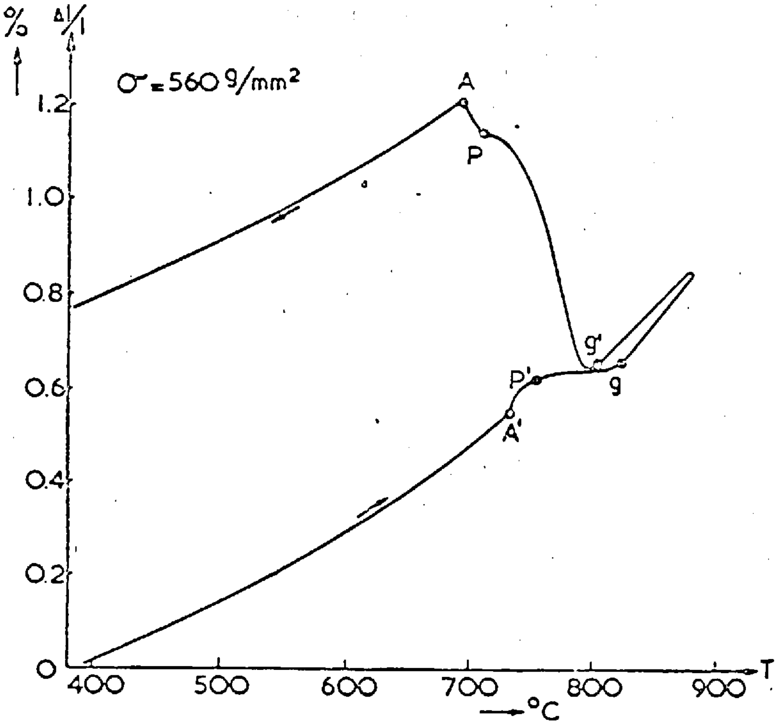

Fig. 1. Change in length of specimen of polycrystalline iron-0.2%C alloy while passing through the whole (α + γ) range both ways under a stress of 560g/mm^2. The rate of heating and cooling was constant, except at the highest temperature. The parts of the curve correspond to the following processes: A': Beginning of the transformation of pearlite to austenite and start of ferrite-austenite transformation; P': Pearlite transformation to austenite completed; g: Transformation into γ-phase completed; g': Start of transformation to ferrite P: Start of pearlite transformation; A: Transformation completed.
(After de Jong and Rathenau)[29]

equal to $1/3$ of $\Delta V/V$. If on the other hand a pure coarse grained iron transforms in a steep temperature gradient along the axis of the specimen, the transformation front may be approximated by a plane perpendicular to the gradient, thus $\Delta l/l$ would be approximately equal to $\Delta V/V$. In fact, in a coarse grained iron, the elongation accompanying the γ to α transformation under a steep temperature gradient has been found to be approximately equal to $\Delta V/V$.

The experiments which have been performed during the phase transformation by the application of tensile stresses to study the superplasticity had to be corrected for thermal expansion. So, using the iron-0.2wt% carbon alloy for their experiments de Jong and Rathenau also studied the extensive plasticity of the steel during phase transformation by the application of torque(33). The experiments on the iron-carbon alloy with different applied torsional moments during phase transformation yielded a straight line relationship between the applied torsional moment and the twist obtained by the application of that moment. The same effect was observed by the application of tensile stresses during phase transformation, which gave a straight line relationship between the applied tensile stress and the elongation caused by the stress application. de Jong also found that the iron alloys are mechanically weak when undergoing allotropic transformation under an applied compressive stress(34).

More recently Porter and Rosenthal have worked on the effect of applied tensile stress on phase transformations in a eutectoid plain carbon steel(35). They developed an apparatus capable of simultaneously recording temperature, electrical resistance and extension of a wire specimen during rapid quenching or isothermal transformation. The material used was 0.040" in diameter, having the following composition: 0.9% C, 0.45% Mn, 0.25% Si, 0.01% P, 0.01% Ni, 0.005% Cr, 0.01% Mo and 0.08% Cu. Porter and Rosenthal studied the bainite, pearlite and martensite transformations under a dead-weight tensile load. Their results on the bainite transformation showed that above a certain threshold stress which is largely dependent on the temperature of its application, the applied stress accelerated the transformation. The threshold stress was found to be approximately 8,000 lb/in^2 at 260°C, 7,000 lb/in^2 at 316°C and 6,000 lb/in^2 at 371°C. When the transformation to bainite started, the specimen also started to extend plastically at a rather rapid rate. This extension continued through the major part of the transformation but ceased before transformation was completed. The progress of rapid extension occurring during transformation to bainite is shown in Fig.2. The total elongation found by the application of tensile stresses was proportional

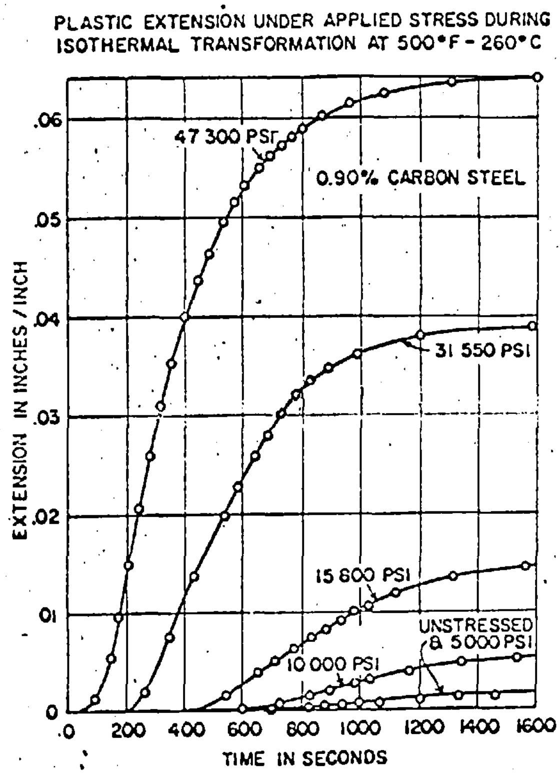

Fig.2. The progress of deformation which occurs coincident with isothermal transformation to bainite at 260°C
(After Porter and Rosenthal)[35]

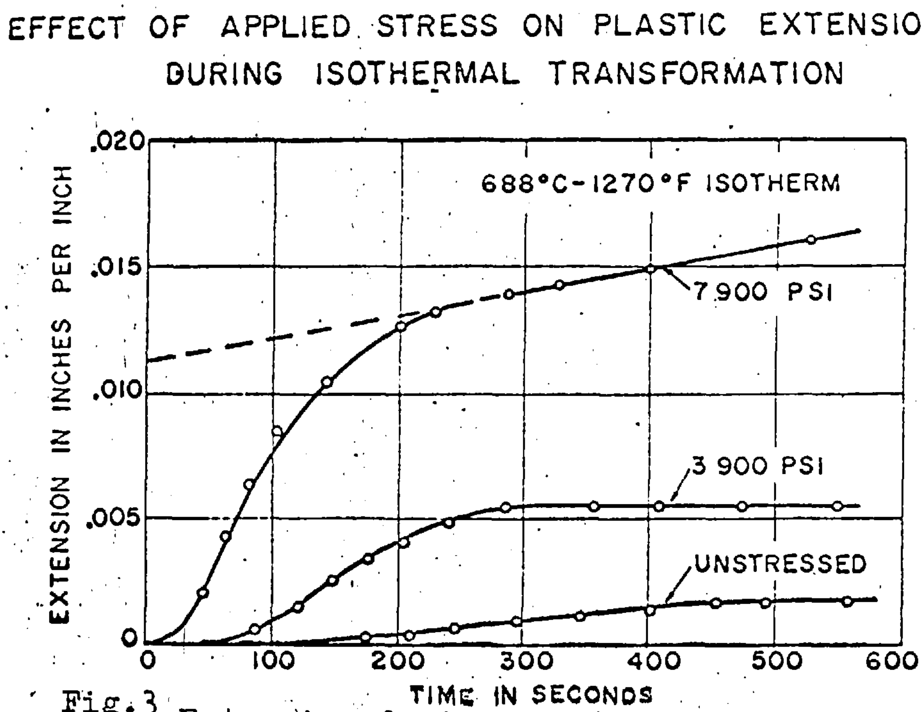

Fig.3 Extension during pearlite transformation at 688°C (After Porter and Rosenthal)[35]

to the amount of applied stress and a straight line relationship existed at lower stresses.

In the results of pearlite transformation, stresses above 2,000 lb/in^2 showed a plastic deformation coincident with the start of transformation; for stresses less than 2,000 lb/in^2 no effect of transformation was observed. Fig.3 shows the extension measurements made during isothermal transformation to pearlite at 688°C.

If the curves in Fig.3 are examined it can be seen that at a stress of 7,900 lb/in^2 extension occurs quite rapidly which is coincident with the start of transformation, and after 300 seconds a steady state creep takes place; within 300 seconds the specimen transforms about 50%. At lower stress values rapid extension and its early completion is noted again but in this case no steady state creep is observed. In short, their results indicated that there is a threshold stress associated with plastic deformation during transformation and in the case of bainite and pearlite formation this stress is also the threshold stress for increased rate of transformation. Thus if a stress exceeds a certain value, it accelerates the transformation rate and during this transformation, the yield stress of the steel is reduced greatly.

III. Transformation plasticity during martensite formation in iron alloys:

Although there is no mention of Wassermann's work in most of the review articles on transformation plasticity, it seems that the first work on low temperature ductility (transformation plasticity during martensite formation) has been done by Wassermann. His work(3) which was published in 1933 was probably the best approach to ~~the~~ understanding ~~of~~ the behaviour of transformation plasticity, though practically all later workers in this field have ~~Unfortunately his work has been largely~~ ignored ~~by the other workers even many years after~~ his detailed studies. ~~in this field.~~

Wassermann's first work in this field was in an iron-nickel alloy containing 30% nickel. Wire specimens of 1 mm. diameter were first austenitized and then brought to room temperature. If the wire was stressed with a stress of 70 kg/mm^2 in this condition, an extension of 0.75% was obtained. When the same wire after quenching in liquid air to bring it to the martensitic condition was stressed with 70 kg/mm^2 at room temperature, the extension obtained was 0.5%. In the third type of experiments the wire in the γ-state was stressed with different stresses and then immersed in liquid air. In this case, the deformation obtained was much higher than in the previous two cases. The start of sudden extension was coincident with the start of transformation to martensite and this sudden extension ceased after about 20 seconds. Table I shows the results of the experiment.

TABLE I.

Transformation plasticity during martensite formation in an Iron-30 wt% Nickel alloy.

Stress in kg/mm^2	32	38	51	70
Deformation %	2.5	3.5	5	11

Because of the similarity between shear transformation and twin formation Wassermann performed some experiments on single crystals of Zn and Cd. No evidence of weakening of the material was observed during twin formation. In his further work in 1937, on single crystals of an iron-30 wt% nickel alloy, Wassermann(3) showed that the transformation plasticity was taking place within the crystal and not at the grain boundaries; but he did not give any explanation for the transformation plasticity mechanism at that stage.

In 1949 Karskii and Sobolov observed an increased ductility and decreased yield strength during decomposition of austenite into martensite in nickel-chromium-molybdenum structural steels(36). They measured the deflection produced by a 2 kg. load on a cylindrical specimen supported at 60 mm. interval undergoing martensitic transformation. The above mentioned load corresponds to maximum stresses of 1.43, 2.45 and 4.75 kg/mm^2 for the specimens of 6, 5 and 4 mm. diameter respectively. Prior to testing, the specimen was heated to 1,000°C for 4-5 minutes, then the stress was applied at the predetermined temperature during quenching. The plastic deformation during transformation was detected by plotting deflection vs. time.

Kayushnikov also observed an increased plasticity in steels when held in a temperature range where martensitic transformation was occurring(37). Use was made of this fact to avoid distortion of steel during quench-hardening(38). For example, he applied this method to the heat treatment of crankshafts.

In their recent work on eutectoid plain carbon steel, Porter and Rosenthal observed an increased plasticity during the martensitic transformation(35). Stresses as low as 1000 lb/in^2 yielded a detectable amount of deformation during transformation while no effect of stresses on M_s temperature was observed until about 28,500 lb/in^2 was reached.

IV. Observations of transformation plasticity in non-ferrous systems:

Besides these experiments on iron and steel, there have been many investigations on other materials. For example, Goffard and Wheeler(39) reported after a study of hot-hardness tests on magnesium alloys that the alloys were softer than pure magnesium at higher temperatures and longer times of hardness testing during the two phase to single phase reaction. But there was no discussion of the implication of the results for the present topic. The most spectacular demonstration of transformation plasticity was obtained during the $(\alpha + \beta) \rightleftharpoons (\alpha' + \beta)$ reaction in an aluminium-zinc alloy with a composition of 80 wt% zinc and at the temperature of 275°C where very severe deformations of about 650% did not produce any failure. One of the most active investigators in the field of superplasticity is Prensyakov. He and his co-workers have studied alloys of Al-Zn(40-42,43,44), Al-Cu(41-45,48,49), Al-Si(42,44,45), Al-Fe(45), Al-Ni(45), Cu-Ni(44,46), Cu-Zn(47). Among the other investigators, Bockvar observed that the hardness of an aluminium-magnesium alloy was the lowest when the temperature of the alloy was at the boundary of two phase region in the equilibrium diagram(50). Similar results have been found by Osipov and Miroshkina(51) on nickel-chromium alloys where again the lowest hardness has been reported near transformation temperatures.

V. A review of the theories on transformation plasticity:

As the experimental work has proceeded on transformation plasticity, various theories have been advanced for the explanation of the observed phenomenon. For example, Konobeyevskii(52) and Osipov(53) proposed that the application of stresses at elevated temperatures can cause profound changes in the solid solution. The stressed condition of an alloy may promote the rearrangement of the atoms which gives rise to segregation, nucleation and growth of the new phases and these new phases are stable under the applied external stresses. Hence the increased rate of diffusion in the direction of the stress gradient would reduce the internal stresses so that the distorted lattice will be restored to an equilibrium condition. Thus the lattice under these conditions could undergo further deformation.

In 1948, Bockvar published a paper(54) where he discussed various mechanisms of plasticity in metallic alloys and presented a theory to explain the enhanced plasticity in two-phase alloys. In his theory, which he called the 'solution-precipitation' theory, he tried to explain the observed facts by the movement of atoms of the two (or more) phases relative to one another which is facilitated by the mutual solution and precipitation of the phases. It was suggested that this process is increased by sharp changes in solubility with temperature, high diffusion rates and fine dispersion of the phases. As a result, the precipitation and resolution of an incipient second phase leads to additional atomic mobility which results in enhanced high ductility. The diffusion that accompanies the tendency for phase change aids the deformation process.

Although Guy and Pavlick(55) supported Bockvar's theory they suggested another mechanism by which the results could be explained. They think that the local structural effects of the deformation process might result in a non-equilibrium formation of embryos of the second phase within the nearly saturated solid solution. This incipient phase transformation facilitates deformation in the vicinity of embryos. However, the mechanism of such a deformation was not specified. The high ductility is interpreted as the result of the deformation in the vicinity of embryos. Moreover the localized disturbance can change the environment of the embryos, hence causing them to redissolve. So, the cycle of precipitation and resolution accounts for the enhanced plasticity.

A different type of explanation for the increased amount of plasticity was given by Bilby(56). It was suggested that the increased ductility is associated with the stress-induced retardation of the phase transformation while the opposite behaviour, a decreased ductility, is associated with stress-induced formation of a second phase. However, Bilby does not specify a mechanism with which such a change in plasticity might occur.

The investigations of Wassermann(3) (as mentioned earlier) and Vorobev(57) are very important because they indicate that diffusional processes, as normally thought of at high temperatures, are not necessary to account for the observed ductility during low temperature phase transformations. The considerably lower strength and the considerably higher plasticity of a steel during any phase transformation has been explained as the result of instantaneous weakening of the interatomic bonds at the moment of phase transformation.

Boas suggested a somewhat different mechanism for the transformation plasticity(58). According to him the high mobility of the atoms at the interface between the parent and product phase, during transformation reduces strength and increases plasticity of the metal. The increase in weakness and plasticity of metal is the same as caused by the high mobility of atoms at grain boundaries at elevated temperatures. However, Boas' suggestion cannot tell the whole story. It is difficult to see how isolated transforming nuclei, such as must be present when the effect is first noted, could, by the mobility of their boundary atoms alone, account for gross plastic deformation of the remaining matrix.

The mechanical weakness during phase transformation has been explained by de Jong and Rathenau as follows(29): First they proved that the mechanical weakness during transformation cannot be due to some creep at the phase boundary. This is so because the elongations measured for iron do not depend on the rate of transformation, and hence are independent of time. The explanation for the mechanical weakness during phase transformations of many crystalline materials is that high internal stresses are set up during phase transformations and then the superimposition of the applied external stresses to this internally stressed system gives rise to excess plasticity.

A very similar explanation for the increased amount of deformation was given by Johnson and Greenwood(59,60). They attributed this phenomenon to the high stresses set up at the transformation front on account of the difference in density of the two phases and proposed a quantitative relationship to calculate the observed extension. It was shown that the extension is a linear function of the specific volume change, applied stress and the yield stress of the weaker phase.

A somewhat different mechanism was also put forward by de Jong and Rathenau(29) as an alternative explanation for the increased ductility. The characteristic of this mechanism is that during transformation, out of possible modes of transformation, those are favoured which allow the largest deformation in the direction of the applied stress. Increased deformation as a result of preferred orientation was also observed by Chang and Read in the gold-cadmium system(61) and by Burkart and Read in the indium-~~titanium~~ thallium alloy during diffusionless phase transformation(62). The work of Kochendorfer and Otto on iron-30% nickel single crystals, with the application of tensile stresses, showed that the first formed martensite plates had a definite orientation, but further cooling the specimen below M_s temperature introduced plates with random orientation(63). The specimens, without any applied stress, when cooled a few degrees below M_s showed no sign of preferred orientation, and all the martensite plates had a random distribution. In the same field the results obtained by Bhattacharyya and Kehl are of great interest(64). They showed the preferential occurrence of bainite formation under high values of applied tensile stresses. Applied stresses exceeding the yield stress of austenite are observed to produce a definite preference in the orientation of low temperature bainite which becomes more pronounced with increasing amount of applied stress.

An explanation for transformation plasticity has been given by Porter and Rosenthal in terms of dislocations(35). It was proposed that the applied stress causes dislocations to move out from their sources. At this moment an extensive yielding is not observed because the dislocations are held up at the grain boundaries, forming dislocation pile-ups. As soon as nucleation starts, it causes a collapse of the resistance offered by those obstacles and thus the dislocations are free to move forward, resulting in a rapid extension during transformation.

Associated with this plasticity the stresses were accelerating the rate of austenite decomposition. Porter and Rosenthal explained the increased rate of transformation by the increased rate of nucleation rather than by an increase of diffusion rate, since it is conceded that the very low stresses involved do not influence diffusion rate. However, Nabarro claims that the normal rate of nucleation of the stable phase may be materially increased when dislocation arrays pile-up at the grain boundaries under the action of applied stress; this is because the internal stress gradient near the first dislocation in a dislocation pile-up may promote the diffusion rate(65). Cottrell(66), Eshelby et al(67) and Koehler(68) have shown that an array of edge dislocations blocked by some obstacle, may cause a high concentration of tensile stress in a considerable region around the dislocation nearest to the obstacle and due to this high tensile stress an extensive dilatation of the lattice can occur below the dislocation. According to Le Chatelier's principle this dilatation of the lattice promotes the formation of product phase and so an increased rate of transformation coincident with rapid extension will be observed.

CHAPTER 3.

EXPERIMENTAL TECHNIQUE

3. EXPERIMENTAL TECHNIQUE.

I. Introduction.

i) Methods for measuring length changes:

The prime factor in the present work was the measurement of the length changes occurring during the cooling of stressed or unstressed specimens. Many methods could have been used for the measurement of the length changes occurring during transformation. These methods may be broadly divided into two categories: In the first method, measurements of changes in length can be made, only at room temperature, between the initial condition followed by a cycle of austenitizing and cooling with or without the application of a tensile stress. The experimental procedure in such a case consists of measuring the length with the aid of a travelling microscope. In the second method the measurement of changes in length of the specimen can be made continuously at any temperature with or without stress being applied during the entire heat treatment. The experimental procedure in this case is the use of a suitable dilatometric instrument, for continuous measurements. This latter method is more informative than the previous one. The present investigation utilizes both the methods in obtaining the necessary results.

ii) Methods to follow the course of decomposition of austenite to martensite:

Various methods have been used to analyse the austenite to martensite transformation characteristics of steels and other alloys.

The most direct of those which are used commonly, is the microscopic method originally used by Greninger and Troiano(69). This involves austenitizing samples of the steel at an appropriate temperature, quenching down to a temperature at which the percentage of martensite formed is to be determined and holding the specimen at that temperature for a few seconds to allow the specimen to attain uniformity of temperature; then the specimen is up-quenched to a suitable tempering temperature and held there for a few minutes to temper the martensite before final quenching to below M_f. The tempering of the initially formed martensite by up-quenching changes the martensite from a slow etching constituent to a rapid etching one. It is possible then to determine the percentage of martensite by microscopical examination of the samples. By varying the temperature to which the specimen is quenched and employing quantitative metallography, the percentage of martensite as a function of temperature below M_s can be determined. This method is still the most reliable of those available for studying steels which undergo the martensitic transformation; it is, however, time-consuming to do quantitative metallography on ~~to~~ a large number of specimens to provide an accurate result. It is consequently often more convenient to follow the course of transformation by measuring the changes in some physical property accompanying transformation and, since the breakdown of austenite is accompanied by an expansion, and changes of length are readily measured, the dilatation method is by far the most widely adopted.

As early as 1919, Chevenard(70) recognized the advantages of dilatometric methods and constructed a differential dilatometer for the study of transformation during continuous cooling. Bain(71) applied a quenching dilatometer to the investigation of the intermediate and martensitic reactions because of the difficulty of evaluating the transformation rates of these fine structures with the microscope. Oakley and Oesterle(72) have described a dilatometer, designed for a heavy tubular specimen (2" long by 0.375" inside diameter, by 0.5" outside diameter) held in compression. Griffiths, Pfeil and Allen(73) have developed a dilatometer for a 0.110" diameter compression specimen shortened to 1.14" to prevent buckling. Parke and Herzig(74) report an instrument with an optical lever for use in the intermediate range, with specimens of 0.07 or 0.034 inches thickness (held in tension).

Techniques involving the measurement of changes in magnetic permeability(75) or electrical resistance(76) have also been used by various investigators. The electrical resistance of metals and alloys is sensitive to alterations in the lattice structure. The electrical resistance of f.c.c. austenite is considerably higher than its product phase martensite. It is this property which is the basis of one of the most sensitive experimental techniques to study the transformation. Heal and Mykura(77), and more recently Gilliam and Cole(78) demonstrated the possibility of following transformations by measuring, as transformation proceeds, the changes in the intensity of a line in the X-ray diffraction pattern of either the γ or the α phase. This method, although experimentally somewhat cumbersome, has the attractive feature of being even more direct than the microscopical method, and also of providing information not otherwise obtainable.

In the present investigation, after a critical examination of many methods which have been used to observe the progress of transformation, the dilatometer and the electrical resistance methods, supplemented by the microscope, have been chosen as they seemed to offer the most valuable and comprehensive technique because of simplicity, accuracy, objectivity of observations and comparative rapidity of determinations.

The other big advantage of the dilatometer method was principally in the provision of a complete record of length changes during the entire heat treatment. This equipment was also used for recording expansion and contraction curves in the determination of transformation temperatures during heating, e.g. the $A_{c1} - A_{c3}$ range for the steel and A_s and A_f temperatures for the iron-20 wt% nickel alloy.

All the methods which depend on the change of a physical property during transformation when used alone do not provide the detailed information obtainable from microscopic examination. Consequently, it is customary to supplement the results obtained by these less direct methods by some microscopical examination.

II. Choice of the specimen material.

Wire specimens of 0.02" diameter of an En 30A steel (BISRA code No. HD) and 0.028" diameter of iron-20 wt% nickel alloy were used in the investigation; the steel composition was as follows:

C 0.31% Ni 4.13% Cr 1.33% Mo 0.12% Si 0.13%
Mn 0.49% S 0.012% P. 0.011%

The iron-20 wt% nickel alloy made by powder metallurgy technique was supplied by the International Nickel Company (Mond) Ltd. and had impurity limits as follows:

C, 0.001% O, 0.005%
S, 0.001%

The steel was selected on the basis of its high hardenability, which enabled martensite to form during free cooling of wire specimens in vacuum; the TTT diagram shows a fairly lengthy incubation period for bainite formation. Since bainite formation may be accelerated under an applied stress(64), it was desirable to have a fairly lengthy incubation period in the experiments where austenite is stressed above M_s.

The specimens were used in the form of 0.02 and 0.028 inches diameter wires so that the heating and cooling cycles could be carried out effectively in vacuum and this enabled rapid and continuous measurements to be made of electrical resistivity, dilatation and temperature. Also in choosing the wire diameter, consideration had to be given to the stress values to be applied so that the loads would not be too large to accommodate in the vacuum chamber.

III. Details of the Apparatus used.

i) Introduction:

The apparatus used was an adaptation of that described by Wells(79). It enabled measurements to be made either of the temperature and electrical resistance of a specimen only during continuous cooling, or of the temperature and length change of a specimen during the entire heat treatment, and it incorporated a device for applying a tensile stress to the specimen at any required temperature. One of the main advantages of this apparatus is that no movement of the specimen from a heating furnace to a quenchant is necessary.

The wire specimen, about 6" long, was vertically mounted, held in small grips in the vacuum chamber, and was heated in a vacuum of about 10^{-4} torr to the austenitizing temperature by the passage of an alternating current. After austenitizing, the alternating current was replaced by a small direct current (required for the measurements of electrical resistance) and the specimen was allowed to cool in vacuum. Chromel-alumel wires (0.002" diameter) were welded to the centre of the specimen length to form a thermocouple. Also, pure iron wires of the same diameter were welded to the specimen at points 2" apart and equidistant about the thermocouple, to form the leads for measuring electrical resistance. The thermocouple and electrical resistance leads were connected to the Y and X axis of an X-Y recorder respectively; thus a continuous record of temperature and electrical resistance was obtained during cooling. By the use of a transducer, a complete record of temperature against length changes during the entire heat treatment was also obtained on a different X-Y recorder. From temperature vs. resistance, or temperature vs. length change curves, the M_s temperature and the % martensite existing at various temperatures below M_s may be determined.

The stressing of the specimen during cooling could be achieved by means of a dead weight loading system, the load being contained within the vacuum chamber and being applied from the outside of the vacuum chamber. By recording the length change under the load against temperature, the progress of deformation was followed. The incorporation of a dynamometer within the vacuum chamber in place of the dead weight loading assembly made it possible to obtain plots of the stress-strain relations of austenite at a certain temperature directly on an X-Y recorder, from which the yield stress of austenite was found.

ii) The vacuum chamber and the specimen assembly:

The vacuum chamber containing the specimen, transducer and loading assembly was a brass cylinder, 6" in diameter and 2 ft. long, mounted vertically. At the lower and upper ends of the vacuum chamber two windows were incorporated to observe the application and removal of a load and also to measure the extension obtained by the application of stress. An Edwards 2" (Model 102) oil diffusion pump and an NGN rotary backing pump (type PSR 2) were used to evacuate the system. The system was equipped with suitable valves so that it could be isolated from the pumps if needed. A needle valve was fitted to the top-plate of the vacuum chamber to let the air into the system when the chamber was isolated from the pumps.

This facilitated the specimen changing by eliminating the necessity to stop the pumps when the vacuum chamber was exposed to the atmosphere. The pressure of the vacuum chamber was measured by means of an Edwards thermal (Pirani type) vacuum gauge (model 7-23), fitted to the top plate of the vacuum chamber.

The specimen which was about 6" long was suspended vertically downwards from the centre of the top plate of the vacuum chamber, and was attached to the top plate with the help of a pin vice and a universal mounting. The role of the universal mounting was to ensure that uniaxial stressing was obtained. The other end of the specimen was gripped with another pin vice (Eclipse pin vice No.121) coupled again with a universal mounting. A hook which was attached to the one end of the universal mounting served to join the universal mounting to the loading device. A schematic diagram of the assembly details is given in Fig.4. The pin vices used to grip the specimens facilitated the mounting and removal of specimens; they also secured a very good electrical and mechanical contact. No slipping was experienced during stressing the specimens. At the rear of the specimen a rigid steel strip support, fixed to the top plate, carried the silica tubes containing the thermocouple wires and potential leads.

Measurements of the temperature of the specimen

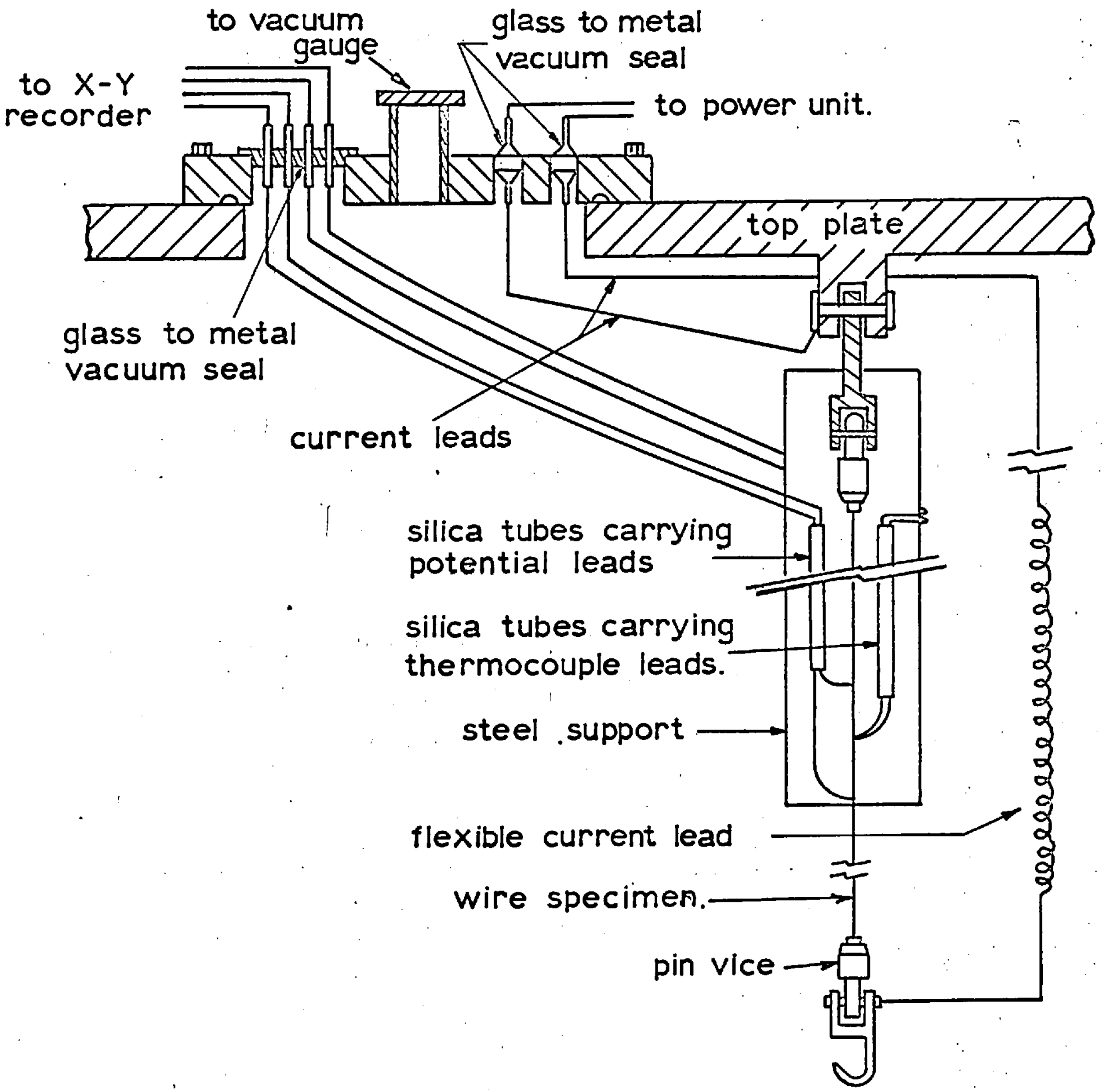

Fig. 4. Schematic diagram of the specimen assembly.

were made using ~~a~~ chromel-alumel thermocouple wire. 30 s.w.g. (0.0124" diameter) wires were introduced to the vacuum chamber and within the vacuum chamber they were enclosed in silica tubes. The final connection of these wires to the specimen was made by using 0.002" diameter wires of the same composition, respectively, and by spot welding them to the middle of the specimen.* The use of these very fine wires made it possible to minimise the heat conduction from the specimen, ensuring uniformity of temperature along the central portion of the specimen. In spot welding these very fine thermocouple wires, care had to be given to ensure that the welded joints met at the same point on the wire specimen so that the A.C. or D.C. current passing through the specimen did not alter the readings of the thermocouple.

Pure iron wires were used to form the potential leads for measuring electrical resistance. These wires were introduced to the vacuum chamber in the same way as the thermocouple leads and in the chamber passed through silica tubes. The final connections were made by using pure iron wires of 0.002" in diameter, by spot welding them to the specimen at points two inches apart, and equidistant about the thermocouple. The use of the thinner wires as final connections allowed free vertical movement of the specimen resulting from the application

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*The spot welding of the wires will be discussed in full detail later in the text.

of stress as well as from the thermal length changes.

The potential leads and the thermocouple wires were introduced to the vacuum chamber through the tubes of an Ediswan (Type 3294) multiple glass-to-metal seal. The seal was soldered to the top plate of the vacuum chamber. After passing the wires through the small metal tubes of the seal, the space around the wire and small tube was filled with soft solder to make it vacuum-tight. The heating current leads were also introduced through the top plate and for this purpose two Ediswan (Type 3366) single glass-to-metal seals, soldered to top plate, were used. The current leads were attached to the pin vices by soldering. The current lead which joined the bottom pin vice to the glass-to-metal seal was flexible to ensure free vertical movement of the specimen.

The transducer leads were introduced to the vacuum chamber from the side of the cylinder by using 3 Ediswan (Type 3279) single glass-metal seals. In the chamber the leads were passed through silica tubes and soldered to the terminals of the transducer. A similar arrangement was made for the leads which joined the dynamometer, except in this case the leads were introduced from the base plate of the vacuum chamber through an Ediswan (Type 3294) glass-metal seal. Again the seal was soldered to the base plate and the space between the leads and the fine tubes of the seal was filled with soft solder to obtain a vacuum tight joint.

iii) Loading device:

In designing an apparatus for the study of transformation plasticity, it was important to apply or remove tensile loads at a selected temperature above, below or at M_s, very rapidly, but without any impact effect, and also to control closely the temperature of stress application or removal.

Two methods could be used; loads could be contained either inside or outside the vacuum chamber. If the load had been located outside the chamber stressing would have been possibly only by the use of a Wilson seal or bellows. At low stress values the frictional force inherent in the seal compared with the total stress applied would be quite large and this would have required a very accurate calibration. For this reason it was decided to contain the load within the vacuum chamber.

The design of the loading device was such that the load contained within the vacuum chamber could be applied from the outside. A schematic diagram of the loading device is shown in Fig.5.

The system consisted essentially of a platform supported by a 1/4" diameter silver steel rod passing through an Edwards seal, in the base plate of the cylinder. Stress could be applied or removed simply by lowering or raising the platform respectively which supports the load. This was achieved by connecting the silver steel rod to

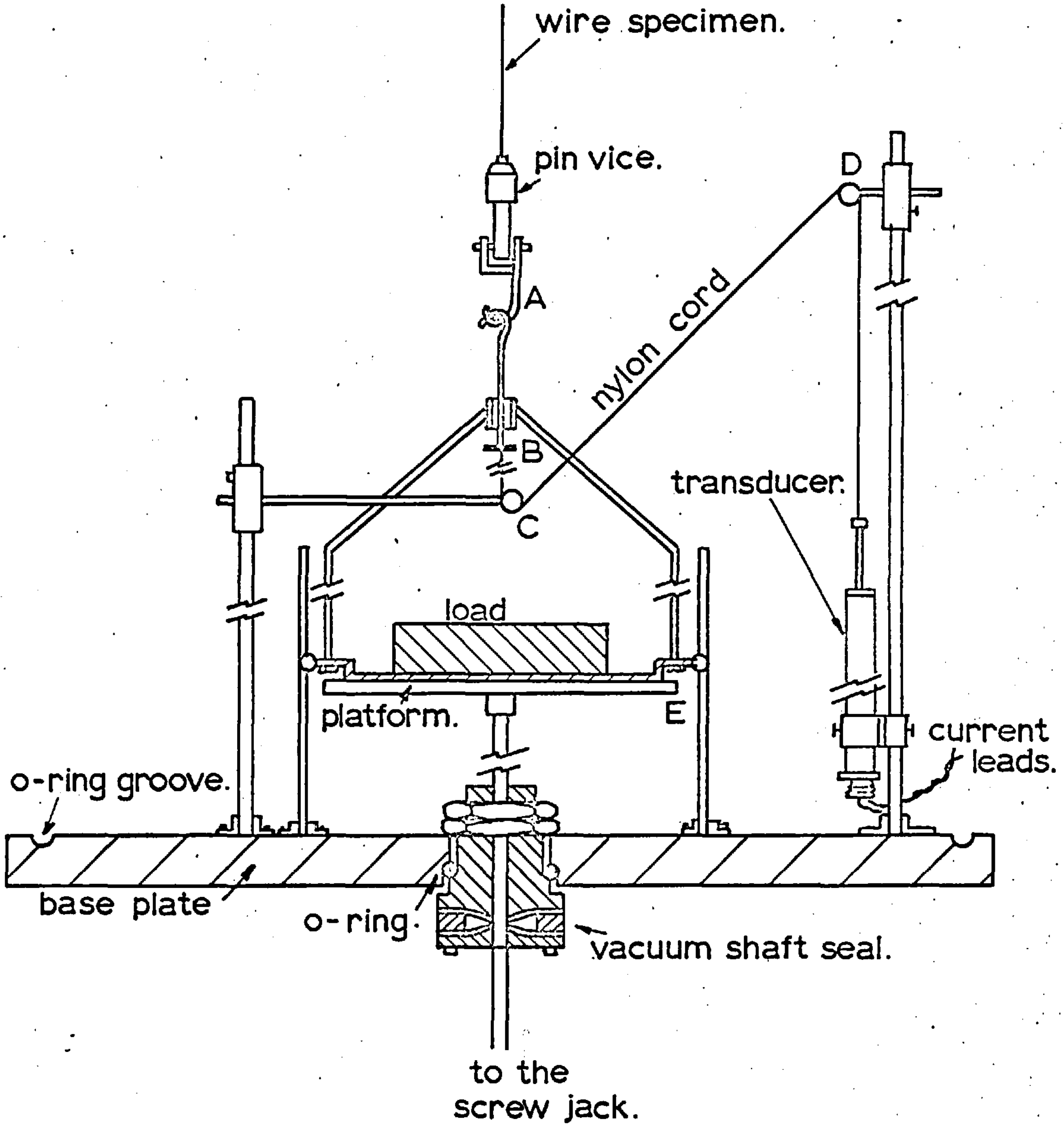

Fig.5. Schematic diagram of transducer and loading system.

a cross-head driven by a screw mechanism. Figs. 6a and 6b show the loading device. A (15 lb/in) reversible induction A.C. motor supplied the power to drive the screw mechanism. A gear box which was incorporated between the motor and screw mechanism enabled the rate of loading to be changed.

The load, resting on the pan, hung from the specimen by the hook as shown in Fig.5. During austenitizing, the load was supported on the platform, and then during cooling it could be applied at the required temperature by a switch which started the motor, and thus lowered the platform. The temperature of stress application or removal could be controlled to within 2°C. To eliminate any impact effect during rapid loading the vertical movement of the pan was guided by two rods, and also to eliminate any friction between the pan and the guiding rods, four ball races were incorporated. The application and removal of stress was observed either by noting the position of the platform with respect to the pan or by noting the position of the hooks between the pan and the lower pin vice. (This will be discussed later in the text.)

To study the effect of stresses during transformation, dead-weight tensile stresses were used. However, for studying the stress-strain relations of austenite at various temperatures, it was essential to strain the

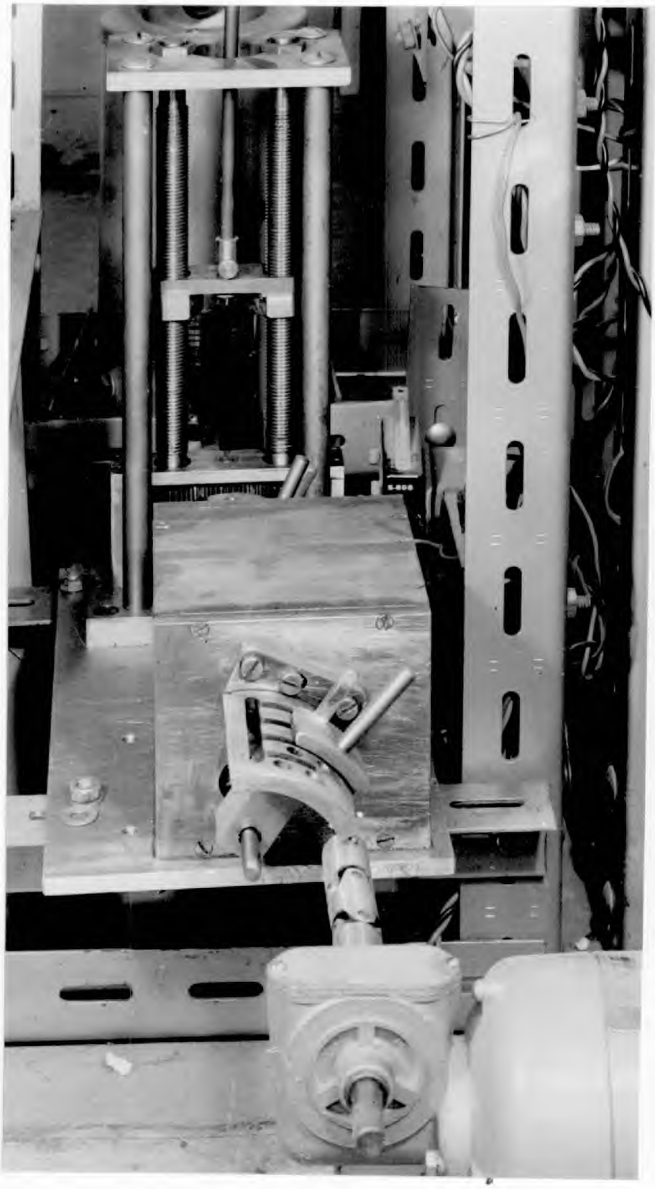

Fig. 6 (a): Loading device

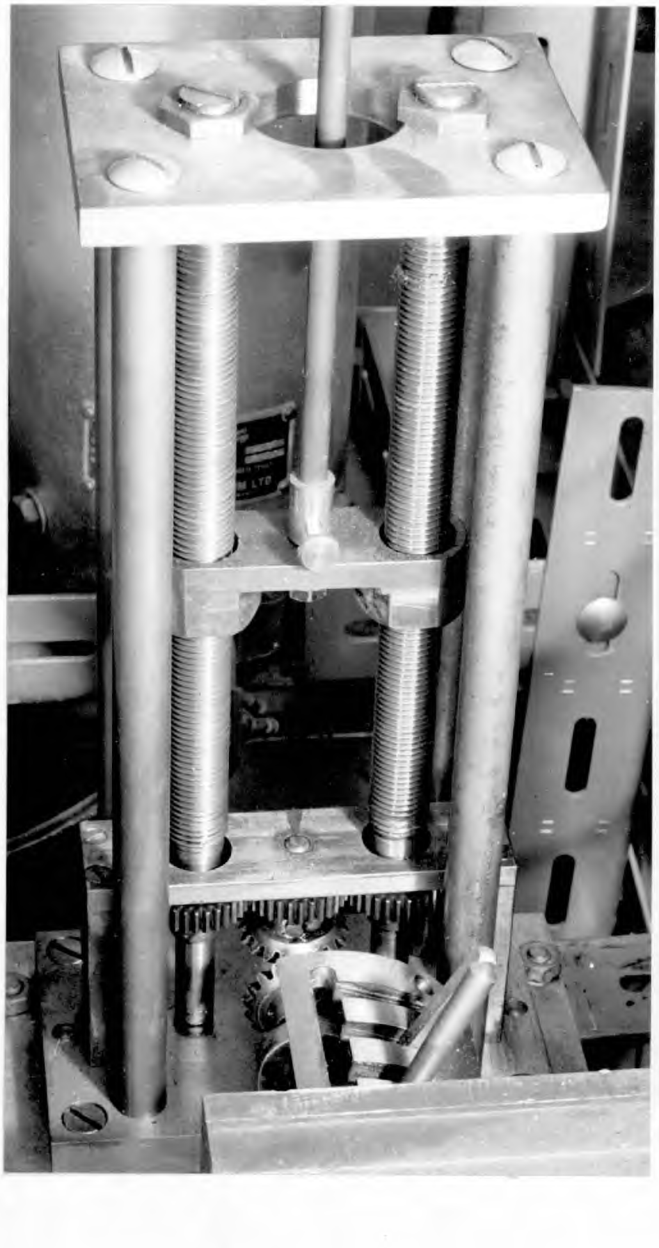

Fig. 6 (b): Loading device

specimens continuously. This involved the incorporation of a dynamometer (J. Langham Thompson Ltd.) within the vacuum chamber having a load range of 0-50 lbs, in place of the dead weight loading assembly. The dynamometer was fixed rigidly to the platform so that the movement of the rod attached to the cross-head of the screw mechanism then gave continuous straining. The incorporation of the dynamometer is shown schematically in Fig.7. The output of the dynamometer was fed to the X-Y recorder.

iv) Dilatometric equipment:

a) Transducer:

It is useful to have a complete record of all the changes in length occurring during each test, for the interpretation of the mechanism of transformation plasticity. Changes in length of the specimen were measured by adopting a transducer, commonly used in dilatometric measurements. Thus continuous records of variation of length during heating or cooling, with or without stress, were obtained. The transducer used was a Boulton Paul Aircraft Ltd. type F 14, and was designed for the measurement of relative displacement over the range of ∓ 1.0 inches.

A schematic diagram of the working principle of the transducer method is shown in Fig.5. The pin vice which is used to grip the lower end of the specimen is

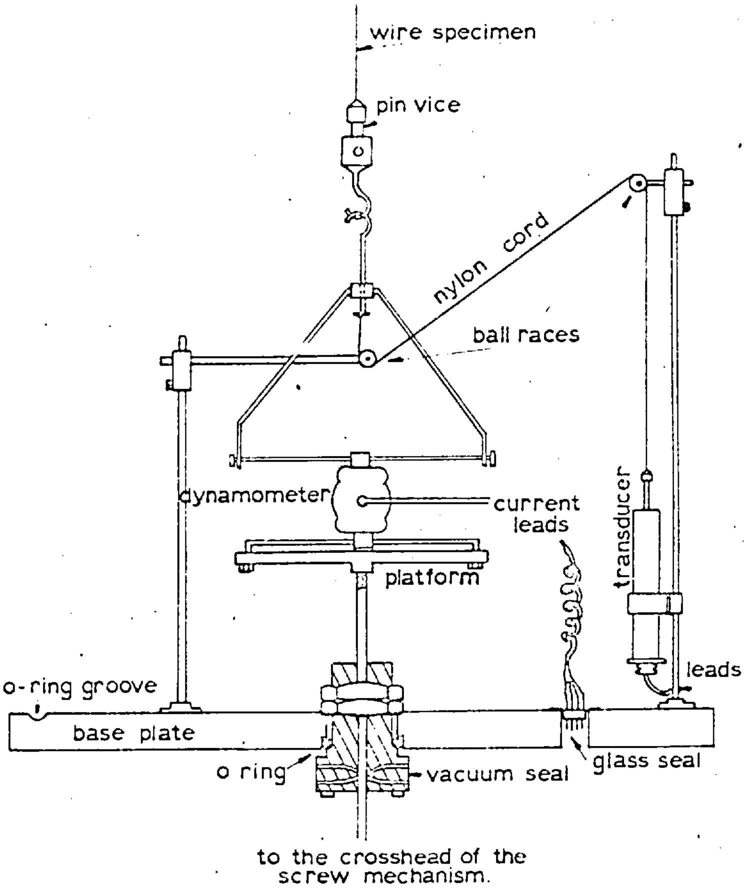

Fig. 7. Schematic diagram of dynamometer system.

connected to the hook 'A'. The hook 'B' which is a part of the pan assembly is then hooked to hook 'A'. A very fine nylon cord connects the hook 'B' via a pulley system to the armature of the transducer. 'C' and 'D' are two ball races used to minimize the friction between the nylon cord and pulley. The weight of the moving parts (armature) of the transducer was approximately 14 grammes and this weight was enough to keep the cord under a constant tension along its full length as the friction on pulleys 'C' and 'D' (ball races) were almost zero. In the 'unloaded' position, the pan with the load on it is resting on the platform. At this position, hook 'B' is not supporting the load in the pan and can move freely with respect to it. Thus during heating and cooling any length change resulting either from thermal or volume change due to the transformation is transmitted from the specimen to the transducer.

If the loading of the specimen at any selected temperature is desired, by switching the motor on, the platform 'E' is lowered and this brings the hook 'B' in contact with the pan assembly. Now the hook 'B' can no longer move freely with respect to the pan, and the movement of the whole assembly is transmitted to the transducer. (As previously mentioned, the application and removal of the stress was observed by the position of the hooks 'A' and 'B',and a mirror which was placed inside the vacuum chamber facilitated the viewing.)

The transducer consists essentially of a movable armature and two series-connected windings. In operation, the two coils are arranged to form two active arms of a bridge network which is fed from an A.C. source having a frequency between 1 and 3 kc/s. As the specimen changes length, the pulley system transmits the movement to the armature of the transducer; this varies the inductance of the two coils differentially, thus unbalancing the bridge and producing an output signal proportional in phase and magnitude to the displacement of the armature from its central position. This signal is fed via an amplifier to a X-Y recorder. By this method on the finest scale of the instrument, an extension of as little as 2×10^{-5} inches could be registered. However, this fine scale was not used in the present investigation, because very small changes in length were giving full scale deflection on the recorder. So with the scale chosen it was possible to measure length changes of as little as 5×10^{-4} inches.

b) Cathetometer:

A second method of measuring the extension of a specimen resulting from the application of a load was the use of a cathetometer (The Precision Tool and Instrument Co. Ltd.), which was placed outside the vacuum chamber. At the upper end of the vacuum chamber a window was incorporated for sighting the specimen. This method enabled measurements to be made only on the difference in length of a specimen between the initial condition and that following a cycle of austenitizing and cooling under stress.

v) The circuit diagram:

The electrical circuit diagram of the apparatus can be conveniently divided into six parts as follows:

a) Alternating current (heating) circuit.
b) Direct current circuit.
c) Extension measuring (transducer) circuit.
d) Stress measuring (dynamometer) circuit.
e) Stressing circuit.
f) Recording circuits.

a) Alternating current heating circuit:

The heating circuit diagram is shown schematically in Fig.8. The heating of the specimen was achieved by passing a high amperage current at a low voltage through it. The current transformer used in the circuit had a current capacity of 30 amperes at the secondary coil. The output of the secondary coil was controlled by a 500 watt variac, connected across the mains, which served to vary the input voltage to the primary coil of the current transformer. An (1.5 Amp) ammeter which was connected in a series with the primary side of the transformer, served

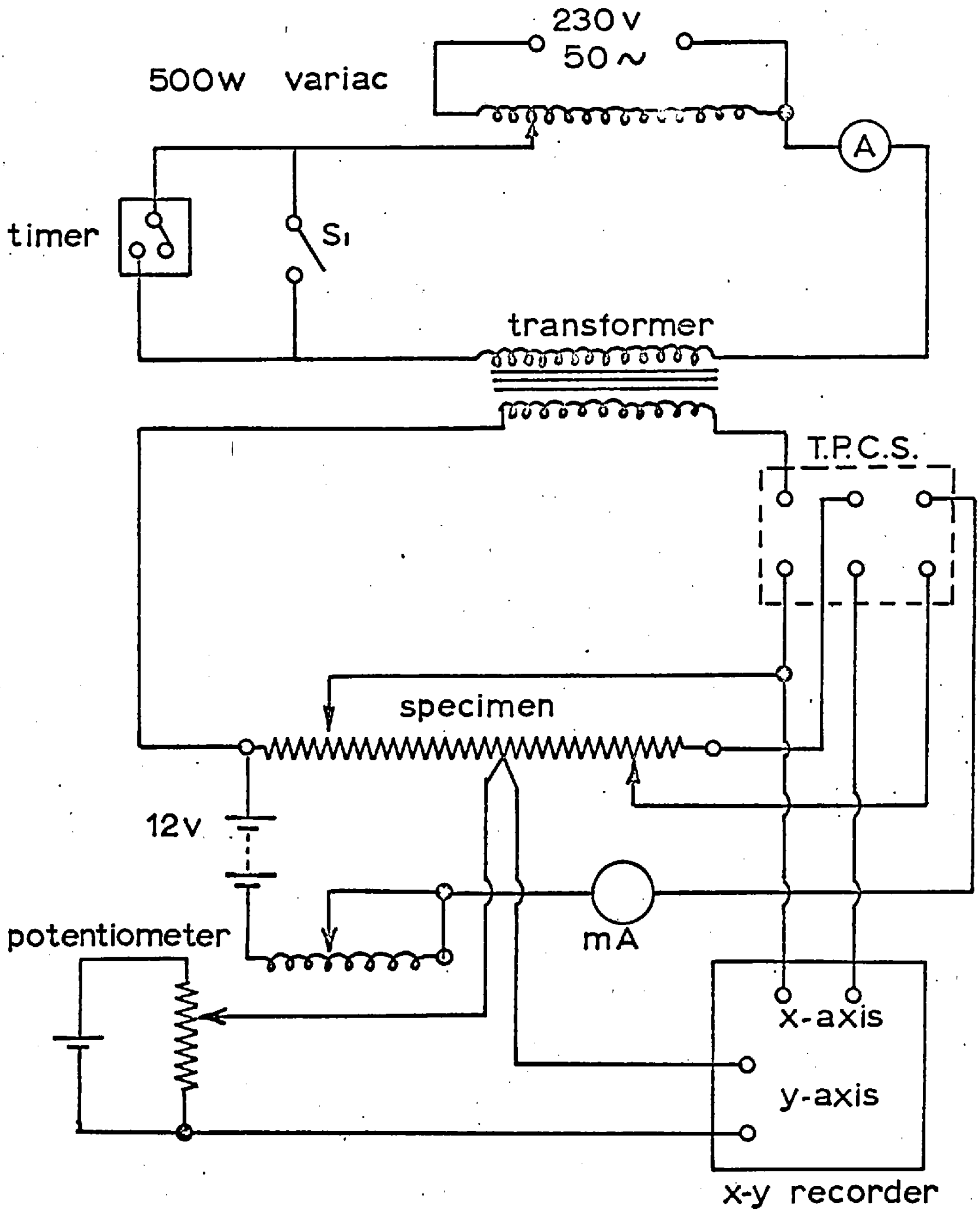

Fig. 8. The circuit diagram.

to note the current in the variac. The wire specimen was connected across the secondary coil with a 2 pole change over switch, denoted by TPCS, as shown in Fig.8. The temperature to which the specimen was heated could be controlled to within ∓ 2°C by means of a variac.

To control the heating rate of the specimen effectively a Venner electronic process timer was incorporated in the A.C. circuit (Fig.8). The timer, which was connected in series with the variac, had a range of 0 to 500 seconds, and it could be left out of the circuit at will by operating the switch S_1.

b) Direct current circuit:

Two 6-volt accumulators were used to supply the necessary current to the resistance measuring circuit. The circuit also contained a variable resistance and a milliammeter connected in series with the specimen. The TPCS switch then completed the circuit (Fig.8). The maximum value of the variable resistance of the circuit was 5 kΩ. A very high resistance was chosen so that any changes of resistance of the specimen do not significantly affect the total resistance of the circuit, since the resistance of the specimen compared with the external resistance would be small; then the direct current passing through the specimen remains substantially constant.

The constant direct current passing through the specimen during cooling or during transformation produces a voltage drop across the potential leads, and this then gives a measure of change in resistance of the specimen.

c) Extension measuring (transducer) circuit:

For the measurement of any length changes occurring during heating or cooling a transducer (Boulton Paul Instrument Company, type F 14) has been used. Fig.9 shows the circuit diagram of the transducer. It consists of a movable armature and two series-connected windings. This system forms an electrical bridge circuit which is energised by a transducer-meter (Boulton Paul Instrument Company, type C 51). Fig.10 shows the balancing circuit of the transducer meter schematically. The output from the transducer-meter, which is proportional in phase and magnitude to the displacement of armature, is then fed via a filter to the X-Y recorder. Since the recorder incorporated a 50 cycles per second chopper input amplifier, a fully smoothed D.C. signal was necessary for its satisfactory operation. To meet this need, a self contained plug-in filter unit (type C 42) has been used between the transducer-meter and the X-Y recorder.

d) Stress measuring (dynamometer) circuit:

To measure the stress applied to the specimen during continuous stressing, a Langham Thompson Ltd. (type 348/50/350) dynamometer was used. Three 6-volt

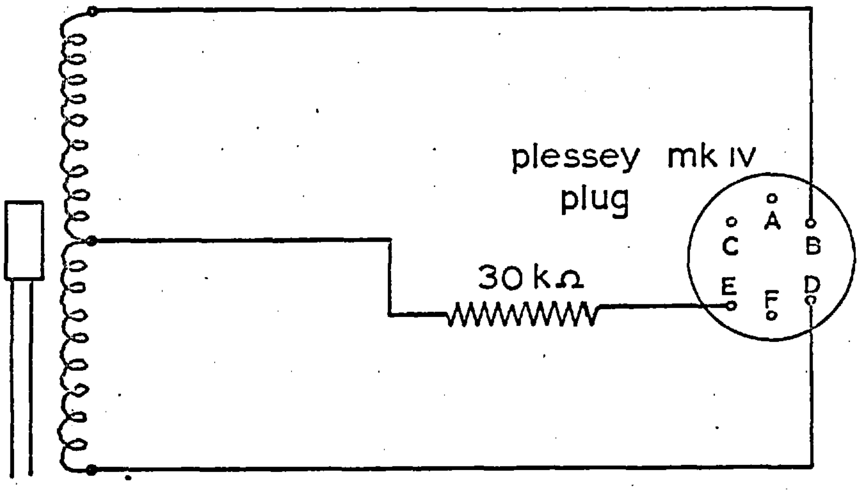

Fig. 9. Transducer circuit diagram.

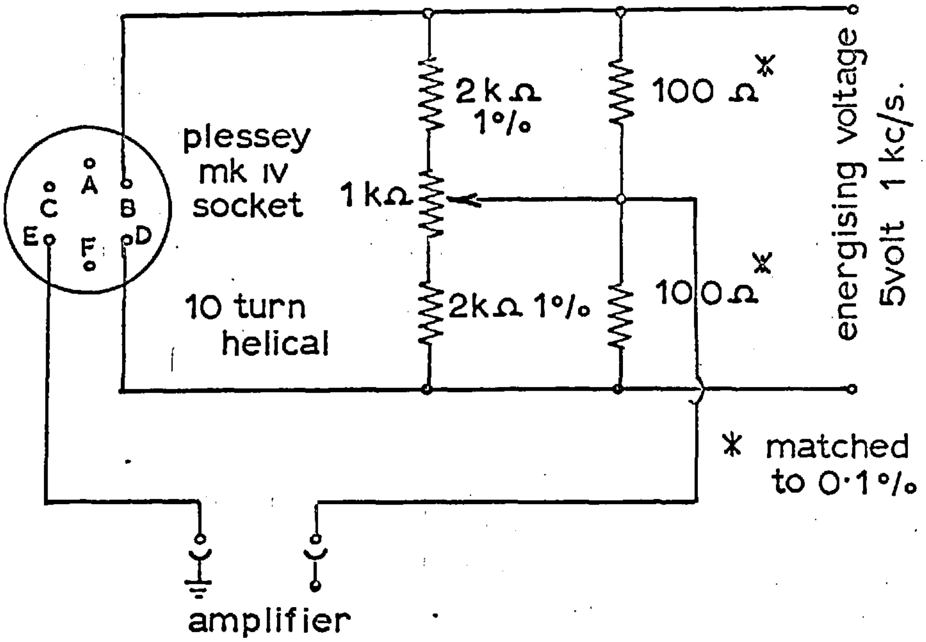

Fig. 10. Transducer balancing circuit.

accumulators, by the operation of a switch, provided the direct current necessary for energising the dynamometer. Before each experiment, the voltage of the batteries was checked by measuring the current flowing across a known 2,000 ohm resistance. For this purpose, a double pole change over switch was incorporated in series with the resistance circuit. The circuit diagram of this system is shown in Fig.11.

The working principle of the dynamometer is very simple. The four resistances shown in Fig.11 form an electrical bridge circuit. An applied stress changes the resistances of the network, thereby giving an output voltage which is proportional in magnitude to the applied stress. Then this output is fed via an amplifier to the X-Y recorder. With this dynamometer, loads up to 50 lbs could be registered. The accuracy of the gauge was ∓ 0.1%.

e) Stressing circuit:

The necessary power for the application or removal of the load to the specimen was obtained from an A.C. induction motor. The motor (M.R. Supplies Ltd.) was reversible, had a fairly constant speed of 62 rpm and a torque capacity of 15 lb/in. The circuit diagram of this system is shown in Fig.12. The neutral lead of the mains was directly connected to the motor. However, the live lead of the mains was connected to the windings of the

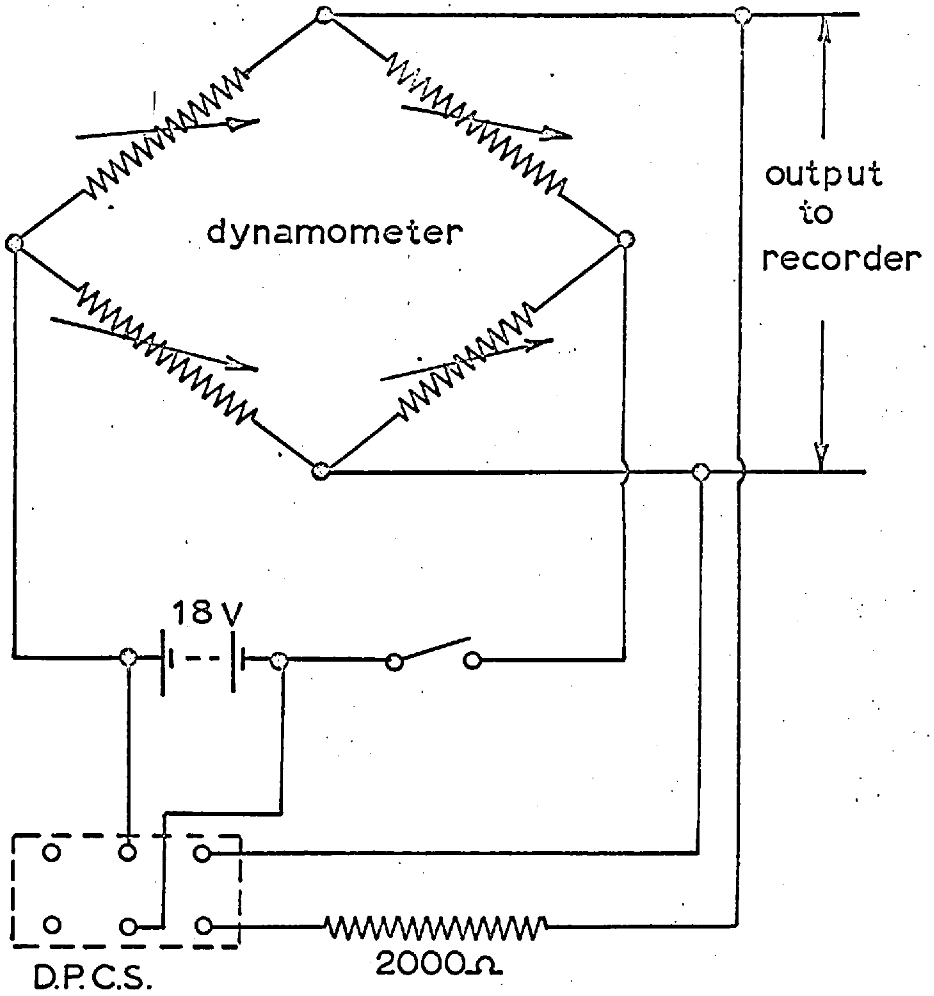

Fig. 11. Stress measuring (dynamometer) circuit.

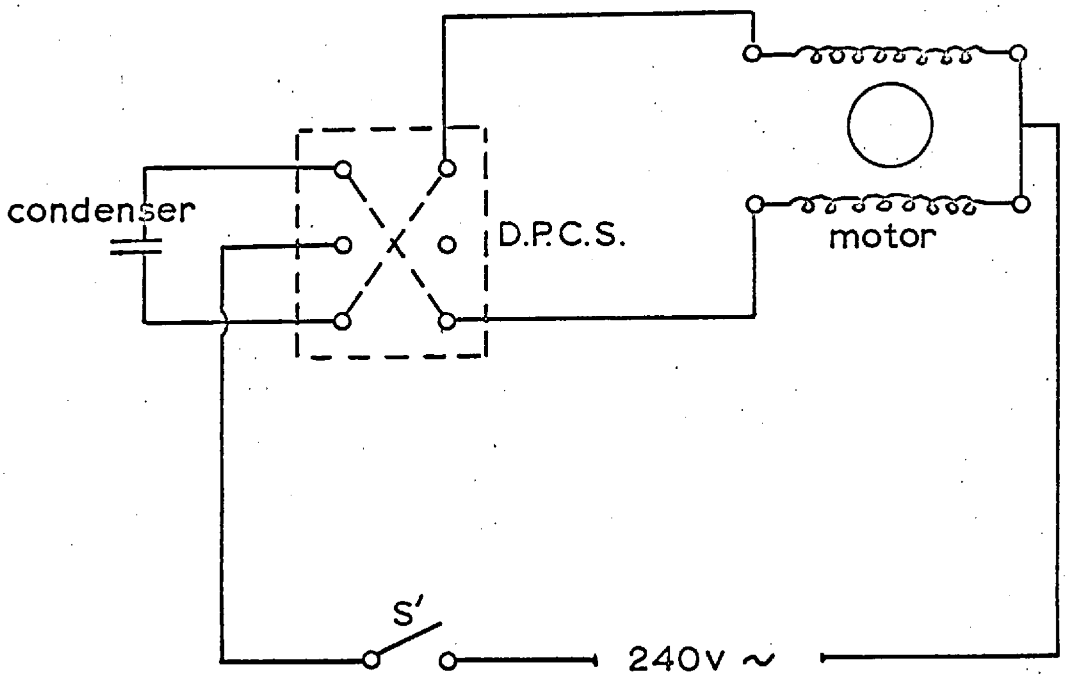

Fig. 12. Stressing circuit.

motor by a double pole change over switch (DPCS). The circuit was also equipped with a condenser to facilitate reversing the motor. One on and off switch was also incorporated in the circuit (as shown in Fig.12) to stop or start the motor.

f) Recording circuits:

To measure continuously the temperature of the specimen a chromel-alumel thermocouple was used, the output of which was fed through a cold junction (0°C) to the Y-axis of a Philips electronic X-Y recorder. In the thermocouple circuit, a Tinsley potentiometer was used to supply a back e.m.f. to oppose the thermocouple potential. This enabled the use of a more sensitive scale on the recorder. Most of the temperature readings were taken using a full scale deflection on the chart of 10 MV. which corresponds to 0-250°C. Using this scale, temperatures could be measured to within ∓ 1°C. No switch was incorporated in this circuit so that the temperature could be measured continuously during heating and cooling cycles.

During cooling, the millivoltage drop across the potential leads was fed through the TPCS switch (Fig.8) to the X axis of the same recorder. Thus a record of the resistance change of the specimen was obtained. Since the alternating current interfered with the direct current while heating, the electrical resistance measurements were done during cooling only. A TPCS switch was incorporated between the alternating current circuit and electrical resistance measuring circuit to ensure the isolation of the X-axis of the X-Y recorder during the heating cycle. The switch in the 'on' position connected the specimen in series with the alternating current circuit, to heat the specimen to the required austenitizing temperature; in the meanwhile, this position of the switch short circuited the resistance measuring leads which protected the recorder from the high alternating current. After an appropriate austenitizing treatment the TPCS switch was switched off and now the terminals of the X axis were in direct connection with potential leads, and at the same time this position of the switch connected the specimen in series with the direct current circuit, thus enabling the electrical resistance measurements to be recorded. No potentiometer was used in this circuit since the potential drop across the leads did not exceed 3 millivolts. Most of the electrical resistance readings were taken using a full scale deflexion on the chart of 5 mv. In this way a direct plot of electrical resistance vs. temperature could be obtained during continuous cooling of the specimen.

The record of any length change that is taking place during heating or cooling with or without the application of a load was obtained by feeding the output signal of the transducer to the Y-axis of a Honeywell X-Y recorder, through a multimeter and a filter. Most of the length change readings were taken using a full-scale deflexion on the chart of 10 mv. It was also possible to change the scale of deflexion recorded, by changing the range selector switch on the multimeter. This enabled the very fine scale to be used to find the thermal critical points quite sharply. During stressing, different scales were employed depending on the magnitude of stress. The output from the thermocouple was fed to the X-axis of the recorder. In this way, a temperature dilatation curve was obtained.

To obtain stress-strain curves, during continuous stressing, the outputs of the dynamometer and transducer were fed directly to the X and Y axis of the recorder respectively. No switch was incorporated in these circuits to obtain continuous reading of stress and elongation during cooling or holding the temperature of the specimen constant at any required temperature by passing an alternating current through it.

vi) The X-Y recorders:

Two different X-Y recorders were used, manufactured by Philips and Honeywell respectively. Most of the temperature vs. electrical resistance measurements were obtained using the Philips X-Y recorder. This recorder had an accuracy of 1% on the X-axis and 0.5%

on the Y-axis. The dilatation vs. temperature or stress vs. strain plots were obtained by using Honeywell X-Y recorder. The calibrated accuracy of this recorder was within ∓ 0.25% of the full scale deflexion. For some of the experiments the Honeywell X-Y recorder was used for temperature vs. electrical resistance and the Philips one for dilatation-temperature measurements, to find whether there was an effect of the recorder used on the observed results. The same results were obtained when the recorders were interchanged.

vii) The calibration of the apparatus and the measuring circuits:

No calibration of the transducer was made since it was calibrated just before purchase. When the output from the transducer was fed to the multimeter, the amount of deflexion could be measured on the 4" scale-meter. This deflexion was checked against the readings obtained on the X-Y recorder.

The readings of the transducer were also checked by measuring the thermal expansion of pure iron between 20° and 300°C. The experiments gave a value of about 14.3×10^{-6} in/in/°C for the mean coefficient of thermal expansion and this value is in very good agreement with the accepted value of 14×10^{-6} in/in/°C for the mean coefficient of thermal expansion between 0° and 300°C.

The calibration of the dynamometer was done by hanging different known loads to the lower end of the dynamometer and feeding the output from it to the recorder. In this way, a relationship between load and output millivoltage was obtained.

To calibrate the X-Y recorders, the X and Y channels were checked against a standard potentiometer from time to time to ensure that correct potentials were being recorded during the experiments.

The calibration of thermocouple was carried out by measuring the temperature of the Curie point in pure iron. At the Curie temperature a definite, though small, change in electrical resistance takes place. This is accompanied by a thermal change as well. The Curie temperature is independent of the heating or cooling rate. A number of experiments gave a value of about 765°C∓5°C. This is in good agreement with the accepted value of 768°C for the Curie Temperature(80). Since very good reproducibility was obtained in the actual measurements carried out and also since the results were essentially comparative, the accuracy reported above was considered very satisfactory.

IV. Specimen Preparation:

Wires 7" long were cut from the stock and then polished with 600 silicon carbide paper to remove the oxides from the surface of wire. Then the wire specimen was very carefully straightened by the use of a pair of tweezers. During this process, great attention was given not to stretch the specimen, since this would introduce necking. Finally, the straightened wire was cleaned with trichloroethylene to remove the grease and dirt.

The top plate of the vacuum chamber was removed from the chamber and placed on a Dexion stand, which was specially designed to facilitate the assembly of the specimen. The specimen was mounted between the pin vices by inserting ½" of specimen on each side to the grips. By doing so a gauge length of 6" of specimen was left between the grips.

The thermocouple and potential leads were attached to the specimen by spot welding. The circuit diagram of the condenser discharge apparatus, constructed by Wells(79) used for the welding process is shown in Fig.13. It consisted of a step down transformer, rectifier and DPDT switch, two condensers and an ammeter. The best welding condition was obtained during discharge when the condensers of 120μF capacity were charged to 30 v. D.C. To make the weld, the wire specimen was connected to the negative lead of the apparatus by means of a crocodile clip. The other lead was connected to a pair of tweezers which was used to grip the 0.002" diameter wires of

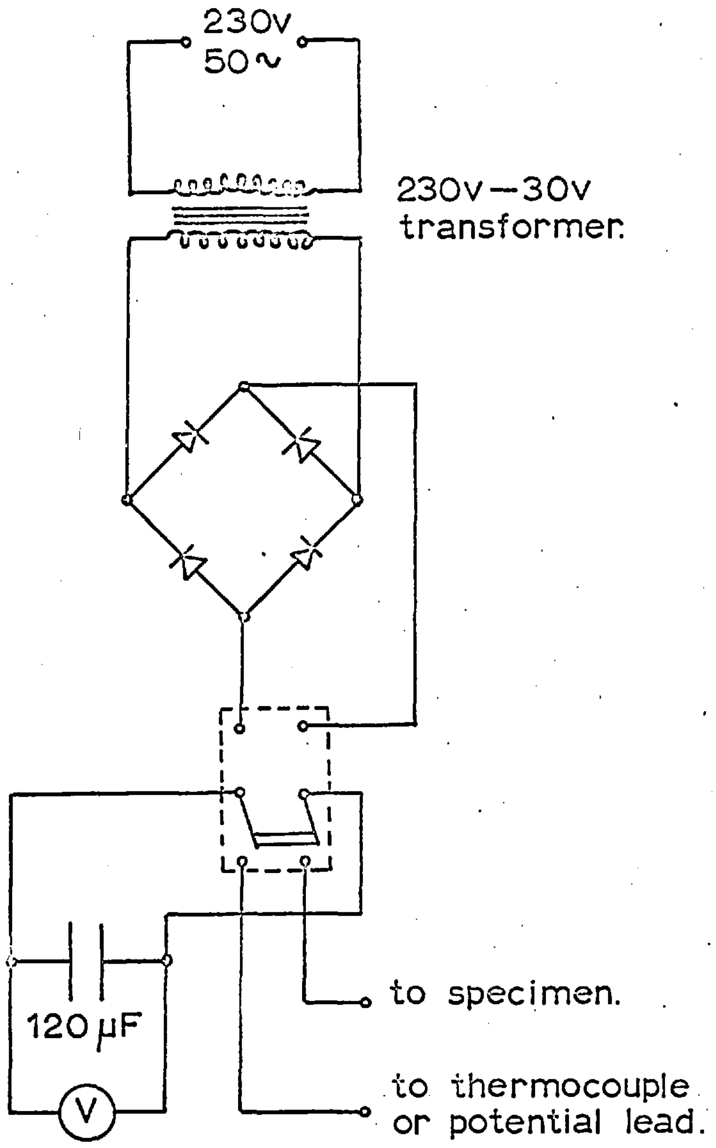

Fig. 13. Circuit diagram of the welding apparatus.

chromel, alumel or iron. The weld is made by the heat produced by the electric current discharged when the condensers were shorted.

Eight welds were thus made, four between the wire specimen and thin wires and four between these thin wires and thicker wires leading from the chamber through the metal-glass-seal. After the welding operation was finished, the top plate was replaced to the vacuum chamber, and by lowering or raising the platform the hook of the pan was engaged with the hook of the specimen assembly.

V. Prevention of Decarburization:

Since martensite transformation is very sensitive to the amount of carbon present in steel it was essential to take the necessary precautions to eliminate the decarburization of the specimen during austenitizing. If decarburization occurred during austenitizing, it manifested itself by an increase in the M_s temperature, by 'rounding off' of the transformation curve near the M_s temperatures and by an increase in the amount of transformation plasticity. It was found that in a vacuum of 10^{-4} torr the decarburization hazard was prevented very successfully. Thus austenitizing the specimens under a very good vacuum made it unnecessary to take further steps to prevent it, such as nickel plating.

The M_s temperatures were reproducible to within ∓ 1°C for ~~some~~ successive heating and cooling cycles. Even after an austenitizing at very high temperature (1200°C), it was possible on a subsequent run at a lower austenitizing temperature (say, 850°C) to obtain an M_s temperature reproducible to within ∓ 2°C of the value characteristic of the lower temperature treatment. Thus in this work it was assumed that the hazard of decarburization was quite successfully eliminated.

VI. Operation of the Apparatus:

The sequence of operation of a typical run may be described in the following order, in two sections;

a) for the study of transformation plasticity,

b) for the study of the stress-strain relationship of austenite.

a) For the study of transformation plasticity:

1. A very carefully prepared specimen was mounted between the pin vices so that the length of the specimen was exactly 6" between the grips. The thermocouple wires and potential leads were spot welded on the wire specimen. The required load was placed on the pan in the chamber. The specimen assembly was placed in the chamber by replacing the top plate. Then by raising or lowering the platform on which the pan was resting the hook of the pan was engaged with the hook of the specimen assembly.

2. The chamber was then pumped down to 10^{-4} torr.

3. The alternating current, direct current and all the measuring circuit leads were connected to the respective terminals on the top plate.

4. The X-Y recorders, the timer, the transducer-meter and the galvanometer were switched on. The potentiometer was checked by the galvanometer, and it was made sure that the cold junction of the thermocouple was at 0°C. The direct current switch was closed and the variable resistance adjusted until a current of 35-40 MA was passing through the specimen.

5. The TPCS switch was closed so that it connected the alternating current circuit and shorted the resistance measuring circuit of the X-Y recorder. The timer was pre-set for the required austenitizing treatment and heating of the specimen was started by pressing the push-button switch of the timer. The temperature of the specimen could be controlled to within ∓ 2°C by controlling the variac.

6. After the required austenitizing treatment, the timer cut off the alternating current, leaving the specimen to cool in vacuum. Now, the TPCS switch was opened to disconnect the alternating current circuit and to connect the direct current and the resistance measuring circuit with the respective terminals.

7. During heating as well as during cooling, the temperature and the length change of the specimen were recorded on an X-Y recorder. The length changes measured during the heating of a specimen proved useful for detecting the A_{c1}-A_{c3} range for the steel and A_s-A_f range for iron-20 wt% nickel alloy for the rate of heating employed.

A continuous record of temperature vs. electrical resistance was obtained on the other X-Y recorder, only during cooling of the specimen. During the recording of temperature the potentiometer was operated to obtain the required trace on the recorder.

8. By the operation of switch S', the stress could be applied to the specimen above, below or at the M_s temperature and the temperature of application or removal could be controlled to within 2°C.

9. The total extent of transformation plasticity in a given experiment was obtained by noting the difference in length at room temperature between a specimen transformed to martensite without stressing and with stressing respectively. Two methods were used for the measurements of length changes occurring during cooling. The transducer method enabled measurements to be made continuously; thus any length change could easily be read from the charts of the X-Y recorder.

A second method of measuring the extension of a specimen resulting from the application of stress involved the use of a cathetometer, placed outside the vacuum chamber, with which the distance between the potential leads was measured to within ∓ 0.05 mm. In order to determine the deformation caused by a stress applied above, below or at M_s temperature and either maintained on cooling to room temperature or removed at any other predetermined temperature, the gauge length was first measured on a fully martensitic specimen at room temperature. The specimen was then austenitized and cooled, the load being applied at a selected temperature during cooling, and the final gauge length being again measured at room temperature. The difference between these two measurements gave the total plastic elongation that occurred under the stress, which is expressed as a percentage of the original gauge length.

10. The M_f temperature was determined either from temperature/dilatation curves or from temperature/electrical resistance curves. When it was determined from temperature/dilatation curves, the specimen was simply heated from room temperature to about 150°C during which the temperature dilatation curve of a fully martensitic structure was obtained. However, if it was determined from temperature/resistance relations, a continuous curve could not be obtained, since resistance readings

could be made only during cooling. Thus the specimen was heated in stages to determine the temperature/electrical resistance relation of a fully martensitic structure.

11. The specimen was then ready for removal and for X-ray, mechanical or metallographic examination.

b) For the study of the stress-strain relationship of austenite from which the yield stress was determined:

1. The specimen was mounted in the usual way between the pin-vices. This time only the thermocouple wires were attached to the specimen. The pan and load assembly was taken out of the chamber, and instead the dynamometer was fixed to the platform. The specimen assembly was placed in the chamber. Then by raising or lowering the platform the hook of the specimen assembly was engaged with the other hook which was incorporated to the free end of the dynamometer.

2. The chamber was then evacuated.

3. The alternating current and measuring circuit leads were connected to the respective terminals on the top plate.

4. The X-Y recorders, timer and transducer-meter were switched on.

5. The TPCS switch was switched on (to the position which connected the alternating current circuit),

and the pre-set timer was started.

6. After carrying out the required austenitizing treatment the switch (s_1) was switched on. This put the timer out of circuit and by controlling the variac the temperature of the specimen (which was registered on one recorder) was brought down to the required temperature. In other words, at the required temperature the cooling of the specimen was stopped by passing an A.C. current of magnitude sufficient to maintain the temperature constant.

7. The switch s' was switched on, starting the reversible A.C. motor to run. During continuous straining the output from the dynamometer was fed to the X-axis and the output from the transducer was fed to the Y-axis of the recorder. Thus a stress-strain curve was directly plotted on the X-Y recorder, from which a yield stress value could be obtained. For measurements of austenite yield stress at temperatures in the range say 330-400°C, the cooling of the specimen was relatively slow, and it was found to be unnecessary to maintain the specimen at constant temperature while stressing; very little accuracy was lost by carrying out the test as the specimen cooled, the relevant part of the test being completed within a range of some 10°C.

8. After considerable yielding was observed, the motor was reversed by the DPCS switch.

9. The specimen was then taken out and the apparatus was ready for the next run.

10. Actually two methods were used for measuring the yield stress of the austenite in the range of temperature from about 330-650°C. The second method involved applying a suitable load at a certain temperature, say 400°C, and using the transducer to determine whether plastic deformation occurred. If no such deformation was observed, a further treatment cycle was carried out, and the load was applied at a temperature 20°C higher than in the first experiment. This procedure was repeated, if necessary, until eventually a temperature was reached at which the stress gave a detectable amount of deformation (viz.appr.0.1%); a yield stress value for this temperature was thus obtained.

VII. Mechanical Properties:

To study the mechanical properties of martensite formed without stressing and with stressing, some tensile and hardness tests were performed on some specimens. To obtain autographic records of tensile tests, i.e. stress-strain diagrams of wire specimens, a Hounsfield tensometer testing machine was used.

In making tensile tests it was important to employ straight lengths of wire, because it was found that small kinks in the wire produced indifferent results, and therefore the specimens with kinks were discarded. To facilitate the specimen mounting a special stand was designed. First the specimen was mounted between the grips on this stand and then it was transferred to the Hounsfield tensometer testing machine. The tests were usually performed on a 4" length of the specimen. Two disc chucks were employed for holding the wire specimen in which the end of the specimen was placed between two brass bushes, then, by means of a screw, the specimen was gripped firmly. No slipping was experienced during testing. Sometimes, the specimen broke near one of the grips and this yielded a very low value of the tensile strength. Therefore, any experiment where the wire broke near the grips was discarded.

For hardness measurements, a microhardness tester or a Vickers pyramid hardness tester was used. Before measurements could be made, specimens had to be mounted either in a cold setting mounting compound or in bakelite. Rough polishing on silicon carbide papers followed by final polishing using diamond dust ~~sheet~~ yielded a bright and scratch free surface.

VIII. Metallographic Examination:

The wire specimens for metallographic examination were mounted either in a cold setting mounting compound or in bakelite. The rough polishing which was done on silicon carbide papers was followed by final polishing on a diamond wheel (0-2 microns). This produced a satisfactory and scratch free surface.

To bring out the structure of the steel, samples were etched for various times in 2 per cent nital or Vilella's martensitic reagent, having the following composition:

Picric acid	1 gm.
Hydrochloric acid	5 ml.
Ethyl alcohol	100 ml.

IX. X-Ray Diffraction Examination:

Some specimens were studied by X-ray diffraction to determine the orientation of the martensite plates that are forming under stress. Experiments with Co K_α radiation were unsuccessful. Mo K_α radiation with an iron filter yielded more satisfactory results. Rotation diagrams were obtained by rotating the wire specimen about its axis by means of an electric motor. The X-ray reflections were recorded on a photographic plate perpendicular to the incident beam. The specimen was placed between the flat film and the X-ray tube. In this way transmission photographs were obtained, in which beams with small reflecting angles were recorded. The preferred orientation was calculated from the recorded circles known as 'Debye-Scherrer circles'.

To make a transmission photograph, the 0.020" diameter specimen was electro-thinned to 0.002" in diameter in an electrolyte. An 'Ellopol' powerpack was used to supply the direct current for thinning, and the electrolyte used and conditions employed were as follows:

Composition:

Orthophosphoric acid . . .	60 ml.
Sulphuric acid	40 ml.

Conditions:

Temperature	50°C
Current density	1.5 A/sq. inch
Cathode	stainless steel (flat)

Thinning was continued in this electrolyte for about 10 minutes.

The X-ray set:

The X-ray set used was a Newton Victor Ltd. Raymax model. The rotation diagrams were obtained at 40 KV and 20 MA in a universal camera, and the exposure times were approximately 14 hours.

CHAPTER 4.

EXPERIMENTAL

RESULTS

4. EXPERIMENTAL RESULTS.

I. Introduction.

The experimental results are most conveniently divided into four groups:

1. The effect of the magnitude of the applied stress and of the temperature of stress application.
2. Experiments to study the effects of austenitizing conditions on transformation plasticity.
   A. Effect of austenitizing temperature.
   B. Short time austenitizing treatments.
   C. Effect of austenite grain size and defect structure.
   D. Yield stress measurements.
3. Experiments to study the structural aspects of transformation plasticity.
4. Experiments to study the effects of stress on the transformation plasticity during the tempering of martensite.

In the first three groups, transformation plasticity during the austenite to martensite transformation is dealt with and in the fourth group transformation plasticity during the tempering of martensite is investigated. Most of the results were obtained by plotting curves of temperature vs. elongation and temperature vs. electrical resistance simultaneously, on two X-Y recorders. The elongation readings of the transducer were confirmed by cathetometer readings.

i) Temperature distribution along the wire specimen.

When a wire specimen is heated for austenitizing it does not attain a constant temperature along its total length because of the vice pins used to hold the specimen in position. The relatively large mass of the vice pins compared with the mass of wire, caused some heat flow from the ends of the specimen. This flow of heat from the ends created a temperature gradient along the specimen; the temperature was highest in the middle zone of the wire specimen and decreased towards the ends of the specimen. Nevertheless, a considerable portion of the middle part of the specimen attained a constant temperature. Fig.14 shows such a temperature distribution along a standard 6" wire specimen when the middle portion is heated to an austenitizing temperature of 850°C. There is a uniformly heated zone in the middle of the specimen. The length of the middle zone where the temperature is uniform is about 3" long, leaving at each end about 1½" of length where the temperature is not uniform. Fortunately, the temperature decreases very rapidly at both ends outside this uniformly heated region. During an austenitizing treatment the complete length of a specimen does not austenitize; there is a region extending about 0.75" in length from the vice pins at

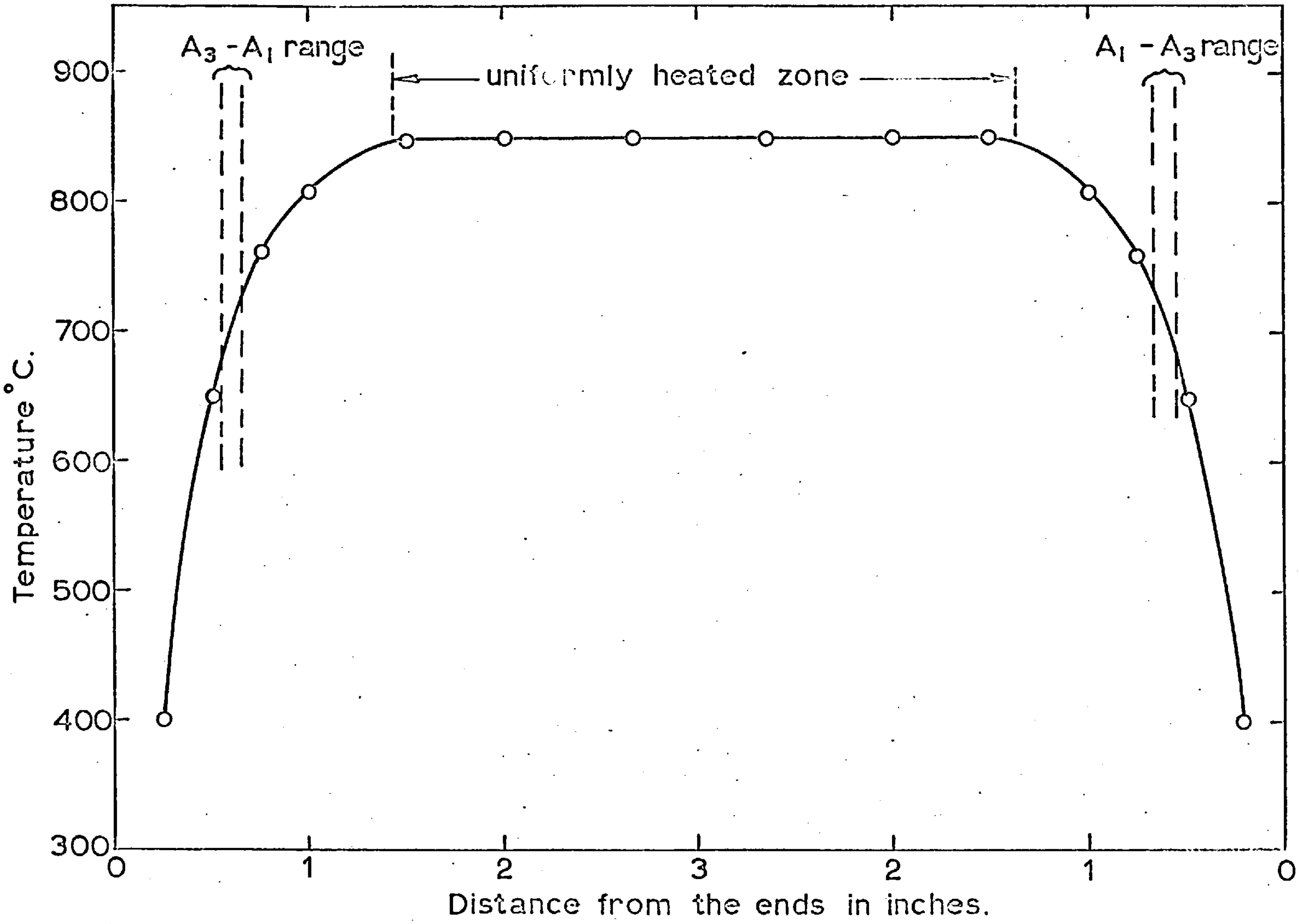

Fig. 14. Temperature distribution along the wire specimen, when the central zone is at 850°C. for En 30A steel.

both ends, where the temperature does not even exceed the A_{c1} temperature, and there is also a region between A_{c1} and A_{c3} where only partial austenitizing is achieved.*

Fig.15 shows the temperature of the central zone of the wire specimen when the temperature at different fixed points away from the grips is at Ms (307°C). From such a curve the maximum length of the specimen which contributes to the (major portion of) transformation plasticity during the application of stress above Ms, can be obtained as follows. When the specimen is stressed at 350°C (about 40°C above Ms) the temperature of a point 1" away from the grip is at 307°C, thus the length of the specimen above the Ms temperature is (6"-2") inches, since the distance between the grips is 6". If the load is applied at this stage, 4" of the specimen length will contribute to the transformation plasticity. On the other hand, if the same specimen is stressed at 307°C (Ms) the temperature of the point 1½" away from the grip is at 307°C and therefore the length of the specimen above the Ms temperature would be (6"-3"=3") inches. Thus a smaller specimen length will contribute to transformation plasticity. The zones between the ends of the grips and points at which the Ms temperature have just

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* The A_{c1} transformation was observed to occur between 695 ∓ 5°C and 720 ∓ 5°C for the En 30 A steel.

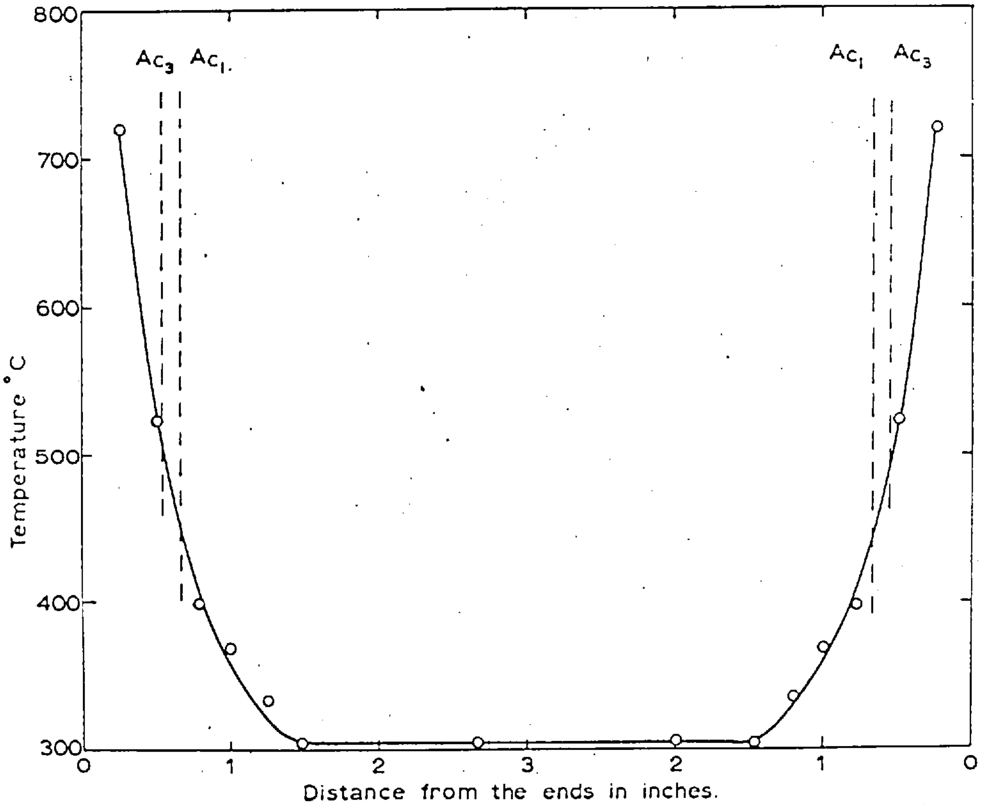

Fig. 15. The centre-zone temperature of the wire specimen, when the temperature at different points away from the grips is at 307°C (M_s).

reached contribute comparatively very little to transformation plasticity since the temperature falls off very rapidly as the grips are approached.

Fig.16 shows the temperature distribution along the wire specimen when the centre zone of the specimen is at M_s temperature. From this figure, it is obvious that, to study the effects of different factors on transformation plasticity, the best temperature of the stress application is when the temperature of the middle zone is at M_s. This is because at this instant all the temperature values outside the uniformly heated middle zone are already below the M_s temperature. Since transformation plasticity has been observed to decrease sharply below M_s, the non-uniformly heated temperature portions will be expected to alter the results only slightly. In the investigation, as mentioned in the previous chapter, by comparison of the results of the transducer and cathetometer techniques, an "effective" gauge length of 3.4 inches (for austenitizing at 850°C) was estimated. A reasonable estimation of an "effective" gauge length from Fig.16 would also give a similar value.

ii) Interpretation of the temperature vs. resistance and temperature vs. length change curves.

If an austenitized specimen cools in vacuum to room temperature the final structure becomes fully

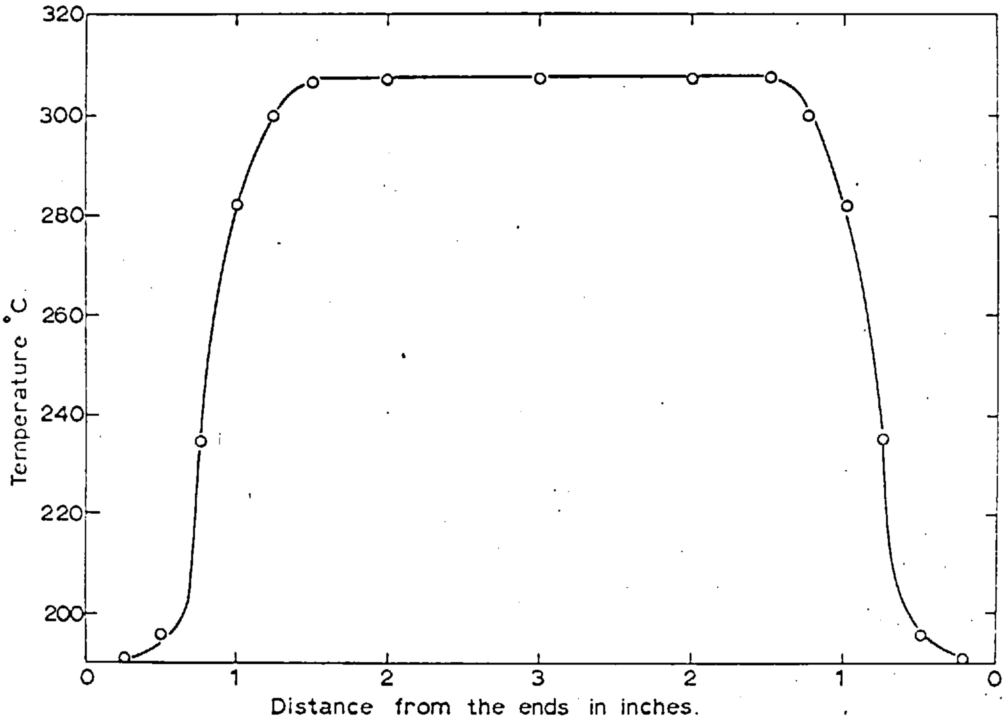

Fig. 16. Temperature distribution along the wire specimen when the centre of the specimen is at 307°C (M_s). for En 30A steel.

martensitic* and the transformation starts at a well defined temperature, designated M_s. During transformation of austenite to martensite there is an associated volume change. As a specimen cools from the austenitic state, it undergoes normal thermal contraction indicated by the linear section of the plot above M_s (Fig.17). This represents (for the En 30A steel) a coefficient of contraction of 23×10^{-6} in/in/°C for austenite. When, however, the M_s temperature is reached, austenite starts to transform to martensite provided that the cooling rate is quick enough to suppress any intermediate transformation to pearlite or bainite. With the start of transformation, a volume change starts as well. The specific volume of martensite is greater than that of austenite, so the increase of volume during transformation manifests itself as an increase in length as well as an increase in diameter of the wire. The increase in length during cooling ceases after the temperature reaches about 235°C, which is approximately 70°C below M_s. With further

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* A small amount of austenite usually remains in low and medium alloy steels at room temperature. Even low carbon steels need to be further cooled below room temperature, for the completion of the transformation to martensite. Keoistinen and Marburger(81) proposed a mathematical expression to describe the course of transformation in steels, which does not predict an M_f point. In the present investigation with X-ray studies it was found that there is always a small amount of retained austenite present at room temperature.

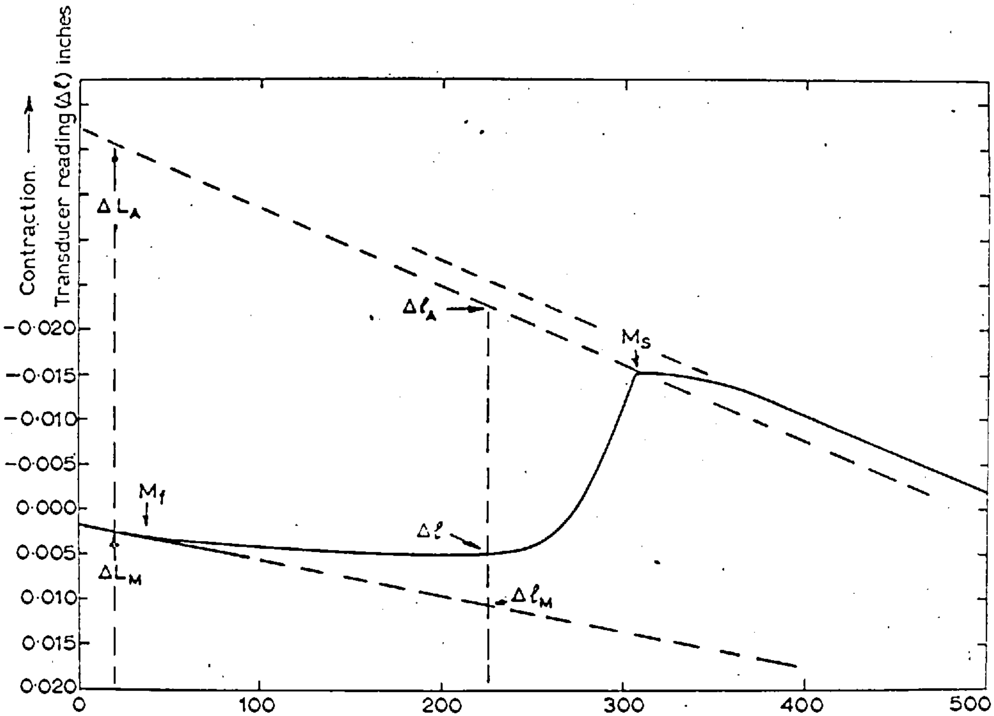

Fig.17. Typical temperature length-change curve.

cooling down to about 150°C, extension due to martensite formation is counterbalanced by thermal contraction and there is essentially no length change as the temperature falls. This is illustrated in Fig.17, where a typical temperature vs. length change curve of a vacuum cooled specimen (without an application of a stress) is reproduced. Below about 150°C, the specimen starts to contract again, indicating that the elongation due to the transformation is smaller than the thermal contraction which is taking place simultaneously. Below about 40°C, which is the M_f temperature for the En 30A steel, the contraction is solely due to the cooling. The two straight lines, shown in Fig.17, represent the thermal contraction (expansion) of completely austenitic and completely martensitic structures respectively. It will be noticed that the thermal contraction line of austenite deviates from linearity as the M_s temperature is approached. This is due to the end-effects of the specimen. The volume change which accompanies the transformation of the end zones causes the curve to deviate from straight line. To correct the curve for this end-effect, a straight line from the M_s point (where there is a sharp break in the curve) is drawn parallel to the linear portion of the austenite contraction curve. Using the corrected curve, the total elongation due to the volume change during transformation without any application of stress has been found to be about 1.43% for the En 30A steel and about 1.5% for the iron-20% nickel alloy. These values are found by assuming that the thermal contraction curve of austenite may be extrapolated down to the room temperature; then the percentage expansion is obtained by the simple proportion from Fig.17.

$$\% \text{ Expansion} = \frac{\Delta l_T}{l} \times 100\%$$

where $\Delta l_T = (l_A - l_M)$ the difference in length between fully austenitic and fully martensitic structures when referred to room temperature, and,

l = The total effective length of specimen.

The results obtained for percentage extension during transformation without any applied stress agree very well with the results obtained by other investigators(82); for example, it was shown that the increase in volume when austenite transforms to martensite, is comparatively insensitive to the carbon and alloy content(82) and turns out to be about 4.6% when referred to room temperature. de Jong and Rathenau(28) have shown that $\Delta l/l$ is equal to one third of $\Delta V/V$ during the γ to α transformation in pure iron. So in the present investigation the volume change during transformation for the En 30A steel would be 3x1.43 = 4.3%, and for the iron-20% nickel alloy, this value would be $3 \times 1.5 = 4.5\%$; these values are very close to the above suggested $^{\Delta V}/V$ values found by other investigators.

The reproducibility of the data obtained using the dilatometric method was excellent and the deviations obtained for $\frac{\Delta l}{l}$ values were within $1.43 \mp 0.01\%$ for the En 30A steel and $1.5 \mp 0.01\%$ for the iron-20 wt% nickel alloy.

The above mentioned elongation takes place during the austenite to martensite transformation, but when a fully martensitic structure is austenitized and then brought back to room temperature with a cooling rate that produces a fully martensitic structure, theoretically there should not be any change in the final length of the specimen, i.e. the volume change (contraction) during the heating of (tempered) martensite into the austenite range should be cancelled by the increase of volume during the reverse transformation. However, it was found that a small permanent extension occurred (up to about 0.06%) when a martensitic specimen was austenitized and then cooled to produce martensite. The total extent of plastic deformation with the application of stress during the austenite to martensite transformation was obtained by noting the difference in the length at room temperature between a specimen transformed to martensite without stressing and with stressing respectively. This is explained in full detail in the previous chapter on experimental technique. The term 'total extension, percentage' as used in the graphical presentation of the results (e.g. Figs.21,22) refers to transformation plasticity measured in this way, and expressed as a percentage of the effective gauge length.

The results for various austenitizing temperatures on the En 30A steel as shown in Figs.27,28 are based only on cathetometer measurements, since the 'effective' gauge length was only determined for austenitizing at 850°C, and it might be expected to vary with austenitizing temperature. In experiments involving austenitizing at 850°C, measurements were made by both the cathetometer and transducer methods; the results shown in the graphs are based on the transducer measurements, expressed on the basis of the 'effective' gauge length.

In the experiments involving the iron-20 wt% nickel alloy the measurements of transformation plasticity for various austenitizing temperatures were made by the transducer technique. It was found that the 'effective' gauge length was the same as that determined for the En 30A steel for austenitizing at 850°C. For higher austenitizing temperatures, the effective gauge length was found to change slightly; e.g. it increased from 3.4" to 3.7" with an increase in austenitizing temperature from 850°C to 1200°C. The increase in effective gauge length was taken into account in the calculation of total extension percentage.

The reproducibility of the results for total extension percentage was good. An indication of the scatter obtained in the measurements of the total extension percentage is shown in Fig.21, e.g. for a stress of 20,000 lb/in^2 applied at M_s. Results of replicate experiments lay between 1.3 and 1.5%, while at higher stress levels, the scatter was nearly twice as great.

In this investigation, two methods, based on the change of some physical property with transformation, have been used to ascertain the progression of martensite formation; namely,

a) Dilatometry,
b) Electrical resistance method.

a) Dilatometry method:

The dilatation measurements of Fig.17 were converted to percentage transformation, by assuming that the percentage of total length difference between pure austenite and completely martensitic structure at a given temperature is equal to the volume percentage martensite present. Thus the percentage transformation is obtained by simple proportion from Fig.17 as follows:

$$\% \text{ martensite} = \frac{\Delta l_A - \Delta l}{\Delta l_A - \Delta l_M} \times 100\%$$

where Δl_A = length (extrapolated value) of the specimen in the fully austenitic condition.

Δl = length of specimen at the required temperature.

Δl_M = length (extrapolated value) of the specimen in the fully martensitic condition.

b) Electrical resistance method:

By virtue of the change in electrical resistance during the austenite to martensite transformation it was possible to follow the kinetics of transformation. The change in electrical resistance of a fully austenitic and a fully martensitic specimen with temperature is fairly linear although the slopes of these two straight lines differ from each other. Austenite has a higher specific electrical resistivity than martensite at any temperature, so that with the start of transformation, the electrical resistance of the specimen decreases. At the M_s temperature there is a sharp break in the linearity of the temperature/resistance curve of austenite. On further cooling the specimen below the M_s temperature, more martensite is formed and thus the electrical resistance drops further. In Fig.18 a typical temperature-resistance curve of a specimen, austenitized and cooled freely in vacuum to form martensite, is shown.

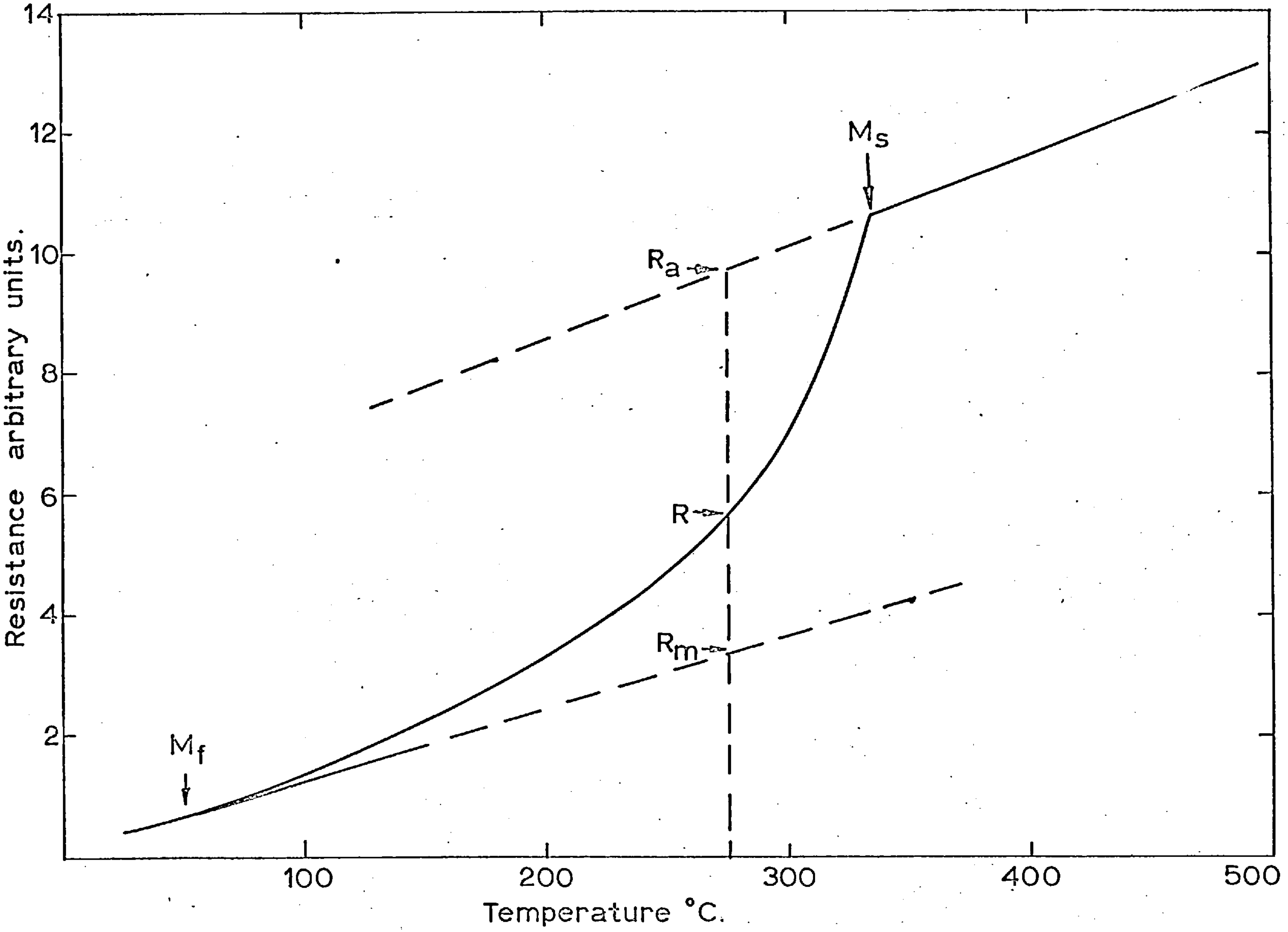

Fig.18. Typical temperature resistance curve.

The change in electrical resistance was recorded in terms of the potential drop across a gauge length of 2" of specimen. The potential measuring circuit had a large external resistance in series with the specimen, so that the current which was flowing through the circuit remained fairly constant during the transformation cycle. Thus the measured potential drop across the gauge length was directly proportional to the resistance of the specimen.

To convert the change in electrical resistivity to percentage transformation, it was assumed that the percentage of total resistance difference between austenite and completely transformed martensitic structure at a given temperature is equal to the volume percent of transformed martensite phase present. By using the approximation that the resistance of retained austenite may be found by extrapolating the resistance curve of untransformed austenite below M_s, then the percentage transformation at any temperature T_i below M_s is obtained by a simple proportion as follows (from Fig.18)

$$\%\ \text{Martensite} = \frac{Ra - R}{Ra - Rm} \times 100\%$$

where 'R' = the resistance of the specimen at the required temperature 'T_i',

'R_a' = the extrapolated value of resistance of pure austenite at the required temperature 'T_i'.

'R_m' = the resistance of the fully martensitic structure at the required temperature 'T_i'.

The agreement between the 'Dilatometry' and 'Electrical resistance' methods was excellent and the scatter was always within 2%, e.g. after an austenitizing treatment at 850°C, the amounts of martensite formed at 305°C and 300°C were found to be 8% and 20% by the resistance method, and 9% and 22% by the dilatometric method respectively.

In the later stages of transformation, the transformation rate decreases considerably, and at room temperature usually a small amount of austenite remains untransformed. Therefore, the accuracy of the observed M_f temperature will depend on the sensitivity of the measuring technique employed. In this investigation, the M_f temperature was obtained from either temperature-resistance or temperature-dilatation curves. To obtain the M_f temperature from the temperature-dilatation curve the specimen in the fully martensitic structure was reheated from room temperature and the point where it deviated from the original cooling curve was noted. In the temperature-resistance case, another cooling curve was obtained by reheating the fully martensitic structure and recooling to obtain a temperature-resistance relation of the fully martensitic specimen and noting the point where the previous cooling curve coincided with the latter, and taking the meeting point as the M_f temperature. In calculating the progress of transformation, the amount of retained austenite was not taken into account.

The electrical resistance of the fully martensitic steel was measured only up to 100°C, since above this temperature tempering occurred rapidly and this lowered the electrical resistance considerably. However, for the iron-20 wt% nickel alloy this curve was obtained up to 250°C. The slight autotempering of the first formed martensite plates in the En 30A steel could not be avoided when cooling the specimens in vacuum.

iii) Transformation characteristics.

The as-received structure of the En 30A steel wire was tempered martensite. Once the specimen was austenitized and then allowed to cool in vacuum the structure became ~~completely~~ martensitic. The M_s temperature for the first treatment was 2-3°C higher than those observed on subsequent cycles, when the specimen was austenitized at the same temperature.

The M_s temperature of a steel is very sensitive to the chemical homogeneity of the austenite. Since the initial structure has a marked effect on the attainment of homogeneous austenite(83) it will have a distinct effect on the subsequent transformation. Therefore for the attainment of a homogeneous austenite, the specimens were initially brought to the fully martensitic condition and then for the subsequent cycle rapidly heated (about 100°C/sec.) to the required austenitizing temperature. A rapid heating rate was employed to reduce the time available for tempering of martensite and hence to facilitate the attainment of homogeneity in austenite.

Experiments were made to determine the effect of austenitizing times at a particular temperature, on the subsequent transformation to martensite. It was found that, when the En 30A steel (in the as-received condition) was austenitized at 1000°C for progressively longer times, the M_s temperature on subsequent quenching in vacuum showed a downward trend until about 2½ minutes of austenitizing time was reached. For times longer than 2½ minutes a constant value for M_s temperature was obtained. This suggests that all the carbides were taken into solution in austenite at 1000°C after 2½ minutes. For the subsequent cycles the austenitizing times above 30 seconds at 850°C yielded constant M_s values. In this investigation the longest austenitizing time employed was ½ hour.

On the basis of these experiments, all samples were initially treated for 5 minutes at 1000°C and were then cooled to produce martensite. After this initial homogenizing treatment, martensitic samples were rapidly heated (at about 100°C/sec.) to the required austenitizing temperature. Standard austenitizing times of 3 minutes and 100 seconds for temperatures in the range 800°-1000°C and above 1000°C respectively were used; these being considered to produce an essentially homogeneous austenite in the En 30A steel from the standpoint of chemical composition. Although it is not possible to know whether these conditions produced a completely homogeneous austenite in the En 30A steel, longer times of austenitizing led to no change in the subsequent transformation and in transformation plasticity. Under these conditions the values of M_s temperature were reproducible within ∓1°C for the same specimen for a number of cycles of austenitizing and cooling. From specimen to specimen the M_s temperature for a given treatment agreed within ∓ 2°C.

Measurements of length changes occurring during heating indicated that for the heating rate used (100°C/sec.) the A_c transformation for the as-received structure occurred between 725°C and 750°C for the En 30A steel. On reheating the same specimen with the same rate of heating from the martensitic state, the transformation to austenite occurred in the temperature range of approximately 695°-715°C. The change from austenite to martensite in the iron-20 wt% nickel alloy appeared to occur largely in the temperature range 550°-575°C.

II. The effect of the magnitude of the applied stress and of the temperature of stress application.

The effects on transformation plasticity of tensile stresses, within the range 200-53,000 lb/in^2, were studied for selected temperatures of stress application, either above, below or at M_s temperature for a standard austenitizing treatment at 850°C. It was found that for stresses less than 1000 lb/in^2, applied to the En 30A steel (at M_s) during cooling, the observed extension was not more than 0.06%, which is the extension obtained after a cycle of heating (for austenitizing) and cooling (to produce martensite) without the application of any stress. It was not possible, therefore, to be certain whether any transformation plasticity occurred with such stresses. Stresses of 1000 lb/in^2 or more when applied at or above M_s gave detectable transformation plasticity; e.g. for a stress of 1000 lb/in^2 applied at the M_s temperature approximately 0.1% permanent total extension occurred. With increasing values of stress, the amount of deformation increased.

When stresses greater than the yield stress of the austenite were applied above M_s and maintained during cooling through the martensite range, the total extension is made up of the plastic deformation that occurs above

M_s due to the stress, and the transformation plasticity associated with the main bulk of martensite formation. Interpretation of the results is then rather complex since the application of such stresses between M_s and M_d induces some martensite formation which may assist yielding by virtue of a transformation plasticity effect. The M_d temperature was estimated by Wells$^{(79)}$ to be approximately 70°-80°C above M_s. Above the M_d temperature only slip occurs, but there is no transformation to martensite.

Although the total extension obtained during cooling of a specimen under a stress to room temperature is greater than that found when the same stress is applied at M_s, experiments reported below showed that the extent of the main transformation plasticity is somewhat reduced by the deformation of austenite that occurs above M_s. Thus, when studying the effects of stresses greater than the yield point of austenite, it is desirable to apply the stresses at the actual M_s and not above it, and this was done in some experiments. With stresses less than the yield stress of austenite, the same results were obtained when the stress was applied at or above M_s.

Fig.19 shows a typical curve which is obtained by the transducer technique, for the change in length

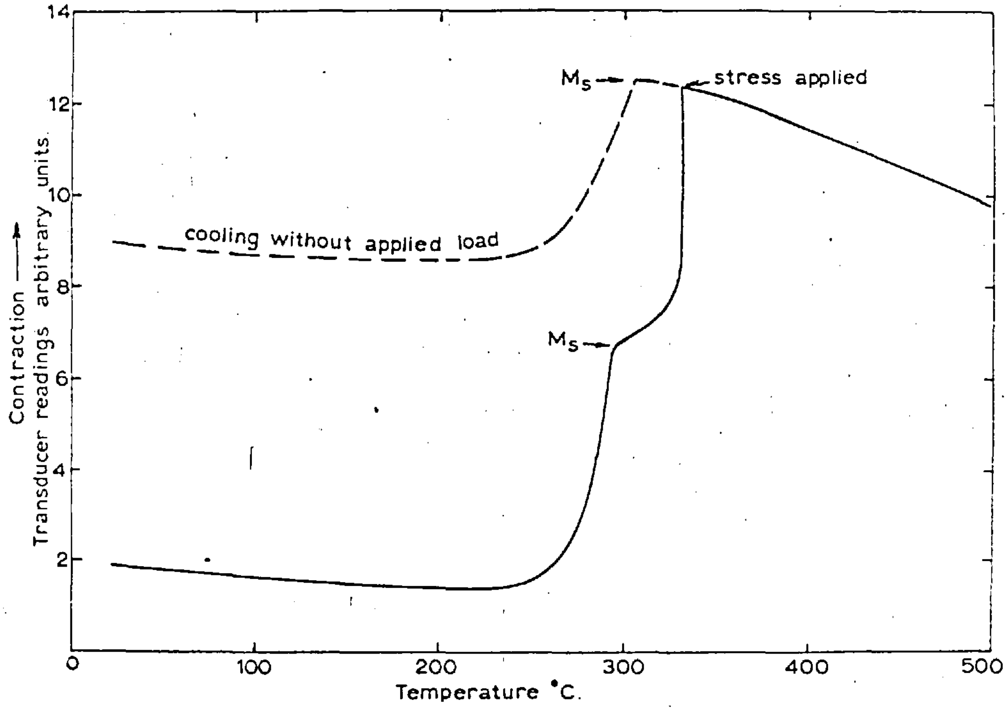

Fig. 19. Temperature vs. transducer readings for specimen stressed above M_S with a stress greater than the austenite yield stress.

of a specimen during cooling to room temperature when a stress greater than the yield stress of austenite was applied above M_s and maintained throughout cooling. When the stress is applied, a marked and sudden extension occurs, associated with the formation of a small amount (burst) of martensite. With decreasing temperature, the specimen continues to elongate, although at a reduced rate. On cooling through M_s there is a further sudden extension of the specimen, which represents the beginning of the main 'transformation' plasticity, and this extension continues as the temperature falls, but at a decreasing rate. It is important, when interpreting the length changes that occur during cooling, to note that the curves include contributions from thermal contraction and from the volume expansion due to martensite formation, as well as from the transformation plasticity.

Fig.20 shows a similar curve, obtained by the transducer technique for the change in length of a specimen during cooling to room temperature when a stress less than the yield stress of austenite was applied above M_s and maintained throughout cooling. When the stress is applied, a small increase in length occurs because of the elastic yielding of the specimen. With falling temperature, the specimen contracts (thermal contraction) as if there were no stress present, until the M_s temperature is reached. The sudden extension of the specimen which

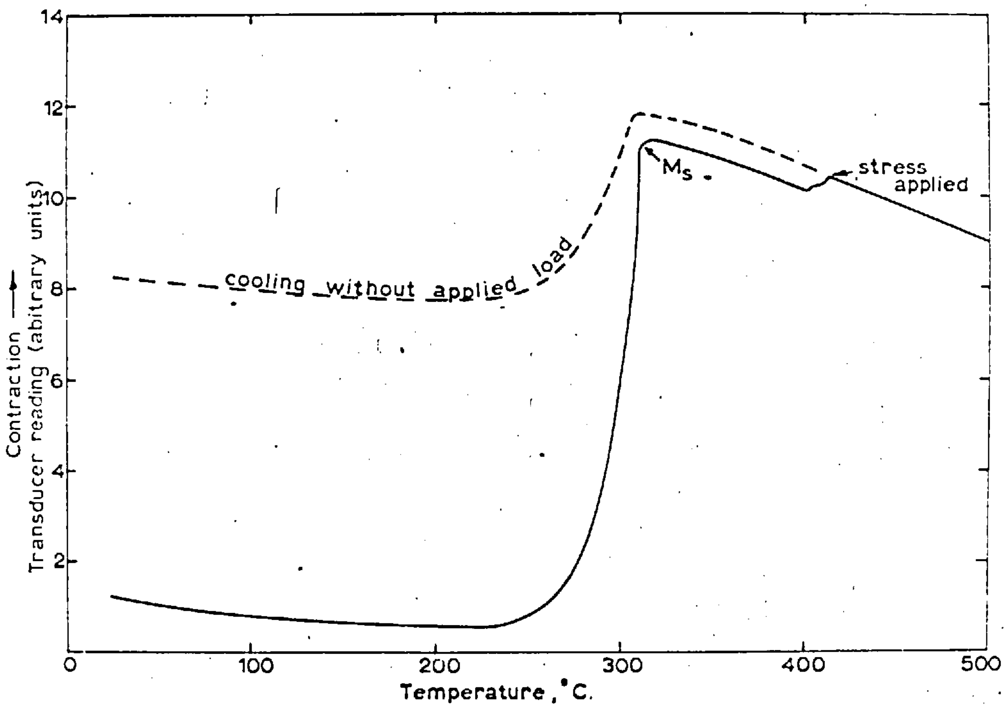

Fig. 20. Temperature vs. transducer readings for specimen stressed above M_s with a stress less than austenite yield stress.

occurs with the beginning of transformation at the M_s temperature is much slowed down at 20°-25°C below M_s.

The amount of transformation plasticity increases with the increase in applied stress; Figures 21 and 22 show this increase in transformation plasticity with the increased amount of applied tensile stress in the En 30A steel and in the iron-20 wt% nickel alloy specimens respectively. For example, a stress of 5,000 lb/in^2, when applied to the En 30A steel specimen at the M_s temperature, produced 0.3% total extension, and this increased to 10% as the amount of stress is increased to 50,000 lb/in^2. Similarly, a stress of 4,000 lb/in^2, when applied to an iron-20 wt% nickel alloy specimen at the M_s temperature, produced 0.5% total extension. An increase in the amount of applied stress to 23,000 lb/in^2 produced 6% total extension. The results of Figs. 21 and 22 show that there is a straight line relationship between the total extension percentage and the magnitude of the applied stress in the lower range of stresses. The departure from linearity occurs at about 21,000 lb/in^2 in the En 30A steel and at about 13,000 lb/in^2 in the iron-20 wt% nickel alloy for stresses applied at or above M_s.

The other observation which has been found to affect the transformation plasticity is the temperature

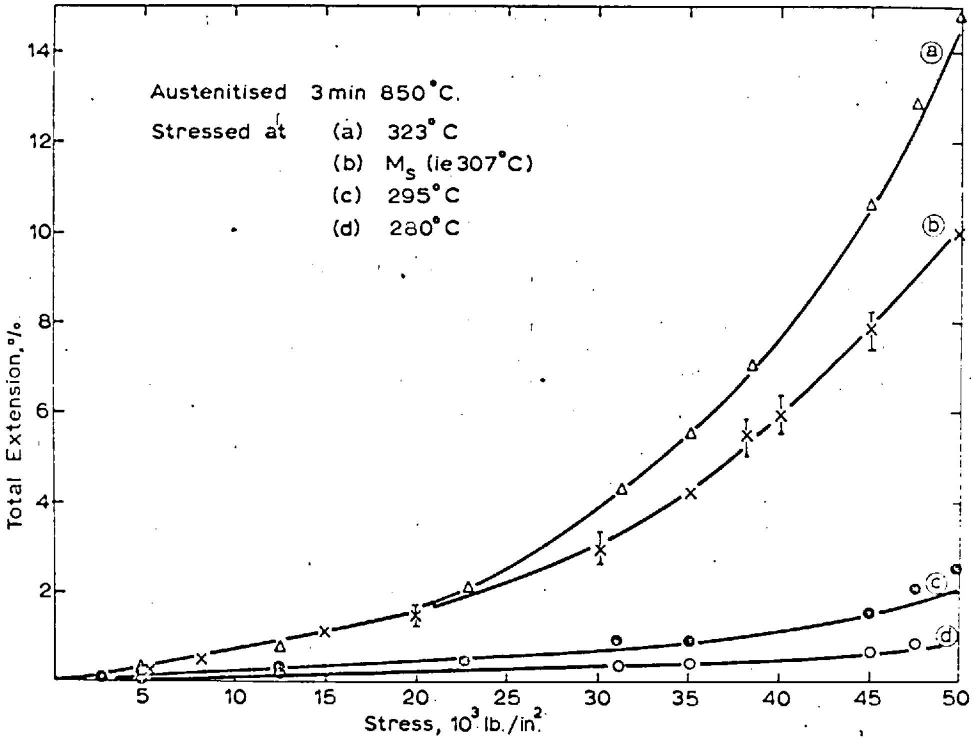

Fig. 21. Effect of magnitude of applied stress on total extension for En30A steel

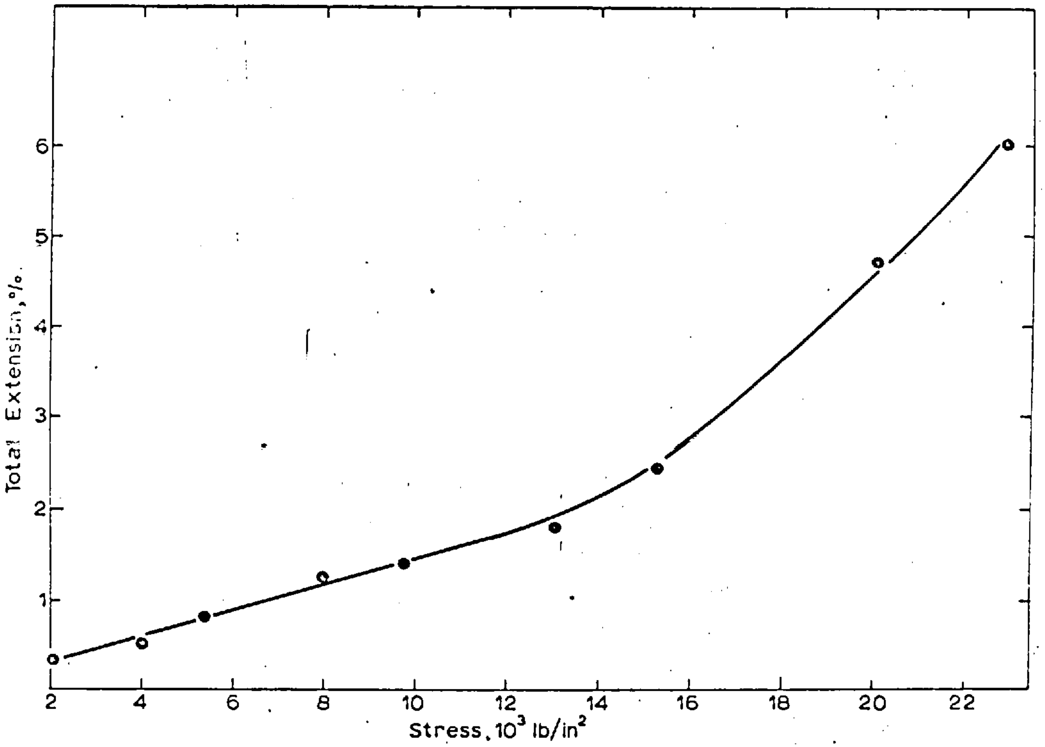

Fig.22. Effect of magnitude of applied stress on total extension for Fe-20wt%Ni alloy stressed at M_s. (ie 251°C).

at which the specimen is stressed. When the stressing temperature was decreased below the M_s temperature, the amount of plasticity decreased. Thus, for the En 30A steel, a stress of 48,000 lb/in^2 produced about 0.8% total extension when the stress was applied at 280°C. An increase to 323°C in the stressing temperature caused an increase in extension from 0.8% to 14.8%.

Fig.23 shows the amount of extension induced by stresses applied to the En 30A steel at different temperature levels below, above and at M_s temperature. For each stress level studied there is a sharp break on the curves at the M_s temperature. In the lower range of stresses, no extra increase in total extension was observed when the specimen was stressed above M_s instead of at M_s temperature. However, larger stresses do show an increase in total extension as the specimens are stressed above the M_s temperature. The results of Fig.21 show that for a given temperature of stress application (e.g. curve 'a') there is a straight line relationship between the total extension percentage and the magnitude of the applied stress in the lower range of stresses. The slope of the straight line is dependent on the temperature of stress application. The slopes increased as the temperature of stress application was increased from 280°C up to the M_s (viz. 307°C) but no further increase of slope was observed when the temperature of stressing

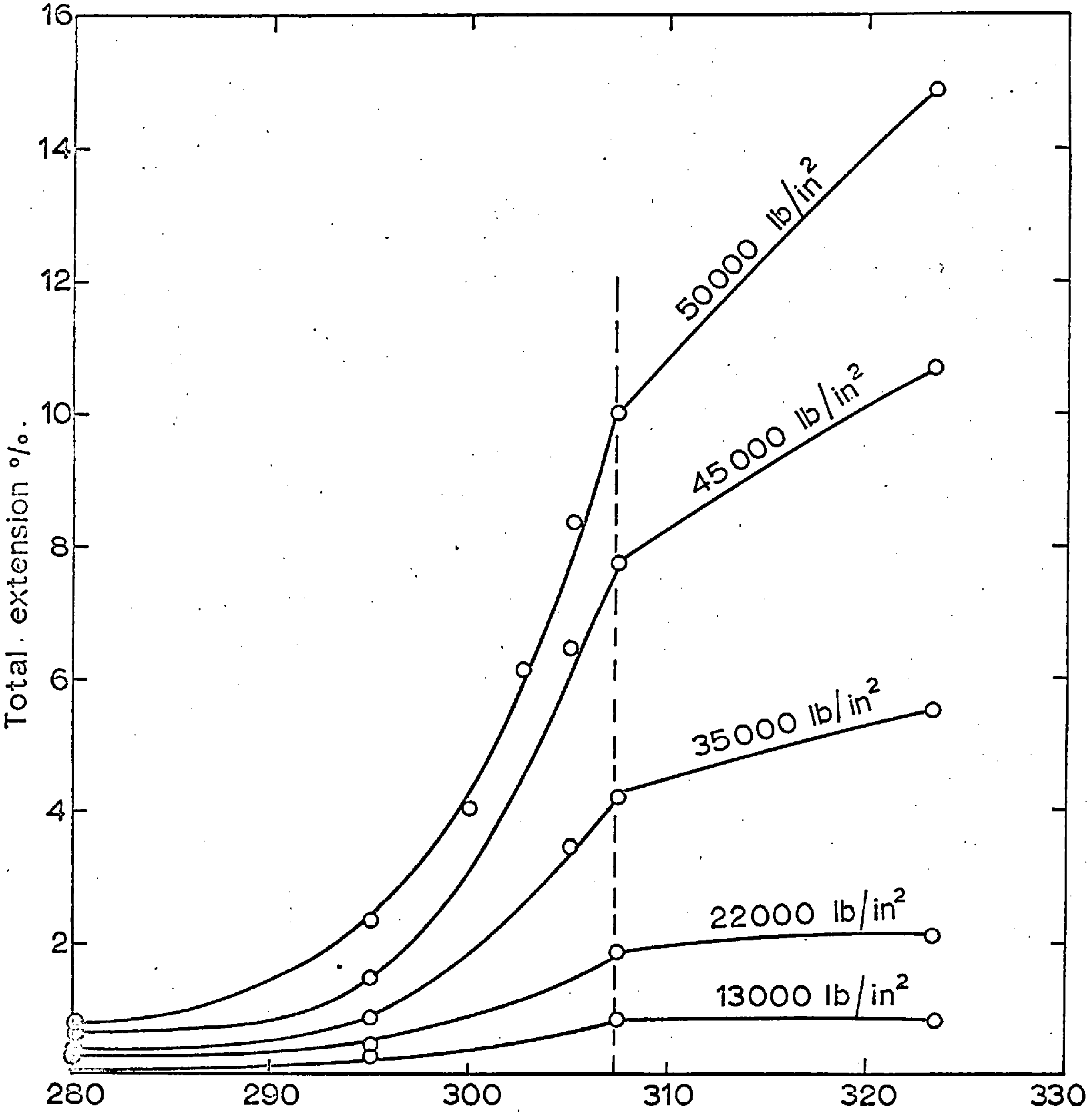

Fig. 23. The effect of stressing temperature on the total extension %, for En 30A steel; austenitized 3 min at 850° C.

was increased to 323°C. The plotted data for the various temperatures showed a departure from linearity beyond a certain stress level. The slopes of the curves increased progressively, the effect being particularly marked when the stresses were applied at or above M_s.

The departure from linearity occurred at higher stress levels when the stressing temperature was below M_s; e.g. at about 35,000 lb/in^2 at 295°C and at about 40,000 lb/in^2 at 280°C.

When the stress was applied to the En 30A steel at temperatures below about 270°C (viz. approximately 40°C below M_s), no transformation plasticity was observed; at this temperature about 70% of martensite had formed during normal unstressed cooling. The iron-20 wt% nickel alloy is very similar to the En 30A steel in this respect. In this alloy the effect of stress also seemed to cease after about 70% martensite had formed.

The curves of Fig.24 have been drawn from the measured transducer results of the En 30A steel, making corrections for thermal contraction and for the volume expansion due to martensite formation. These curves show that the major part of the deformation takes place within the 5°C below M_s during which about 70-80% of total deformation is completed. With decreasing temperature the specimen continued to elongate, but at a rapidly

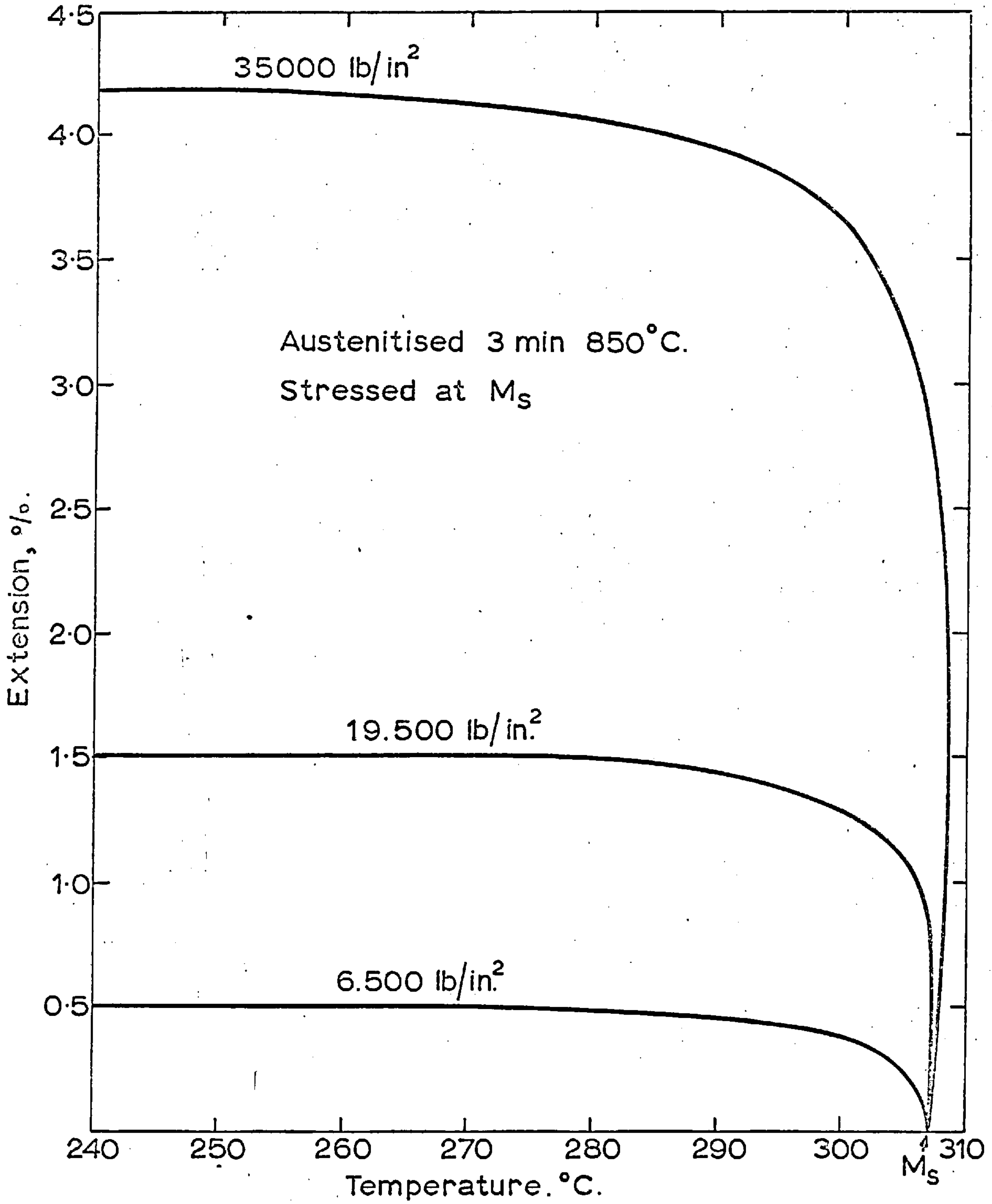

Fig. 24. Actual extension occuring during transformation for En30A steel.

reducing rate. The transformation plasticity was completed (within the accuracy of measurements) when the temperature had fallen to about 260°-270°C.

As shown in Fig.21 (curves 'a' and 'b'), when a stress in the non-linear portion of the graph is applied above M_s, the total amount of extension that occurred was greater than that when the same stress was applied at M_s; this being due to the occurrence of plastic deformation of austenite for stresses greater than the austenite yield stress. Experiments were made on the En 30A steel specimens to investigate the effects of such plastic deformation on the subsequent main transformation plasticity, and the results are shown in Figs. 25A and 25B. Curve 'a' shows the total plasticity found when stresses ranging up to 50,000 lb/in^2 were applied at 323°C (i.e. 16°C above M_s) and were maintained down to room temperature. Curve 'b' shows the values of total extension obtained when the specimens were stressed at M_s. The results of curve 'c' represent the total extension resulting from applying the stress at 323°C and removing it at M_s. The difference between curves 'a' and 'c' gives the amount of transformation plasticity associated with the main martensite formation under conditions when the specimen was stressed at 323°C with stresses greater than the austenite yield stress. This "difference curve" is curve 'e' of Fig.25B, and is seen

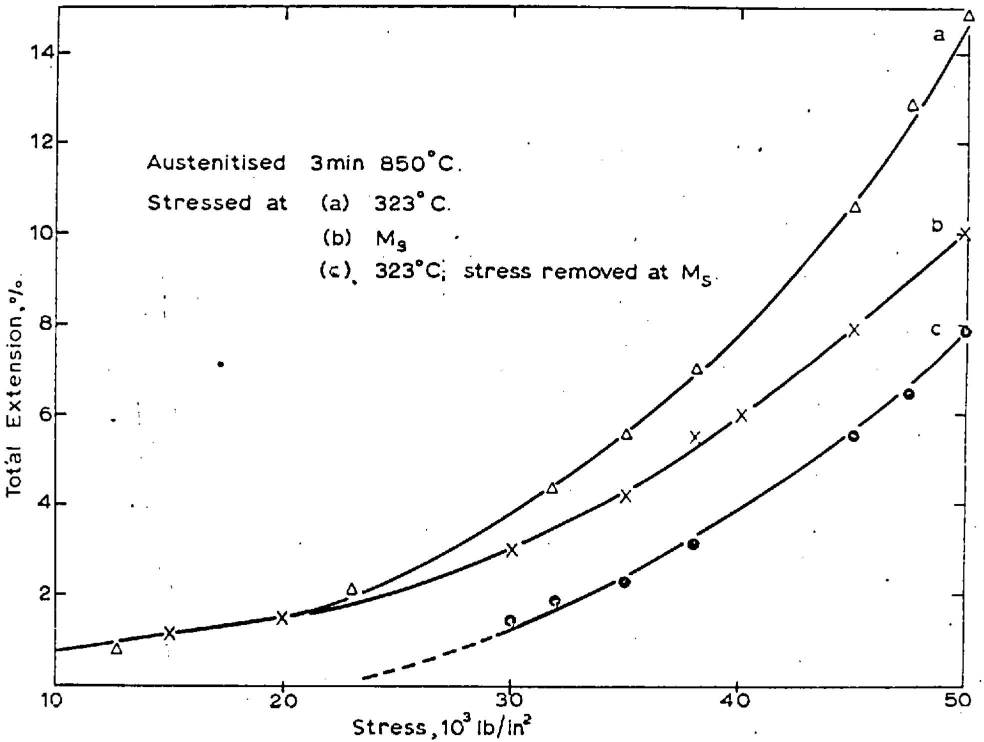

Fig. 25A. Effect of stressing temperature on total extension for En 30A steel.

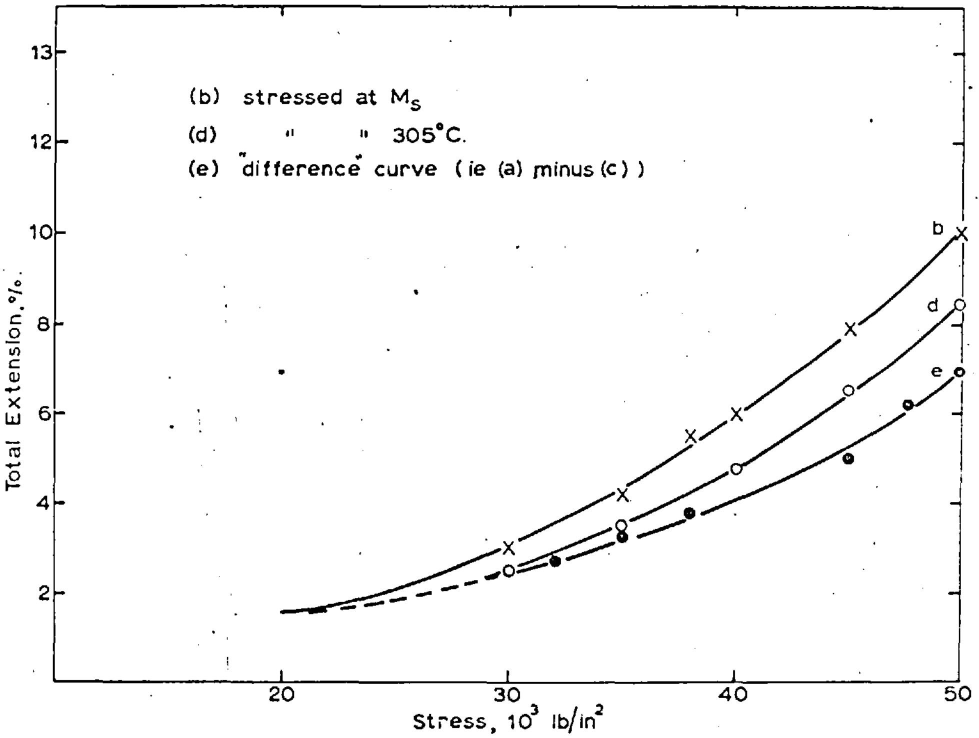

Fig.25B. Effect of stressing temperature on total extension for En30A steel.

to lie significantly below curve 'b'; this shows that when stresses that cause plastic deformation of austenite are applied above M_s the effect is to reduce the amount of transformation plasticity that occurs on cooling through M_s. This reduction in transformation plasticity by the application of the stress above M_s is considered to be due to the strain-hardening of the austenite, and to the formation of a certain amount of martensite. The effect of the existence of a relatively small amount of martensite, viz. 7%, in reducing transformation plasticity is illustrated by curve 'd' which shows the amount of plasticity following the application of a stress 2°C below M_s. This amount of martensite is about the same as that which forms at 323°C when a stress of 50,000 lb/in^2 is applied.

The amount of transformation plasticity decreases when the austenite is thermally stabilized. This is shown in Fig.26. The En 30A steel specimens were thermally stabilized by arresting cooling at different temperature levels below M_s for certain isothermal holding times. The specimens were stressed after vacuum cooling was restarted with a stress of 50,000 lb/in^2. The amount of plasticity decreased as the amount of stabilized austenite increased with increasing isothermal holding time. Thus a stress of 50,000 lb/in^2, when applied at 302°C without interrupting the cooling,

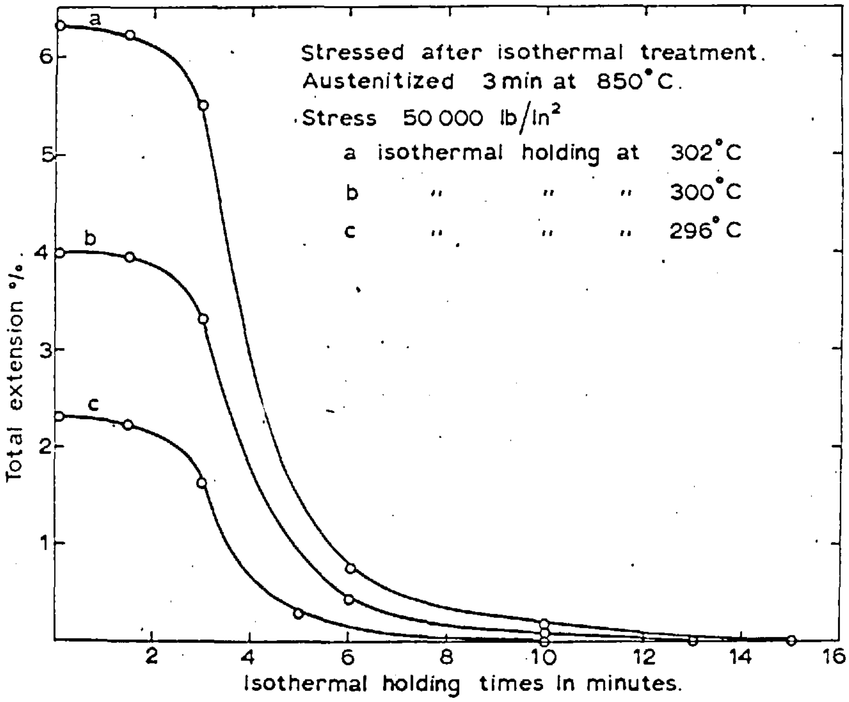

Fig. 26. The effect of thermal stabilization on the total extension for En 30A steel.

produced 6.3% deformation. The same stress, when applied at the same temperature after an interrupted cooling for 90 seconds at 302°C, produced 6.2% deformation and this decreased to 5.5% when the time was increased to three minutes. It is believed that after three minutes of isothermal holding time, bainite formation started.[*] Above the isothermal holding times of 10, 13 and 15 minutes at the temperatures of 296°C, 300°C and 302°C respectively, no indication of plasticity was observed. This proves that within this time a continuous phase must have been formed.

Experiments were also made to determine the effect of cooling rate on transformation plasticity. Specimens of the iron-20 wt% nickel alloy and the En 30A steel were cooled in vacuum until the M_s temperature was reached and then stressed and cooled by an air stream. The results obtained from (quick cooling) air cooled and (slow cooling) vacuum cooled specimens were identical and well within the experimental error. Thus it can be concluded that for the rates of cooling employed, there was no effect of cooling rate on transformation plasticity.

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[*] It is fairly well established that the incubation period of bainite transformation is reduced in the presence of martensite.

III. The effects of austenitizing conditions on transformation plasticity:

i) Effect of austenitizing temperature:

Experiments were made to determine the effect of austenitizing temperature on transformation plasticity. Austenitizing treatments were carried out at various temperatures in the ranges 800-1200°C and 600-1200°C for the En 30A steel and for the iron-20 wt% nickel alloy respectively. For austenitizing temperatures at and below 1000°C a time of three minutes was used and only 100 seconds was allowed for higher temperatures. These were the standard times adopted. Stresses ranging up to 53,000 lb/in^2 were used with two different temperatures of stress application. In one series of experiments on the En 30A steel, a temperature of 330°C was chosen for stress application, bearing in mind that the M_s temperature increases as the austenitizing temperature is increased, this temperature being above the M_s temperature for all the austenitizing treatments. In other experiments on the En 30A steel and the iron-20 wt% nickel alloy specimens, stresses were applied at the M_s temperatures for each of the austenitizing treatments.

The measurements of total extension percentage for various austenitizing temperatures on the En 30A steel specimens, when the loads were applied at 330°C, are shown in Fig.27. It is seen that for a given stress the extension increases as the austenitizing temperature is increased; also for a given austenitizing temperature, the extension increased with stress. The increase in plasticity with increase in austenitizing temperature is particularly marked at the higher stress levels; thus, for a stress of 53,000 lb/in^2, an increase of austenitizing temperature from 850°C to 1200°C gave an increase in extension from approximately 16.3% to 24.8%.

Specimens stressed with large stresses showed a rise in temperature (i.e. 2-5°C depending on the temperature of stressing and amount of stress) when stressed above M_s. This was accompanied by a burst of martensite and a sudden permanent extension. After this initial recalescence, the temperature curve followed a path parallel to the continuation of the cooling curve just before the load was applied. But when the M_s temperature was reached, the temperature of the specimen increased markedly after which a normal cooling was resumed again. The results of temperature measurements on the En 30A steel specimens showed that with increasing austenitizing temperature and stress, the rise in temperature at M_s increased. But, for a given austenitizing temperature and stress value, the recalescence at M_s increased as the temperature of stress application was decreased towards M_s, and at the M_s temperature, the maximum recalescence

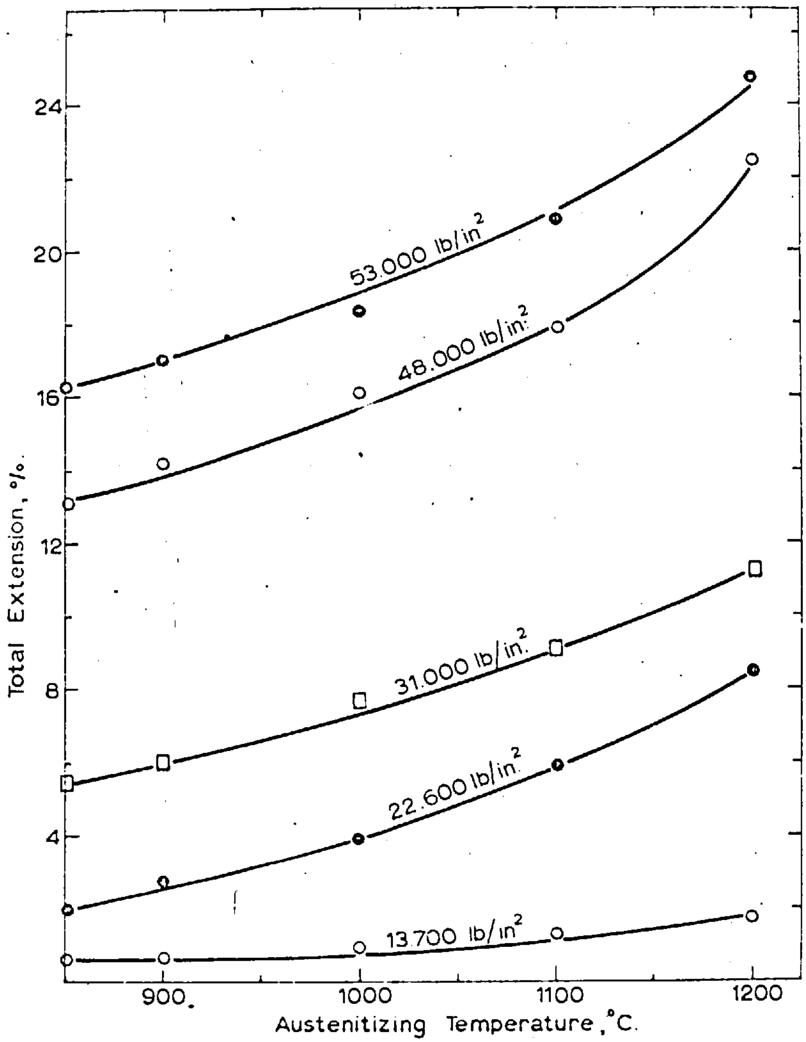

Fig.27. Effect of austenitising temperature on total extension for En30A steel; stressed at 330°C

was observed. If after an austenitizing treatment at 1200°C, the specimen was stressed 7°C above M_s with 50,000 lb/in^2, the amount of recalescence at the temperature of stressing and at M_s were about 4°C and 30°C respectively, whereas, when the same specimen was stressed at M_s with the same stress, the rise at M_s amounted to 50°C. If, however, the stress (50,000 lb/in^2) was applied 1°-3°C below M_s, only a slight rise in temperature was observed which never exceeded 3°C.

The results of experiments on the En 30A steel and on the iron-20 wt% nickel alloy, in which the stress was applied at the respective M_s temperatures, are shown in Figs. 28 and 29 respectively. The results showed that for both of the alloys investigated the amount of transformation plasticity increases with an increase in austenitizing temperature. But in the iron-20 wt% nickel alloy the increase in transformation plasticity with increasing austenitizing temperature ceased at 700°C. Austenitizing the specimens between 700°C and 1000°C did not alter the amount of plasticity. However, austenitizing the specimen above 1000°C increased the plasticity by a small amount. The effect of stress was particularly marked in the case of austenitizing between the temperature range of 850° to 1000°C in the En 30A steel and in the temperature range of 600° to 700°C in the iron-20 wt% nickel alloy. Thus in the experiments involving

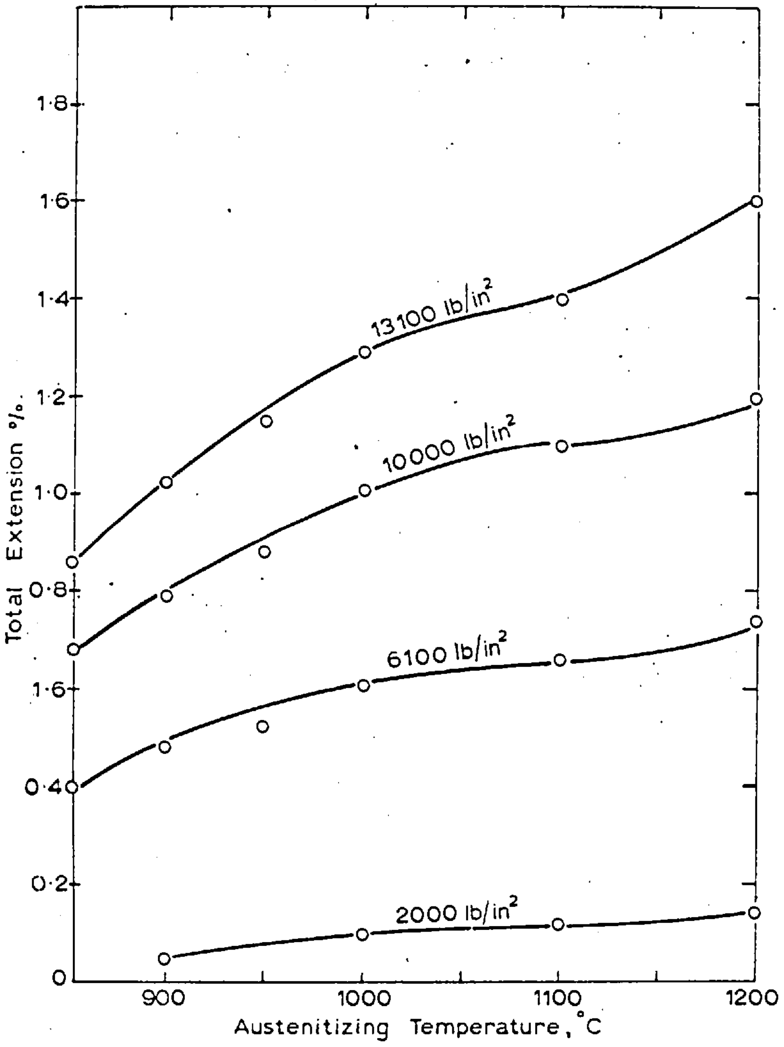

Fig 28. Effect of austenitising temperature on total extension for En30A steel; stressed at M_s.

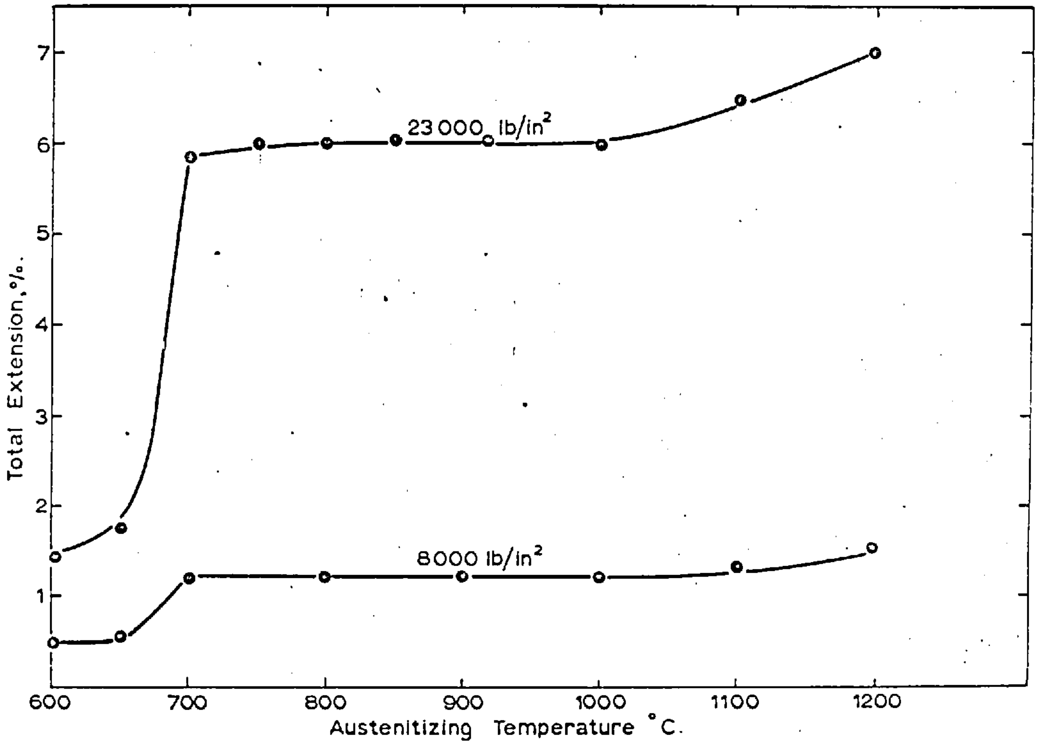

Fig. 29. Effect of austenitising temperature on total extension for Fe-20wt°/o Ni alloy austenitised 3min 850°C stressed at M_s in each case.

iron-20 wt% nickel alloy, for a given stress of 23,000 lb/in^2 it was found that the amount of extension during transformation increased from 1.45% to 5.8% with an increase in austenitizing temperature from 600° to 700°C. For the same stress, increasing the austenitizing temperature from 1000°C to 1200°C resulted in an increase in extension from approximately 6% to 7% only.

The various austenitizing temperatures did not appear to have any marked effect on the temperature where transformation plasticity is completed, as measured by the transducer readings. In the early stages of transformation the amount of martensite formed, for a given degree of undercooling below the M_s temperature, was larger for the lower austenitizing temperature. However, at any fixed temperature which is below the M_s temperatures obtainable for any austenitizing temperature, the trend was reversed; the higher the austenitizing temperature the higher was the amount of martensite formed. The later stages of transformation were sluggish and at a temperature sufficiently below the M_s temperatures obtainable for any austenitizing treatment, the amount of martensite for all the austenitizing conditions was found to be almost the same. In the case of the En 30A steel samples, the transformation plasticity was completed at a temperature of approximately 270°∓20°C, during which about 70°∓10% transformation to martensite occurred for all the austenitizing treatments investigated.

ii) <u>Effect of short time austenitizing treatments</u>:

The effects of very short time austenitizing treatments on transformation plasticity have been studied on the En 30A steel. A short-time austenitizing treatment involved a very rapid heating of the specimen (at approximately 200°C/sec.) to a preselected austenitizing temperature for very short times, e.g. one or two seconds, followed by cooling in vacuum and stressing at a predetermined temperature. The total time a specimen is in the austenitic condition depends on the rate of heating up to the austenitizing temperature, in addition to the actual times of holding at the maximum temperature of austenitizing. The results of short time and normal austenitizing treatments at 850°C for various stresses applied at the M_s temperature are shown in Fig.30. The total extension percentage is seen to be much less for the short time treatments, the difference between short-time and normal treatments being particularly marked at the higher stress levels. For example, a stress of 15,000 lb/in^2 when applied at M_s to a specimen heated at a rate of 200°C/second from room temperature to 850°C and held there for one second gave a total extension of 0.6% as compared with the normal value of 1.1% for a standard austenitizing treatment at that temperature. However, for a stress of 30,000 lb/in^2, the extension

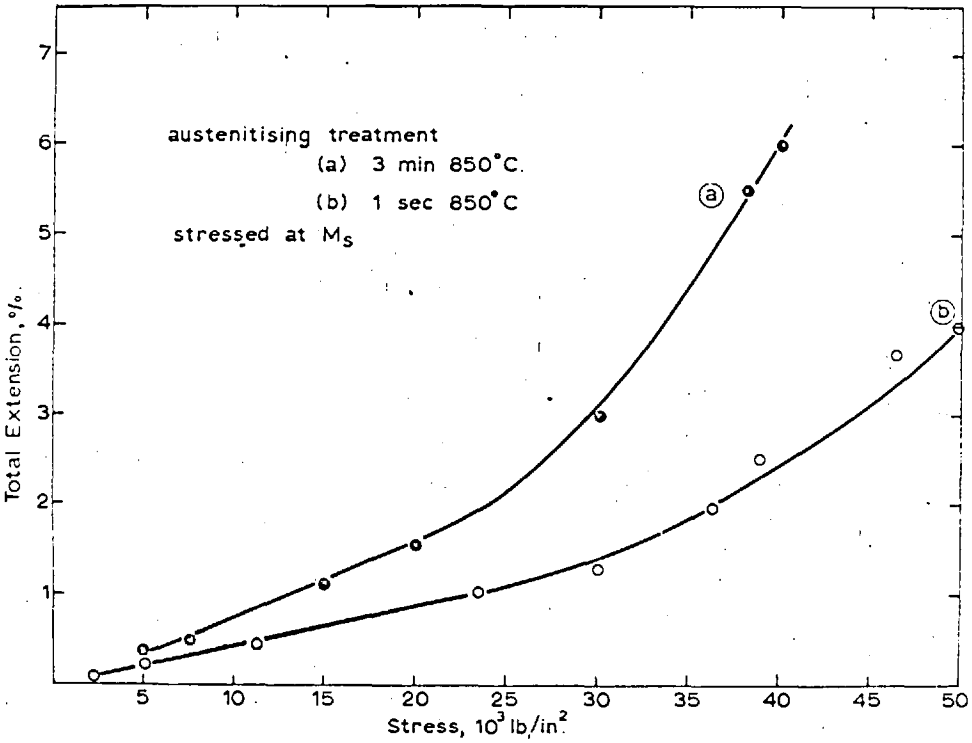

Fig.30. Effect of short-time austenitising on total extension for En 30A steel.

values (transformation plasticity) for 'short-time' and 'normal' austenitizing treatments at 850°C were approximately 1.25% and 3% respectively.

The results of Fig.30 show that, after a short-time austenitizing treatment, there is also a straight line relationship between the total extension percentage and the magnitude of the applied stress in the lower range of stresses. The slope of the straight line portion of the curve for the short-time treatment was less than for the normal austenitizing treatment. The plotted data for the short-time austenitizing treatment showed also a departure from linearity beyond a certain stress level and the departure from linearity occurred at about 35,000 lb/in^2.

The results of Fig.31 show that the difference between the two types of treatment decreases as the austenitizing temperature increases. Thus, for a given stress of 30,000 lb/in^2 and stressing temperature of 330°C, the extension values for short-time and normal austenitizing treatments at 850°C were 1.25% and 5.5% respectively, as the austenitizing temperature was increased to 1200°C the extension values were approximately 9% and 10.5% respectively, the difference between the extension values being reduced considerably.

iii) Effect of austenite grain size and defect structure:

Experiments with different austenitizing

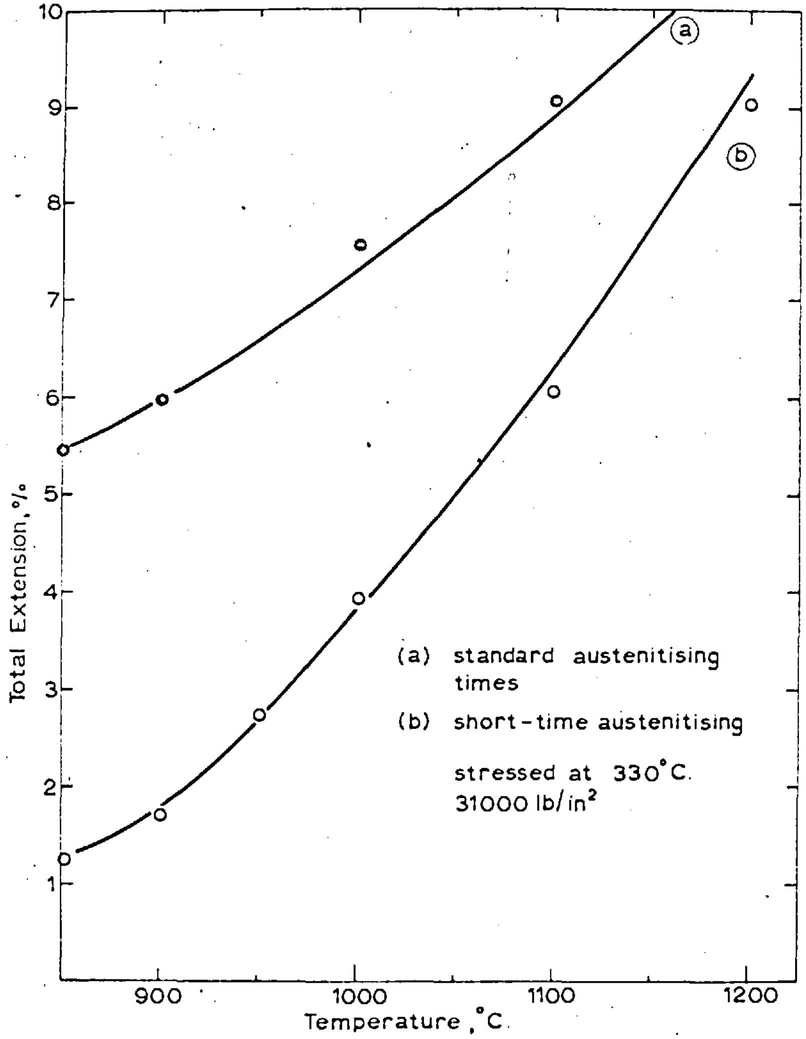

Fig. **31**. Effect of austenitising temperature on total extension for En30A steel.

conditions were also made to study the influence of various important factors contributing to the transformation plasticity as the result of different austenitizing treatments. Some double austenitizing treatments were carried out on the En 30A steel in an attempt to separate the grain size effect from some other submicroscopic effect which varies also with the austenitizing temperature. Treatments were made involving initial austenitizing at a high temperature for the specified standard time, to establish a grain size, characteristic of that temperature, and bringing down the temperature to a pre-selected holding temperature and for a predetermined time (still within the γ region) to allow any submicroscopic change to occur in the austenite and finally cooling and stressing. The double austenitizing treatment experiments have shown that once a homogeneous austenite is formed with a certain grain size, holding the specimen at a lower temperature did not alter the transformation plasticity; i.e. the values obtained were characteristic of the higher temperature treatments. For example, holding the specimen at 700°C for two hours after initial austenitizing at 1200°C for 100 seconds yielded the same percentage extension as for a single austenitizing treatment for 100 seconds at 1200°C.

Experiments were also made on the En 30A steel by varying the heating rate through the A_{c1}-A_{c3} range during austenitizing to separate the effects due to the austenite grain size and the austenite defect structure on transformation plasticity. During the formation of austenite, defects are introduced in the structure because of the volume change involved. Thus it is possible that the faster the heating rate at which the austenite is formed, the greater might be the defects introduced in the structure. It was expected, therefore, that, by employing very slow and very quick heating rates through the A_{c1}-A_{c3} range it might be possible to study the effect of various defect concentrations on transformation plasticity.

Specimens were heated from the martensitic structure at a standard rate of 100°C/second to a temperature of 690°C (5°C below A_{c1}) and then heated from A_{c1} to the A_{c3} temperature (730°C) very slowly in 1 hour. Above A_{c3} the specimens were quickly heated to the austenitizing temperature (800°C) and after holding there for three minutes, specimens were cooled and stressed at M_s to study the transformation plasticity. It has been shown(84) that for the range of conditions used in the present work, the austenite grain size is dependent only on the time at the highest austenitizing temperature and independent of the heating rate through A_{c1}-A_{c3}. The results of the experiments on transformation plasticity are given together with those for normal austenitizing treatment involving rapid heating at 100°C/sec. to the austenitizing temperature, in Fig.32. The extent of transformation plasticity was found to be greater for the slow heating than for the rapid heating; e.g. for a stress of 30,000 lb/in^2 the strain was approximately 3.2% as compared with 2.2% for rapid heating austenitizing.

iv) <u>The necessity of phase transformation for the occurrence of transformation plasticity</u>:

When a stress was applied to a specimen with a stable structure, such as tempered martensite, or pearlite and ferrite structure, no plastic deformation was observed during cooling. This was confirmed in one of the experiments in which fully martensitic specimens were tempered at 500°C and cooled to room temperature. In the other type of experiments some specimens were first austenitized and then cooled to 600°C where they were kept for one day. During this isothermal treatment all the austenite transformed to ferrite and pearlite structure. The specimens were then heated to 500°C and stressed at about 330°C during cooling. Tensile stresses ranging up to 40,000 lb/in^2 for tempered samples and 50,000 lb/in^2 for ferrite and pearlite structures were used. Since no plastic deformation was observed for these experiments, the effects described hitherto are attributed to transformation plasticity.

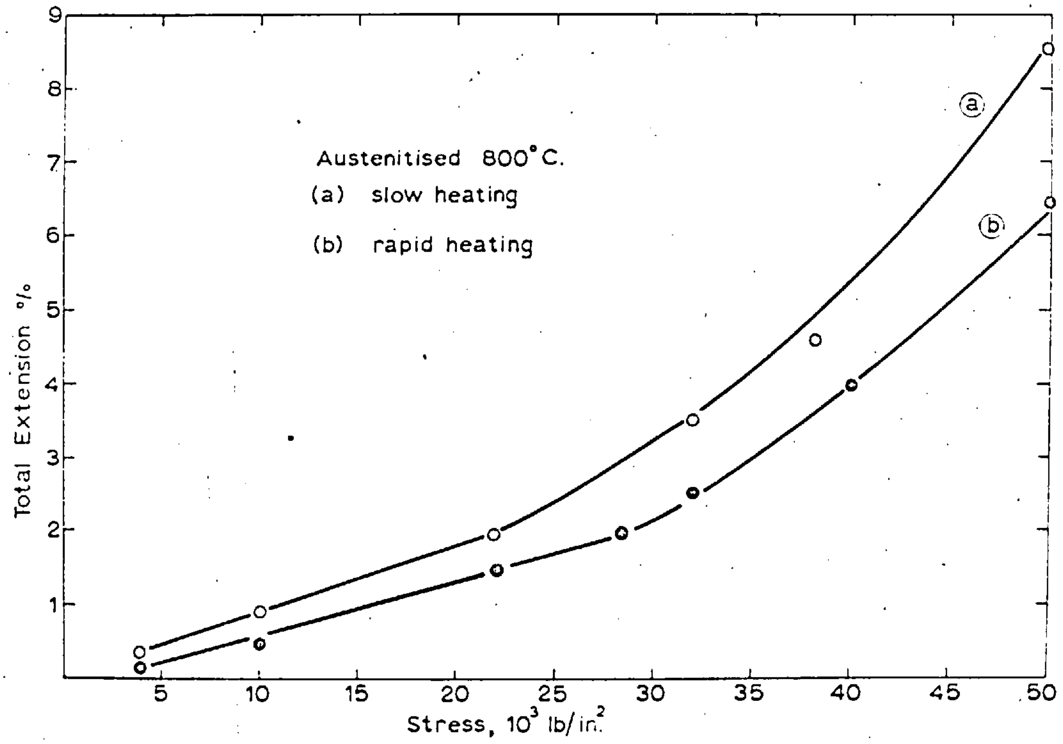

Fig. 32. Effect of heating rate through critical range on total extension for En 30A steel.

v) Austenite yield stress:

In the measurements of the strength of the austenite, the yield stress was taken as the stress corresponding to appr. 0.1% ~~0.2%~~ strain and the term yield stress, used throughout the present work, has this meaning. Experiments were made on the En 30A steel and the iron-20 wt% nickel alloy to find the yield stress of austenite in the approximate temperature ranges 330° - 650°C and 300°-600°C respectively. The results of the measurements of the austenite yield stress of the En 30A steel, following various austenitizing treatments, are shown in Fig.33. Large differences in the strength of the austenite were found for the various austenitizing treatments; the highest strength values were obtained with the austenite produced by austenitizing for one second at 850°C, and the strength progressively decreased by increasing austenitizing time at 850°C or by increasing the austenitizing temperature. The magnitude of the difference between the two extremes of treatment used may be illustrated by considering the yield stress at, say, 330°C. At this temperature, the yield stress of austenite produced by a one second treatment at 850°C is nearly three and a half times greater than that of austenite produced by a 1200°C treatment, the values being approximately 37,000 lb/in^2 for the one second treatment at 850°C and 11,000 lb/in^2 for the standard

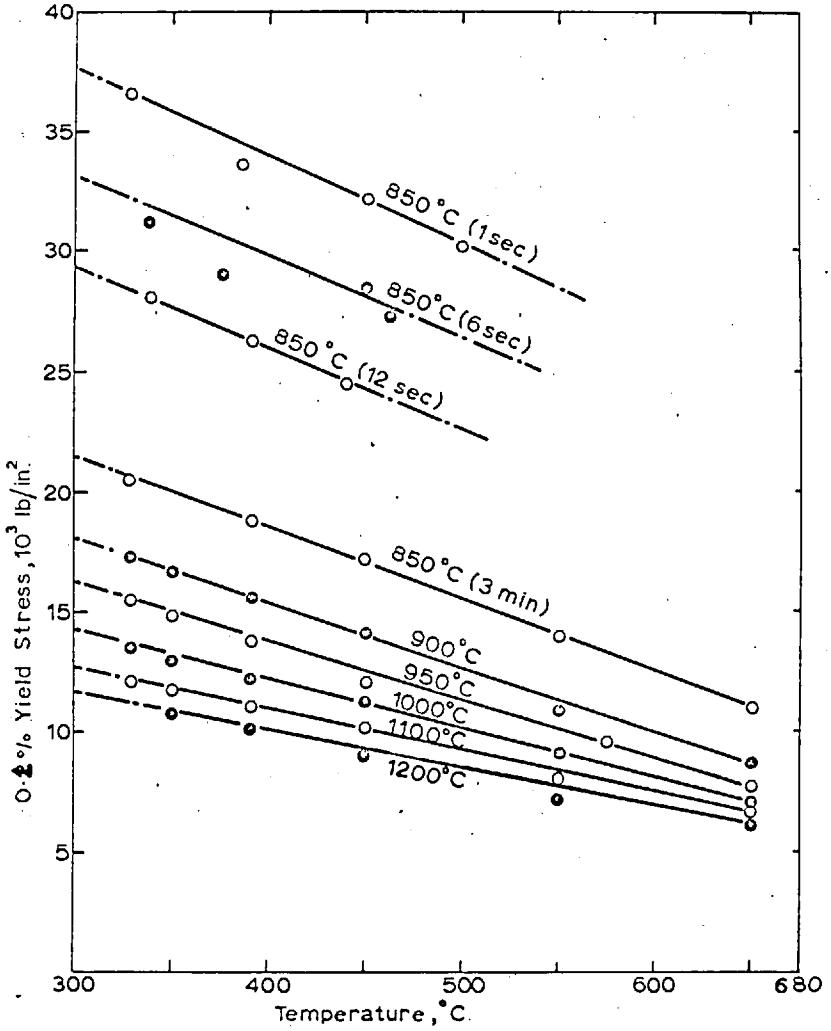

Fig.33. 0·2 °/o yield stress of austenite as a function of temperature for various austenitising treatments for En 30A steel.

treatment at 1200°C.

The results of measurements of the austenite yield stress for the iron-20 wt% nickel alloy are shown in Fig.34, for the various austenitizing temperatures studied. As in the case of the En 30A steel the yield stress for the lowest austenitizing temperature used is considerably higher than that obtained with the highest austenitizing temperature; e.g. for an austenitizing treatment at 600°C the yield stress of austenite at 300°C was approximately 29,000 lb/in^2. With an increase in the austenitizing temperature to 1200°C, the yield stress decreased to approximately 8000 lb/in^2. The fact that little change in yield stress was brought about by changing the austenitizing temperature between 700°C and 1000°C was of special interest; differences in yield stress in this temperature range appear to be within the experimental scatter.

The slopes of the various curves of yield stress vs. temperature show that the temperature dependence of yield stress showed some differences for the various austenitizing temperatures: the reduction of yield stress with increasing temperature was the greatest in the iron-20 wt% nickel alloy for the lowest austenitizing temperature and, in the case of the En 30A steel, for the short time austenitizing treatment.

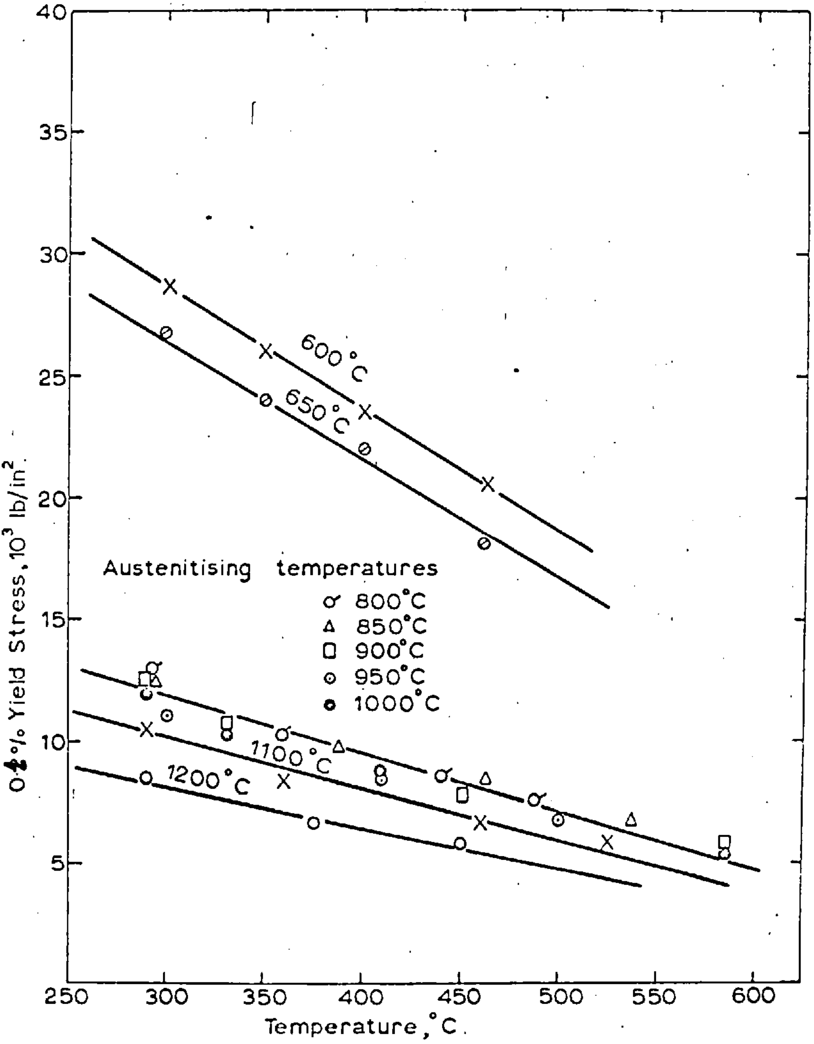

Fig.34. 0·2% yield stress of austenite as a function of temperature for various austenitising treatments for Fe-20wt% Ni alloy.

It is interesting to note that in both the En 30A steel and in the iron-20 wt% nickel alloy investigated, the plot of the yield stress of austenite after different austenitizing treatments vs. the M_s temperatures is a straight line, suggesting a direct correlation between the strength of austenite and M_s temperatures, as shown in Figs.35 and 36 respectively.

IV. Experiments to study the structural aspects of transformation plasticity:

i) Metallographic examination:

Some metallographic work on the En 30A steel using the Greninger-Troiano quench temper technique was done to study the martensite formed under various conditions of stressing.

In one series of experiments with a standard 850°C austenitizing treatment, a selected stress was applied at M_s, and the specimen was allowed to cool 5°C below M_s; the stress was removed and the specimen was rapidly heated to 500°C and held at this temperature for one minute to temper the initially formed martensite followed by air quenching to room temperature. This tempering temperature is in the bay of the T.T.T diagram and was used to minimize the tendency for bainite formation. It is known that the incubation period of the bainite transformation is reduced in the presence of martensite, and

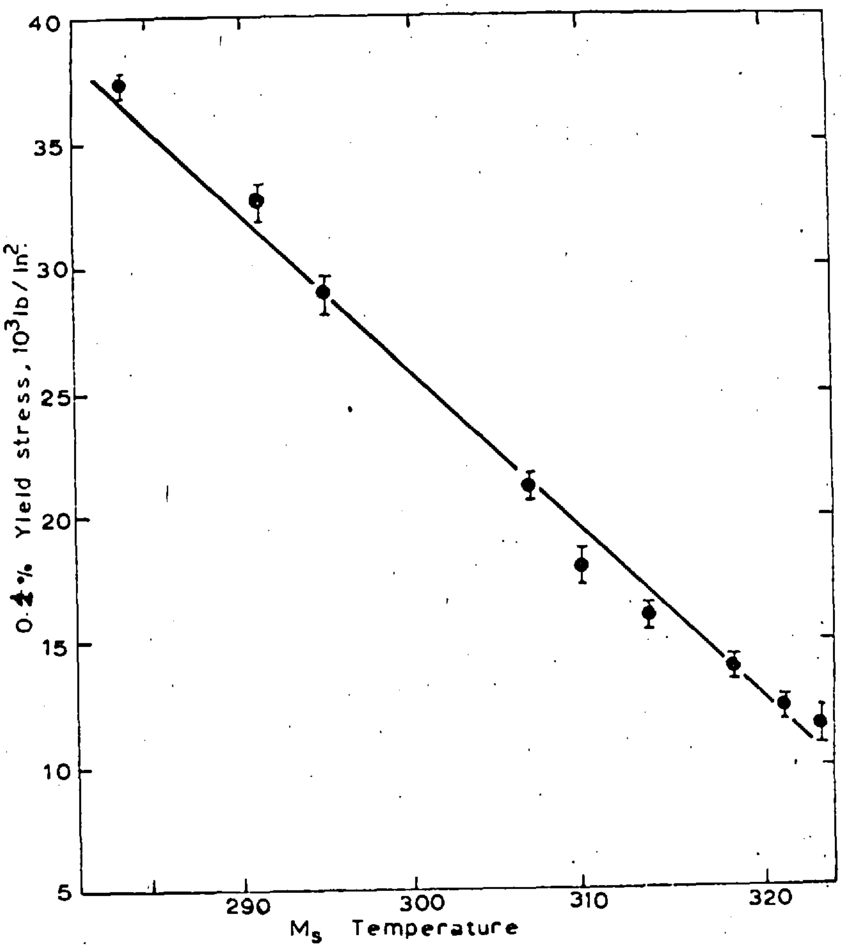

Fig. 35. Variation of M_s with yield strength of austenite in En 30 A steel.

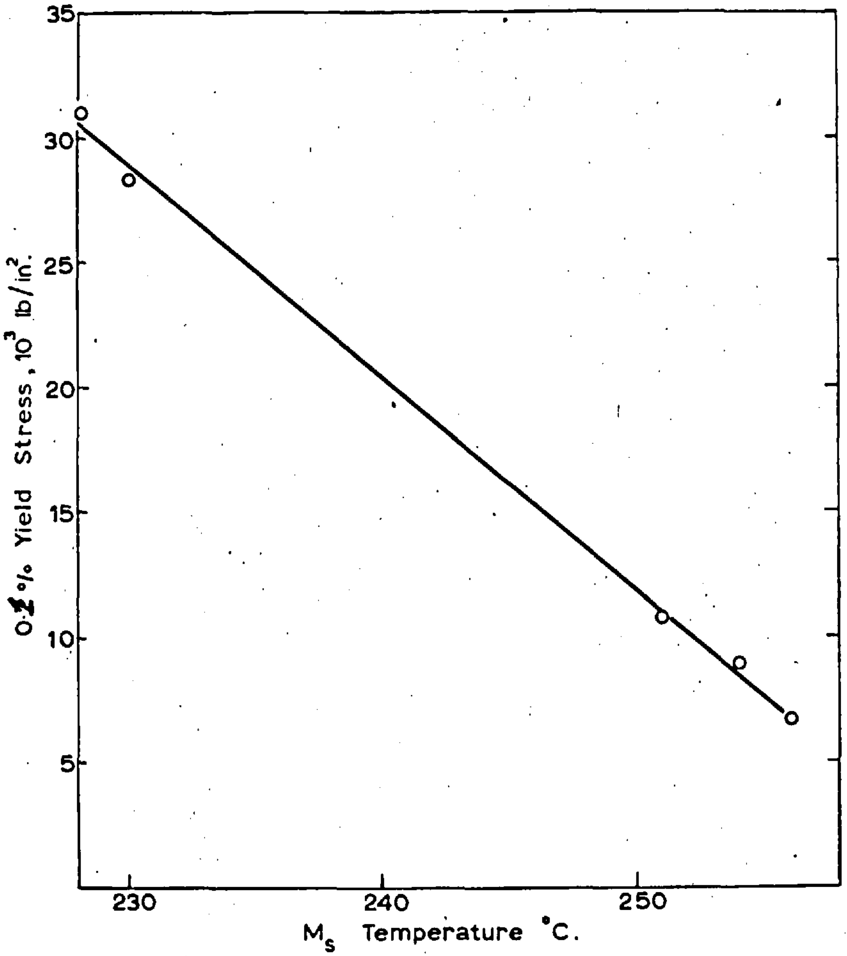

Fig. 36. Variation of M_s with yield strength of austenite in Fe-20% Ni alloy

the tempering time was therefore also kept to a minimum. Five stress levels were studied, namely, 8,000; 23,000; 25,000; 35,000; 53,000 lb/in^2 respectively. The martensite formed with a stress of 8,000 lb/in^2 was indistinguishable from that which formed in experiments where no stress was applied and the martensite plates showed no preferred orientation. Fig.37 shows the martensite plates formed on cooling 5°C below M_s under a stress of 8,000 lb/in^2. However, with the stress of 23,000 lb/in^2, the martensite plates were found to have a preferred orientation, the plates being inclined at about 40° to the tension axis. Figs. 38, 39 and 40 are the optical photomicrographs of an appropriately treated specimen showing the preferential occurrence of martensite plates under the above mentioned tensile stress. Similarly, Figs. 41, 42, 43 and 44 are showing the preferential occurrence of the martensite plates under a tensile stress of 25,000, 35,000 and 53,000 lb/in^2 respectively. With an increase in applied stress, a larger number of martensite plates showed the preferred orientation effect. As can be seen in Figs. 38, 40, 42 and 43, there is an indication that the fineness of the martensite plates increases also with increase in applied tensile stress.

The martensite plates that are induced by the applied stress during deformation between M_d and M_s have

Fig.37: Martensite plates formed 5°C below M_s under a stress of 8,000 lb/in^2. The direction of stressing is shown by white arrows. x 1200.

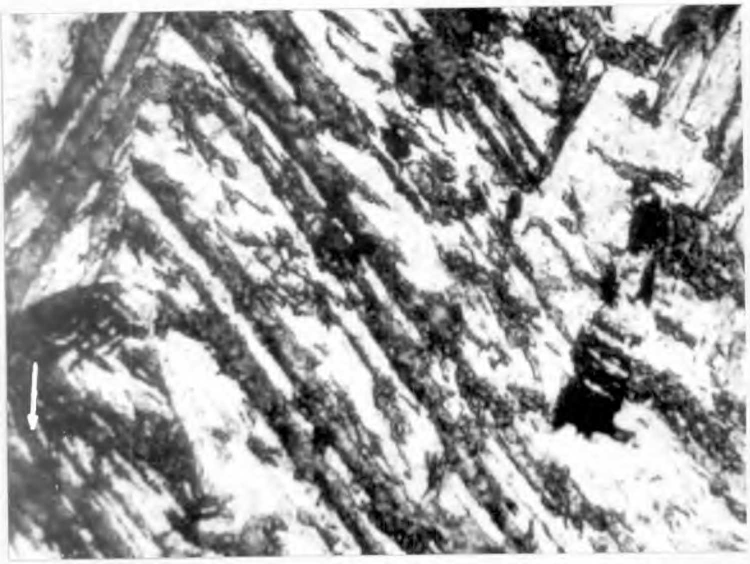

Fig. 38: Martensite plates formed 5°C below the M_s under a stress of 23,000 lb/in^2. x 1200.

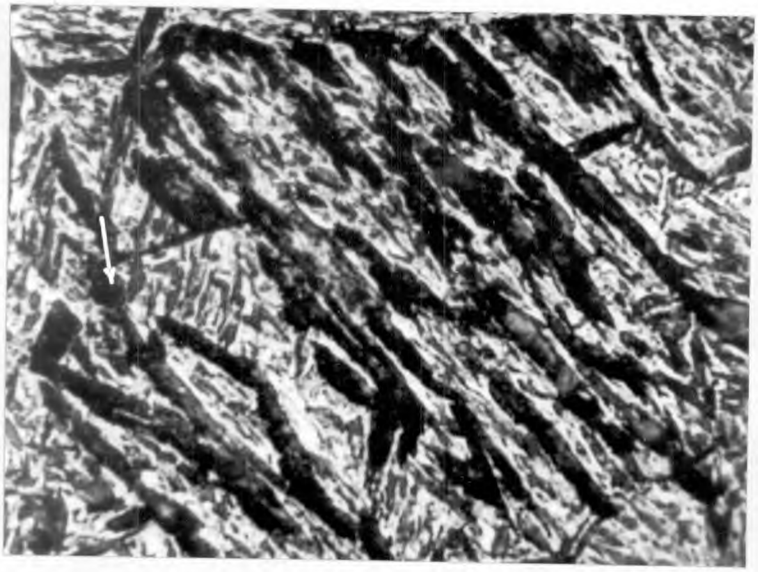

Fig.39: Martensite plates formed 5°C below M_s under a stress of 23,000 lb/in^2. x 1200

Fig.40: Martensite plates formed 5°C below M_s, under a stress of 23,000 lb/in^2. x 1200

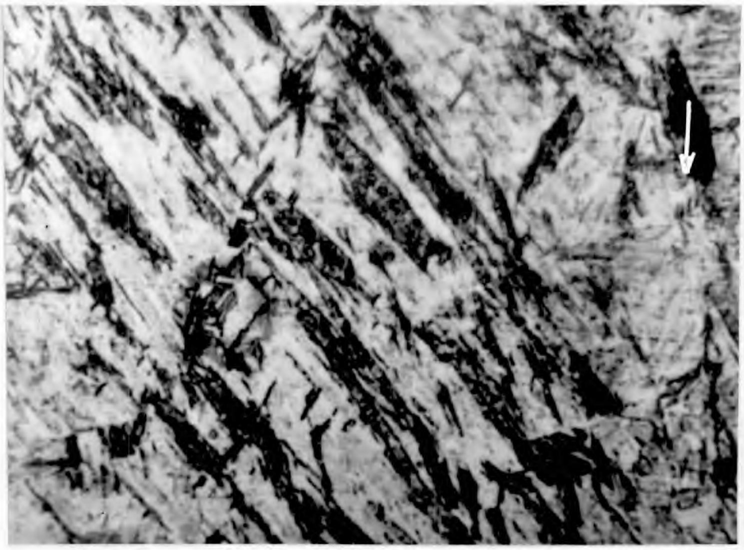

Fig.41: Martensite plates formed 5°C below M_s under a stress of 25,000 lb/in^2. x 1200

Fig.42: Martensite plates formed 5°C below M_s under a stress of 35,000 lb/in^2. x 1200

Fig.43: Martensite plates formed 5°C below M_s under a stress of 53,000 lb/in^2. x 1200

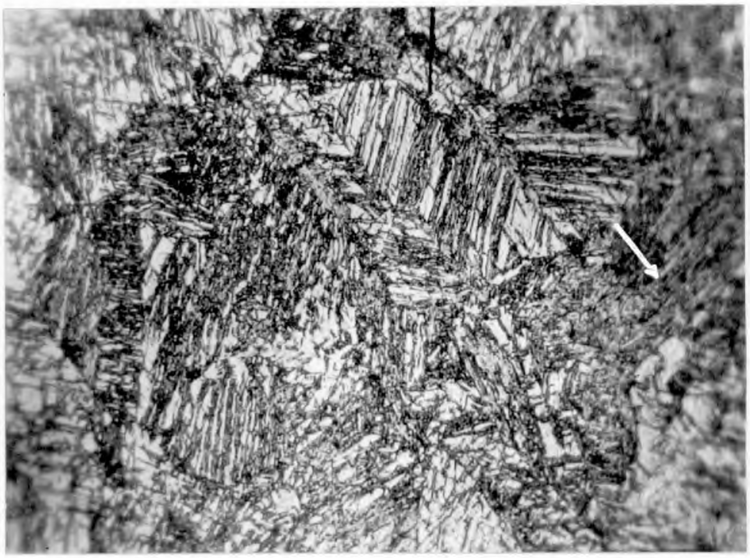

Fig.44: Martensite plates formed 5°C below M_s under a stress of 53,000 lb/in^2. x 700

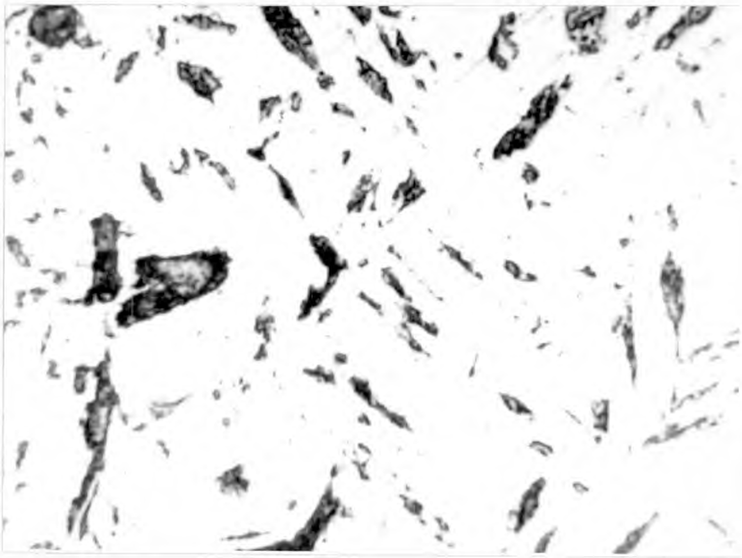

Fig.45: Martensite plates induced by a stress of 31,000 lb/in^2 applied approximately 40°C above M_s. x 1200

been found to be smaller than those that form just below the normal M_s under the same stress. Fig.45 shows the martensite formed by applying a tensile stress of 31,000 lb/in^2, approximately 40°C above M_s, and then tempering the stress induced martensite for one minute at 500°C and quenching. The size of these plates may be compared with those formed below M_s, in the photomicrograph of Figs.41 and 42.

Experiments were also made in which a stress was applied 10°C below M_s and maintained during cooling through a further 5°C; in this case the martensite formed while the stress was applied, contained within a structure in which some martensite plates were already present. Microscopical examination did not reveal any significant preferred orientation effect in these experiments.

It is also significant to note that the application of stress affects the transformation kinetics, viz., the rate of martensite formation. It is found that, as a stress was applied at or above or below M_s temperature, it increased the kinetics of transformation depending on the magnitude of applied stress and temperature of stress application. As assessed by the results of point counts, the larger the applied stress, (the ~~faster~~ greater is the ~~rate~~ amount of transformation to martensite, and hence) the greater the amount of martensite formed for any particular temperature below M_s. However, in the later stages of transformation, the amount of martensite present was similar to that which transformed freely without the application of any stress. This temperature corresponds to the temperature where approximately 70∓10% martensite is formed and was about 270∓20°C for the En 30A steel.

The amount of martensite formed at 305°C (2°C below M_s) after a standard austenitizing treatment at 850°C is shown in Table 2, for the different stress levels employed. As can be seen, the greatest effect was observed with the application of large tensile stresses. A point which is worth mentioning is that during the experiments, the stress was always removed at 305°C. With the application of large stresses, recalescence occurred, and this increased the temperature range of martensite transformation. For comparison purposes, the amount of martensite formed without any application of stress at different temperature levels, is given in Table 3.

TABLE 2.

The results of point counts* on the En 30A steel.

Austenitizing treatment: 3 minutes at 850°C.
Stress applied at: M_s temperature (307°C).
Stress removed at: 305°C.

Stress level used (lb/in^2)	Martensite percentage
10,000	10
20,000	18
50,000	41

---

*An average of about 600 counts on each experiment has been made.

TABLE 3.

The results of point counts on the En 30A steel.

Austenitizing treatment: 3 minutes at 850°C.
M_s temperature: 307°C.

Martensite formed at (°C)	Martensite percentage
305	7%
300	22%
285	53%
270	70%

The application of a large stress at or above the M_s temperature after a high temperature austenitizing treatment, affected the transformation kinetics considerably. Adiabatic heating occurred and thus the temperature range in which plasticity occurred is altered. When a stress of 50,000 lb/in^2 was applied at the M_s temperature after a standard austenitizing treatment at 1,200°C, a sudden extension occurred during which the temperature virtually stayed constant. However, the further increase in length was accompanied by a rise in temperature. The rise was quite appreciable and amounted to about 50°C. More than half of the total amount of extension occurred during this adiabatic heating. After this increase, the transformation plasticity almost ceased. Metallographic observations revealed that during this recalescence, about 60∓10% martensite was formed.

ii) <u>X-ray diffraction examination</u>:

Some X-ray work has been done to study the martensite formed under various stressing conditions. Debye Scherrer rings were obtained of the wire specimens using MoK_α monochromatic radiation.

Since the positions of the 'intensity maxima' on the Debye Scherrer circle depend on the angle between the reflecting plane and the rotation axis, which is the axis of the specimen, the orientation of the martensite plates relative to the rotation axis was determined by measuring these positions (see appendix 1). These experiments confirmed the existence of a preferred orientation effect. It was found that the fibre axis of the martensite formed under a stress of 50,000 lb/in^2 was [320]. There was also some scatter of the [320] direction about the wire axis, since the reflections on the film were long arcs rather than sharp spots.

However, the orientation of the martensite is more complex than might appear to be the case at first sight; thus it could not be fully established. This is apparent when photographs such as Fig.46 are studied carefully. From Fig.46 it is obvious that the orientation, even after transformation under very high stresses, was not perfectly developed and there was sufficient scatter about the ideal peak positions on the diffraction rings to observe minor differences which might be critical.

Unless there is a high degree of perfection in the orientation of martensite, it is not possible to specify the fibre axis very accurately. Thus the [320] direction which is found to be the fibre axis is not claimed to be very accurate.

iii) Mechanical Properties of Martensite after Transformation Plasticity:

The mechanical properties of martensite formed

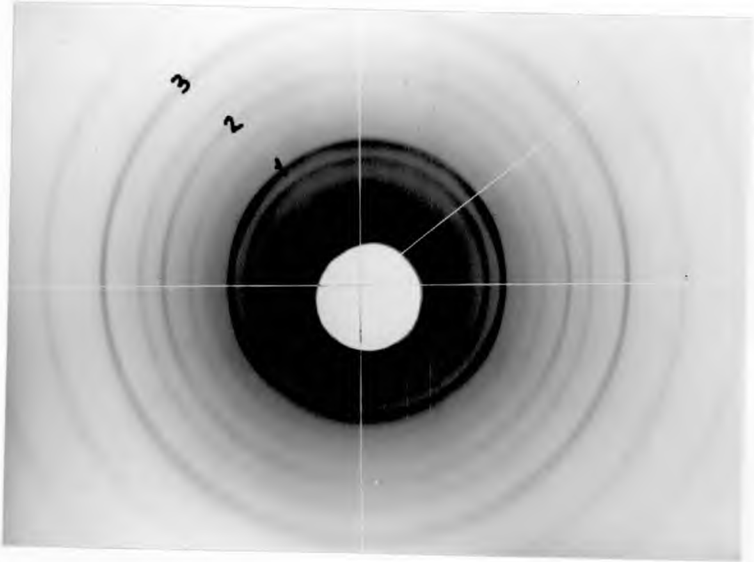

Fig. 46: En 30A steel specimen transformed to martensite under a stress of 50,000 lb/in^2 MoK$_\alpha$ radiation.

under varying stressing and austenitizing conditions are of potential interest; thus microhardness and tensile tests were carried out on some (En 30A steel) martensitic samples. The values of ultimate tensile strength of martensite, formed after different austenitizing treatments with or without application of stresses are given in Table 4.

TABLE 4.

Results of tensile test on the En 30A steel.

Treatment	U.T.S. values for martensite formed without the application of stress	U.T.S. values for martensite formed under a stress of 25,000 lb/in^2	U.T.S. values for martensite formed under a stress of 50,000 lb/in^2
	lb/in^2	lb/in^2	lb/in^2
1. Standard austenitizing of 3 minutes at 850°C.	236,000	236,000	237,000
2. Standard austenitizing of 3 minutes at 1000°C.	230,000	231,000	231,000
3. Standard austenitizing of 100 seconds at 1200°C.	227,000	227,000	228,000

All tests were carried out with a Hounsfield tensometer. The figures quoted in Table 4 are the average of two tests made on two samples treated identically. The limitations of accuracy come from the non-uniform deflection of the tensometer beam, from the bending moment applied to the specimen by the effect of the weights of the grips (as the specimen is stressed horizontally) and from the difficulty of perfect alignment of the wire specimen between the disc chucks.

The results of microhardness tests on the martensite formed after different austenitizing treatments, with or without the application of stress, are given in Table 5.

TABLE 5.

Results of microhardness tests on the En 30A steel.

Microhardness values converted to VPN.

Treatment	Microhardness value. (Martensite formed without the application of stress	Microhardness values. Martensite formed under a stress of 25,000 lb/in^2)	Microhardness values. (Martensite formed under a stress of 50,000 lb/in^2)
1. Standard austenitizing of 3 minutes at 850°C.	570	560	573
2. Standard austenitizing of 3 minutes at 1000°C.	560	565	565
3. Standard austenitizing of 100 seconds at 1200°C.	540	540	550
4. Short time treatment at 850°C for 1 second.	610	620	626

All tests were carried out on a Reichert Microhardness tester at a load of 100 grammes.

It is apparent from tables 4 and 5 that there is no marked effect of stressing the specimen during transformation on the mechanical properties of the final structure. However, it seems that there is a slight tendency for the strength of martensite to increase with decreasing austenitizing temperature and time.

V. Experiments to study the effects of stress on transformation plasticity during the tempering of martensite:

To study the plastic deformation during tempering, the En 30A steel wire specimen is brought to the martensitic condition and stressed at room temperature, after which it is heated at 100°C/minute to the required tempering temperature (300°C or 400°C), held at this temperature for 5 minutes and then cooled back to room temperature. If during this process no load was applied, the length of the specimen virtually did not change.

The threshold stress, for the occurrence of detectable plastic elongation during tempering at 300°C was found to be about 10,000 lb/in^2. Although no systematic work has been done on the effect of tempering temperature on the threshold stress, the general trend was a decrease in threshold stress with increasing tempering temperature. Increasing the stress above the threshold value or increasing the tempering temperature, increased the amount of plastic extension. The 0.08% plastic deformation which resulted from the application of a 10,000 lb/in^2 stress during tempering at 300°C, increased to 0.3% with a stress of 42,500 lb/in^2. Further, this 0.3% plastic deformation increased to about 0.5% with the increase in tempering temperature to 400°C under the stress of 42,500 lb/in^2. The results of Fig.47 show this behaviour; it also reveals that for a given tempering temperature there is almost a straight line relationship between the total extension and the magnitude of the applied stress. The slope of the straight line increases with increase in tempering temperature.

Some Vicker hardness tests were performed on plain tempered and stress tempered specimens and it was found that the hardness slightly increases with stress. For example, the hardness of a specimen plain tempered (at 400°C) was 360 HV. and this increased to 420 HV. with stress-tempered (42,500 lb/in^2) specimens.*

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*It seems that the effect of stress, for the tempering time employed, is to increase the hardness slightly. The very little amount of work which is done in this field is not enough to reach a conclusion. There is still much to be learnt about the influence of stress, tempering temperature and time on the mechanical properties of the material.

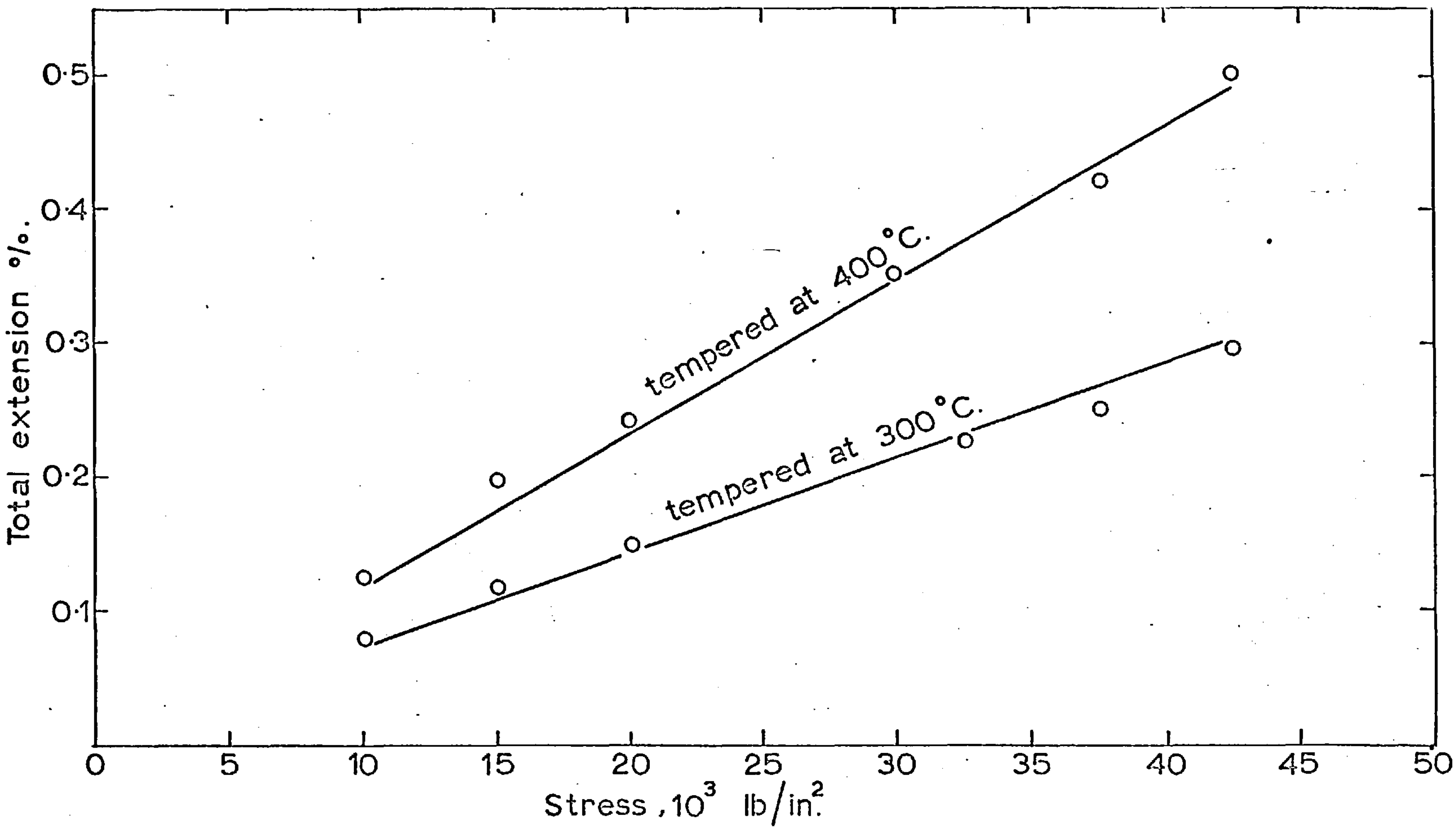

Fig. 47. The effect of the magnitude of applied stress and tempering temperature on the total extension, for En 30A steel.

Some fully martensitic specimens were first tempered at 500°C and cooled to room temperature. They were then reheated under a stress to 400°C. No plastic deformation was observed for these experiments described above. On the basis of this evidence, it appears that the length changes that occurred during tempering under stress are due to a transformation plasticity effect.

CHAPTER 5.

DISCUSSION

5. DISCUSSION:

I. General:

The principle findings of the present investigation are as follows:

i) For a given temperature of stress application and up to a certain stress level corresponding to the yield stress of the weaker phase, the extent of transformation plasticity increases linearly with increase in applied tensile stress.

ii) The extent of transformation plasticity decreases as the temperature of stress application is decreased below M_s and no effect of stress on transformation plasticity is observed for stressing temperatures lower than about 40°C below transformation temperature.

iii) The extent of transformation plasticity varies with the austenitizing conditions employed; lower transformation plasticity values are obtained with short time austenitizing treatments at relatively low temperatures, and the higher plasticity values with higher austenitizing temperatures.

iv) Stabilizing the parent phase by mechanical or thermal means decreases the total extension due to the transformation plasticity.

The results of the present investigation suggest that the magnitude of transformation plasticity associated with martensitic (diffusionless) transformation is a linear function of the applied stress (up to a stress corresponding to austenite yield stress), an inverse function of the austenite yield stress and a linear function of the change in specific volume accompanying phase transformation. This is in accord with the predictions of the macroplasticity approach to the phenomenon of transformation plasticity developed by de Jong(29,33), and Johnson and Greenwood(59,60) and which hitherto does not appear to have been applied specifically to martensite formation. Johnson and Greenwood(59,60) have proposed a quantitative relationship based on the fact that when a product phase forms in a continuous parent phase, the associated volume change produces a stressed region at the phase interface, and under this condition the material yields to the superimposed external stress at a lower value, because of the pre-existing internal stresses. Excess elongations are to be expected at even very small external stresses if these stresses cause slip in the potential slip systems which are activated by the internal stresses set up due to the change in volume during transformation. Thus the magnitude of the external stress which will just start the plastic yielding, i.e. the threshold stress for increased extension, will be dependent on the internal stresses produced by the change of volume. If the volume change is great enough to cause the stressed region to reach the yield stress of the weaker phase, excess elongations are expected to occur even with the smallest applied stress.

The quantitative interpretation of the transformation plasticity giving the increment of total permanent deformation in the direction of a small applied stress, as derived by Johnson and Greenwood (see appendix II), is as follows:

$$\varepsilon = A \frac{\sigma}{Y} \frac{\Delta V}{V} + B \frac{\Delta V}{V} \quad \ldots\ldots\ldots \text{(i)}$$

where, ε = the permanent deformation occurring during the transformation from one phase to another.

σ = the applied stress

Y = the yield stress of the weaker phase at the transformation temperature (i.e. austenite in present case).

$\frac{\Delta V}{V}$ = the fractional volume change associated with 100% transformation.

A = a constant, having the value of 2/3.

B = a constant, whose value is dependent on the geometry of the transformation front.

The term '$B \frac{\Delta V}{V}$' represents the plastic deformation due to the transformation alone, without the application of an external stress. The constant 'B' is sufficiently small so that for the present purpose the second term can be neglected. In fact, experimental results have revealed that the plastic deformation due to the transformation alone was always less than 0.06%.

The linear relationship of the above equation is expected only to apply for small stress; for stresses greater than the yield stress of the weaker phase there will be a departure from linearity in the direction of greater amounts of plasticity.

The effects of different austenitizing and stressing conditions on transformation plasticity during martensite formation have been analysed quantitatively by the use of the above relation as follows. Several preliminary points should be noted.

First, it must be shown that the associated volume change during transformation produces a stressed region in the austenite (which is the weaker phase in the present investigation) and that this stress reaches the yield stress of austenite. The internal stresses set up due to the volume change can be roughly estimated in the following way: during the calculations it will be assumed that the transforming crystal is completely embedded in bulk austenite and also the martensite is considered to be isotropic, incompressible and perfectly plastic; stress fields produced by neighbouring transforming centres will be neglected. If the crystal is approximated by a sphere of radius r_o during transformation the volume of the sphere r_o would increase by ΔV. This increase in volume causes a radial displacement, Δr, of the austenite outside the sphere.

The original volume of the sphere r_o is,

$$V = 4/3 \, \pi r_o^3$$

differentiating both sides we get

$$dV = 3 \tfrac{4}{3} \pi r^2 dr = 4 \pi r^2 dr$$

and dividing both sides by $V = \tfrac{4}{3} \pi r_o^3$ we get

$$\frac{dV}{V} = \frac{4\pi r^2 dr}{\tfrac{4}{3}\pi r_o^3} \text{ or}$$

$$dr = \tfrac{1}{3} \frac{dV}{V} \frac{r_o^3}{r^2}$$ thus the radial displacement is

$$\Delta r = \tfrac{1}{3} \frac{\Delta V}{V} \frac{r_o^3}{r^2} \; ; \quad (r \geqslant r_o)$$

Again by definition the radial strain is

$$\varepsilon_r = \frac{dU}{dr}$$ where U is the radial displacement (i.e. Δr)

thus $$\varepsilon_r = \frac{d\left(\tfrac{1}{3} \frac{\Delta V}{V} \frac{r_o^3}{r^2}\right)}{dr} = \tfrac{1}{3} \frac{\Delta V}{V} r_o^3 \, (- 2r^{-3})$$

therefore $\varepsilon_r = -\frac{2}{3} \frac{\Delta V}{V} \frac{r_o^3}{r^3} \quad (r \geqslant r_o)$ . . . . (1)

The displacements are maximal near the surface of r_o so that any plastic deformation will start at this surface. Let the concentric sphere 'C' of Fig.48A be the surface separating the plastic ($r_o \leqslant r \leqslant C$) from the elastic region ($C \leqslant r \leqslant b \to \infty$).

In the elastic region Hooke's Law holds, giving

$$\varepsilon_r = \frac{1}{E}\left[\sigma_r - \nu(\sigma_\theta + \sigma_t)\right]$$

$$\varepsilon_t = \frac{1}{E}\left[\sigma_t - \nu(\sigma_r + \sigma_\theta)\right]$$

$$\varepsilon_\theta = \frac{1}{E}\left[\sigma_\theta - \nu(\sigma_r + \sigma_t)\right]$$

where

$\varepsilon_r, \varepsilon_t$ and ε_θ = the principal strains (see fig.48B)

E = The Youngs modulus of the austenite surrounding the sphere r_o

$\sigma_r, \sigma_t, \sigma_\theta$ = the principal stresses (see fig.48B)

ν = Poisson's ratio.

From symmetry $\sigma_\theta = \sigma_t$ and thus

$\varepsilon_r = \frac{1}{E}(\sigma_r - 2\nu\sigma_t)$ or

$E\varepsilon_r = \sigma_r - 2\nu\sigma_t \quad (r \geqslant C)$ . . . . (2)

and

$$\varepsilon_t = \varepsilon_\theta = \frac{1}{E}\left[\sigma_t - \nu(\sigma_r + \sigma_t)\right]$$

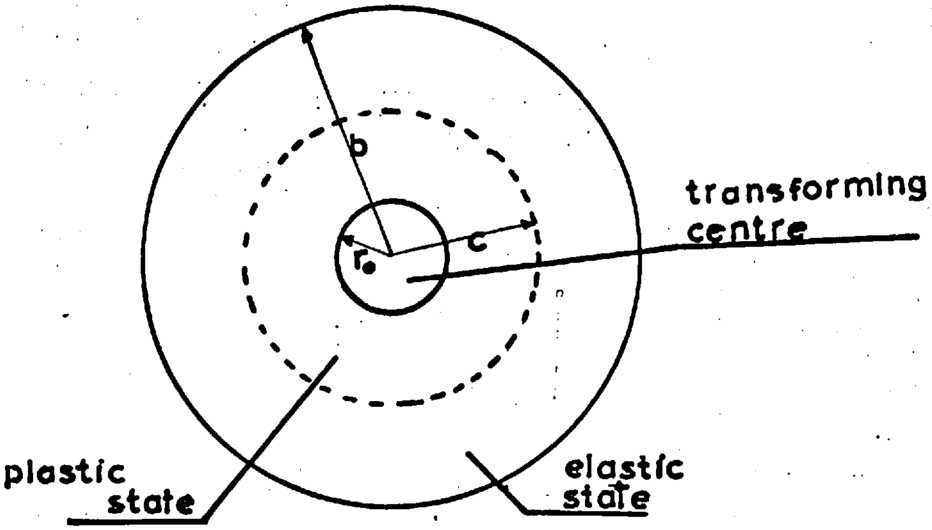

Fig. 48 A. Schematic drawing of a spherical transforming centre with radius r_o completely embedded in bulk material. The surface of the sphere with radius c separates the plastically deforming region from the elastically deforming one.

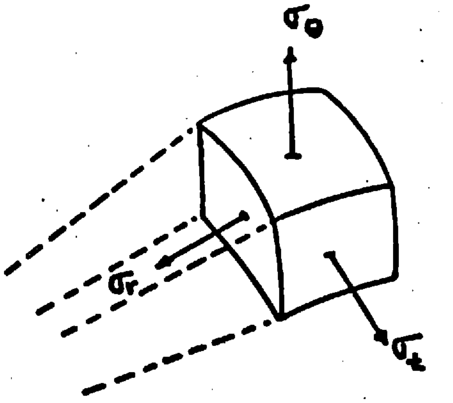

Fig.48 B. The principal stresses acting at any one point within the sphere b.

Since $\varepsilon_r + \varepsilon_t + \varepsilon_\theta = 0$

$$\frac{1}{E}(\sigma_r - 2\nu\sigma_t) + 2\frac{1}{E}\left(\sigma_t - \nu(\sigma_r + \sigma_t)\right) = 0$$

$$\sigma_r - 2\nu\sigma_t + 2\sigma_t - 2\nu\sigma_r - 2\nu\sigma_t = 0$$

$$2\sigma_t(1 - 2\nu) + \sigma_r(1 - 2\nu) = 0 \qquad \text{so}$$

$$2\sigma_t = -\sigma_r \quad \ldots\ldots\ldots\ldots\ldots\ldots \quad (3)$$

In the plastic region the stresses satisfy the yield criterion(88), for which the criterion of maximum shear stress is adopted here. According to this criterion, plastic flow sets in when the maximum shear component $\tau_{max.}$ of the stress acting on a point equals the critical shear stress τ_o. As a first approximation $\tau_o = \frac{1}{2}Y$ where Y is the yield stress measured in tension(89). The maximum shear component of the stress is given as(88)

$$\tau_{max} = \frac{\sigma_t - \sigma_r}{2}$$

The material yields when

$$\tau_{max} = \frac{\sigma_t - \sigma_r}{2} = \tau_o = \frac{1}{2}Y \quad \text{or}$$

$$\sigma_t - \sigma_r = Y \ \ldots \ (r_o \leq r \leq C). \ \ldots\ldots \quad (4)$$

At the surface r = C (fig.48A) both the equations (1) and (2) are valid. Thus

$$E.\varepsilon_r = \sigma_r - 2\nu\sigma_t$$

$$\underline{Y = -\sigma_r + \sigma_t}$$

$$E\varepsilon_r + Y = \sigma_t - 2\nu\sigma_t$$

$$\sigma_t = \frac{E\varepsilon_r + Y}{1 - 2\nu} \quad \ldots\ldots\ldots\ldots (5)$$

Substituting the values of σ_t from equations (3) and (5) into equation (4), it follows,

$$\frac{E\varepsilon_r + Y}{1 - 2\nu} + 2\frac{E\varepsilon_r + Y}{1 - 2\nu} = Y \quad \text{or}$$

$$3(E\varepsilon_r + Y) = Y(1 - 2\nu) \quad \text{or}$$

$$\varepsilon_r = \frac{Y - 2\nu Y - 3Y}{3E} = \frac{-2Y}{3E}(1 + \nu) \quad \ldots\ldots\ldots\ldots (6)$$

and from equation (1)

$$\varepsilon_r = -\frac{2}{3}\left(\frac{\Delta V}{V}\right)\frac{r_o^3}{r^3} \quad \text{and at } r = C$$

$$\varepsilon_r = -\frac{2}{3}\,\frac{\Delta V}{V}\,\frac{r_o^3}{C^3}$$

substituting this into equation (6) it follows,

$$-\frac{2}{3}\,\frac{\Delta V}{V}\,\frac{r_o^3}{C^3} = -\frac{2}{3}\,\frac{Y}{E}\,(1 + \nu) \quad \text{or}$$

$$\frac{C}{r_o} = \left[\frac{1}{1 + \nu} \cdot \frac{E}{Y} \cdot \frac{\Delta V}{V}\right]^{1/3}$$

The above expression denotes that when $C > r_o$, i.e. if $\frac{C}{r_o} > 1$, the change in volume $\Delta V/V$ of the transforming centre causes plastic deformation. On the other hand, when $r_o > C$, i.e. if $\frac{C}{r_o} < 1$, then only elastic deformation will occur. Substituting the proper values of 42×10^5 lb/in^2 for Young's modulus$^{(90)}$ at about (300°C) M_s temperature and 22,000 lb/in^2 for the yield strength of austenite near the transformation temperature and 4.3% for the volume change and the value of 0.3 for Poisson's ratio, into the above equation, it follows

$$\frac{C}{r_o} = \left[\frac{1}{(1 + 0.3)} \cdot \frac{42 \times 10^5}{22{,}000} \cdot 0.043\right]^{1/3} \cong 1.8$$

Since the value obtained for C/r_o is larger than one, C is greater than r_o; thus, the volume change due to the transformation is great enough to cause plastic deformation in austenite surrounding the martensite plate.

In fact some metallographic examinations have shown that plastic deformation takes place in austenite during the martensitic transformations in steels. In this line the work of Kelly and Nutting is of great potential interest(91,92). By transmission electron microscopy they showed that the austenite surrounding a martensite plate was plastically deformed, and the observed dislocation density particularly near the tip of a martensite plate was as high as 10^{12} dislocations per sq. cm.(91).

The spontaneous deformation of austenite during martensitic transformations was also studied by Edmondson and Ko(93). They have shown that in the iron-34% nickel alloy:

a) Plastic deformation was taking place during the formation of martensite mainly by slipping in the austenite. They also observed the slip lines in austenite in 1% carbon, 18% nickel and 1.5% carbon and 5% nickel steels.

b) Plastic deformation also occurred during the reverse transformation from martensite to austenite during heating. This occurred mainly by deformation in the regions originally occupied by martensite, although some slipping also occurred in the untransformed austenite. The residual deformation in austenite after a complete cycle of cooling and heating transformation consisted of slipping in austenite, severe deformation in the formerly martensitic regions and permanent surface tilt of the formerly martensitic regions. All these observations show that a spontaneous deformation in austenite adjacent to martensite plates occurs during transformation of austenite to martensite.

The second point which should be noted is that transformation plasticity occurs only in a range of some 40°C below M_s during which range only approximately 70% of martensite forms in En 30 A steel and in the iron-20 wt% nickel alloy. The value of total extension "ε" in the equation (i) is associated with (100%) complete transformation. The total extension obtained during transformation is a linear function of the amount of material transformed at a constant stress when the crystals of the new phase grow in arbitrary directions(29). Thus, the first term in the equation (i) "$\frac{2}{3}\frac{\sigma}{Y}\frac{\Delta V}{V}$" should be multiplied by the volume fraction of the martensite transformed in the temperature range where transformation plasticity occurs.

Another point to note is that the deformation occurs as transformation of austenite to martensite proceeds with falling temperature. From the simple dependence of yield stress on temperature it is obvious that the yield stress of austenite will not be a constant. Apart from the variation of yield stress with falling temperature, there is the added complication that the formation of martensite strengthens the remaining parent phase. However, a large proportion of the plastic deformation occurs only in a small temperature range (i.e. 10°C), and the change in yield stress, due to the simple temperature dependence, over the range of 40°C is only about 1-2000 lb/in^2. Therefore it was decided, for En 30A steel, to use the value of austenite yield stress at 300°C and for the iron-20 wt% nickel alloy the value of austenite yield stress at 250°C in all the calculations.

II. Effect of the magnitude of applied stress:

It has been shown that stresses less than the normal yield stress of austenite produce considerable plastic deformation as the specimen cools through M_s and the extent of the deformation, for a given austenitizing treatment (e.g. three minutes at 850°C), has been found to increase with increasing stress. Figs. 21 and 22 show that the initial portions of the curves, at low stresses, are in agreement with the predictions of the equation (i), since there is a linear relationship between the extent of transformation plasticity and the magnitude of the applied stress as predicted. The applicability of the equation (i) to transformation plasticity during martensite formation was checked by considering the effect of the magnitude of applied stress for a given austenitizing treatment which determines the value of austenite yield stress. The yield stress of austenite in the vicinity of the normal M_s (307°C), obtained by a standard treatment (3 minutes at 850°C), is about 21,000 lb/in^2 as found from Fig.33 for En 30A steel. Reference to Fig.21 shows that this value agrees well with the value of stress where the total extension vs. stress curve departs from linearity. A calculation from equation (i) gives a value of 2% for a stress of 21,000 lb/in^2, which is in good agreement with the experimental value of 1.7% (curve b of Fig.21) for En 30A steel. In the case of the iron-20 wt% nickel alloy the linear relationship in the initial part of the plot of total extension percentage vs. stress for austenitizing at 850°C (Fig.22) is also in agreement with the equation. Taking the value of austenite yield stress at M_s after a standard austenitizing treatment at 850°C as approximately 13,000 lb/in^2, a calculation from equation (i) gives values for total extension percentages of approximately 1.4% and 2.2% for stresses of 8,000 lb/in^2 and 13,000 lb/in^2 respectively applied at M_s. These are in good agreement with the experimental values for the extension of approximately 1.3% and 1.9% for these respective stresses (Fig.22). The departure from linearity in Fig.22 should occur at a stress corresponding to the austenite yield stress and this is, in fact, the case.

In the higher stress range (above the yield stress) the amount of transformation plasticity increases at an increasing rate as the stress is increased for both the En 30A steel and the iron-20 wt% nickel alloy. This is most probably associated with plastic deformation resulting from stresses in excess of the yield stress.

III. Effect of temperature of stressing:

With reference to Fig.23, the amount of extension obtained during transformation to martensite is very closely dependent on the temperature of stressing. The slopes of the linear portions of the total extension vs. stress curves (Fig.21) are substantially less when the specimen has been stressed below M_s than when stressed at M_s, and increase as the stressing temperature is increased towards M_s. With decreasing slopes of the linear portions of the extension vs. stress plots the extent of the linear portion of the curves increases. This can be attributed to the following factors:- In the first place, when the specimen is stressed below the M_s temperature there will already be a certain amount of martensite present resulting in a smaller amount of volume change occurring between the temperature of stress application and that at which transformation plasticity ceases. The effect of this is to reduce the amount of plasticity that occurs for a given stress, because the total extension is dependent on the amount of martensite transforming at a constant applied stress. Secondly, the yield stress of the austenite is expected to be increased by the formation of some martensite. As discussed below, this is due to the introduction of fresh dislocations into the austenite.

As a martensite plate forms, more imperfections are generated in the surrounding austenite by plastic deformation due to the volumetric and shear strains. By increasing the obstructions to the movement of slip dislocations in the slip planes by the increased number of imperfections of various kinds, plastic deformation of austenite will naturally be rendered more difficult. Moreover, the martensite plates present at the moment of stress application are themselves barriers to the dislocation movement. Thus, the amount of transformation plasticity will decrease as the specimen is stressed below M_s. These factors which reduce the transformation plasticity, should also increase the stress range within which the linear relationship holds.

With reference to the curves 'c' and 'd' of Fig.21, it is difficult to determine the stress values at which the curves depart from linearity. However, a reasonable estimation would give values of about 30,000 lb/in^2 and 40,000 lb/in^2 for stressing temperatures of 295°C (12°C below M_s) and 280°C (27°C below M_s) respectively. Assuming that these are the proper values of austenite yield stress at the temperature of stress application and using the appropriate $\Delta V/V$ values corresponding to the volume change occurring during the period between stress application and that at which transformation plasticity ceases, it is calculated that the total extension values at the end of the linear portion of the graphs should be approximately 1% and 0.3% for curves 'c' and 'd' respectively. The agreement between these values with the experimental results is quite reasonable.

IV. Effect of Austenite Yield Stress:

The results have shown that austenitizing conditions influence the total amount of deformation that occurs when loads are maintained as a specimen cools through the M_s temperature. Since the yield stress of austenite varies considerably with the austenitizing treatment given, the dependence of transformation plasticity on the yield strength of austenite can be studied by using the results of the experiments on various austenitizing treatments. The variation of austenite yield stress with various austenitizing conditions is shown in Fig.33 for En 30A steel and in Fig.34 for iron-20 wt% nickel alloy. The yield stress is considered to be controlled: a) by the dislocation concentration, and b) by austenite grain size both of which are determined by the austenitizing conditions.

a) The effect of dislocation density and defect structure on the yield stress of austenite:

It is believed that high concentration of defects is introduced into austenite when a martensitic alloy is rapidly heated into the austenite range, as a result of the transformation volume change. When austenite forms either directly from martensite (as in the case of iron-nickel alloys) or from a ferrite and carbide aggregate (as in the case of steels) during rapid heating, it is expected that a relatively high concentration of lattice defects (viz. dislocations, vacancies and stacking faults) will be introduced into the austenite due to the contraction in volume associated with the transformation. The basis of this idea has been previously proposed by Kurdjumov(94).

Evidence is available relating to the dislocation structure of austenite produced by the reverse martensitic transformation in iron-nickel alloys. As mentioned earlier, plastic deformation of the austenite as a result of this transformation has been reported by several workers(93,94). Tangled and jogged dislocations, dislocation loops, and high dislocation densities (up to $10^{11}cm^{2}$) were observed in iron-30% nickel alloy specimens heated to about 50°C above A_f (i.e. 450°C), and austenite was found to be strengthened by this reverse transformation effect (95,96). Krauss discussed the aspect that vacancies created at the interface as a consequence of the changes in specific volume may be important in producing the dislocation arrangements.

These defects in austenite will tend to anneal-out at a rate depending on the austenitizing temperature and time, involving processes of dislocation climb and interaction, and interaction of dislocations and austenite grain boundaries; the stacking fault energy of the austenite will influence the 'annealing-out' process.

In particular, raising the austenitizing temperature will not only increase the austenite grain size but will also increase the mobility of dislocations, thereby decreasing the dislocation density. Moreover, it may be suggested that, during the grain growth, the boundaries will move and will act as a sink for the dislocations in austenite and will further decrease the dislocation density. In general it may be suggested that the treatments which increase the dislocation density strengthen the austenite, and any treatment which reduces the dislocation density reduces the austenite yield stress.

b) Effect of grain size on the yield strength:

With regard to the contribution of grain size to austenite yield strength, this is not known accurately, but it is considered likely that the grain size provides very little strengthening contribution.

The relation between yield stress and grain size has been differently reported by different workers. Edwards, Pfeil(97), and Edwards, Jones and Walters(98), and Winlock and Leiter(99) found a linear relation over most of the range when plotted against 1/grain size and Petch(100) and Stroh(101) a linear relationship when plotted against $1/(G.S.)^{1/2}$.

Petch has discussed the relation, supposing that yield takes place when a critical stress is reached at the head of a pile-up of dislocations pressing on a grain boundary, and developed an equation to show the relationship between yield stress and grain size. The dependence of yield stress on the grain size was given according to an equation shown below:

$$\tau = \sigma_i + ky,(d)^{1/2}$$

where 'τ' is the yield stress of the material, 'σ_i' and 'ky' are two constants and 'd' is the grain diameter.

It is worth pointing out that the theory established by Petch implies that the interior of the ferrite grains has the same properties in fine and coarse grained steel. However, it should be remembered that the experiments which are required to find the grain size dependence of yield stress involve a number of samples with different grain sizes, obtained by various austenitizing treatments: it may be questioned whether these different austenitizing treatments result in identical dislocation structures with the same dislocation density and identical distribution of point defects in coarse and fine grained material.

If one looks more carefully into Petch's equation, it can be seen that the terms 'σ_i' and 'ky' are dependent on the dislocation density and dislocation configuration of the material. 'σ_i' is the lattice friction stress which resists slip because of the forces exerted by solute atoms, precipitates and pinned dislocations on dislocation arrays.

The second term 'ky' is the shear stress required ahead of the pile-up to unlock a dislocation and is made up of two terms:

$$ky = T_s l^{1/2}$$

where, 'T_s' is the stress needed to move a dislocation just ahead of a slip band; it measures the strength with which the dislocation is locked in position.

'l' is the distance between the head of a slip band and the nearest source of dislocations.

The dependence of the two constants on dislocation density and configuration suggests that the yield stress of a material is much more dependent on the dislocation density than grain size. In fact some experiments have shown that in conditions of strong dislocation locking, the grain size dependence of yield stress disappears completely(102). There are some other data in the literature(103) for titanium and titanium-aluminium alloys in which yield stress shows no dependence on grain size.

The results obtained in the present investigation on En 30A steel provide evidence that grain size, as such, within the range studied has little effect on the yield strength of austenite, since a constant value of grain size is obtained for slow heating and normal austenitizing treatments giving different yield stresses.

On the basis of the present results, it seems reasonable to assume that the predominant factor controlling the yield strength of austenite is the dislocation structure (i.e. the defect structure of austenite), introduced during heat treatment. As the austenitizing temperature is increased, the mobility of dislocations increases and they consequently interact and tend to annihilate more easily, thereby lowering the dislocation density and the yield stress. Thus, a high value of total extension during martensite formation is expected to occur from a low value of yield stress associated with a low dislocation density, obtained by high temperature austenitizing treatment. When the dislocation density is high as in short-time, low temperature austenitizing treatments the deformation during transformation is low and the yield stress is high. In general, for the standard austenitizing times or times longer than these, the actual temperature of austenitizing will be the main factor determining the final dislocation content, and thus, the yield stress of austenite. However, in the lower austenitizing temperature range, and for relatively short austenitizing times, the actual time will be important.

It is of interest to note that the change of austenite yield stress with austenitizing temperature in En 30A steel is more marked in the lower temperature portion of the austenitizing range, i.e. below approximately 1000°C, although increasing the austenitizing temperature above 1000°C leads to further reduction in austenite yield stress. In terms of the ideas discussed above it appears that the reduction in initial dislocation density of the austenite with increase in temperature is most marked in this lower temperature range; the rate of annealing-out of the defects seems to be greater the higher their actual concentration.

The variation of austenite yield stress with austenitizing conditions is somewhat different in the iron-20 wt% nickel alloy. In this alloy, heating a martensitic structure to 600°C or 650°C produces austenite of relatively high strength. The concentration of defects in the austenite is expected to be high, and possibly (if it occurs), recrystallization is not complete in this temperature range with the austenitizing time used.

With regard to the observed grain sizes, it is of interest to note that the work by Krauss and Cohen(96) on iron-nickel alloys containing 30% nickel, showed that the austenite grain size remained unchanged through a cycle of forming martensite and reverting it to austenite.

Work by Sastri(84) on iron-20 wt% nickel alloy showed large differences in austenite grain size obtained by using different austenitizing temperatures, viz. 3×10^{-2} mm at 600°C, and 8×10^{-2} mm at 1000°C. However, in the present investigation, with the same alloy, the austenite grain size was found to have changed only from 7.9×10^{-2}mm at 600°C to 9×10^{-2}mm at 1000°C for the same times of austenitizing as used by Sastri(84). The latter results seem to agree more closely with the conclusions of Krauss and Cohen(96) regarding the constancy of austenite grain size than with those of Sastri. Thus, there appears to be some uncertainty in this alloy about the recrystallization behaviour of the austenite during the martensite to austenite transformation. (See appendix III).

It was interesting to observe that austenitizing the specimen at 800°C rendered the alloy very brittle and the microstructure showed very heavily etched grain boundaries; this might indicate that some grain boundary segregation might have occurred. This is shown in Fig.49. Annealing the alloy at higher temperatures seemed to remove this intergranular brittleness and the boundaries did not etch so heavily; possibly

Fig. 49. Heavy etching of the grain boundaries, heated to 800°C for 3 minutes. X200

dissipation of the segregated elements had occurred.

Experiments have shown that increasing the austenitizing temperature leads to a reduction in austenite yield stress. This can be explained as follows; Dislocations can be eliminated either during recrystallization by the migration of grain boundaries or by climb or cross-slip. All these phenomena are time and temperature dependent and thus at high austenitizing temperatures, the annealing-out of defects would occur more rapidly. The fact that a change in austenitizing temperatures between 700°C and 1000°C does not appreciably affect the austenite yield stress on iron-20 wt% nickel alloy suggests that an essentially constant dislocation density may be attained. The reduction in austenite yield stress of the same alloy for austenitizing temperatures of above 1000°C could be due to a significant decrease in dislocation density. The explanation of the difference in behaviour between the iron-20 wt% nickel alloy and En 30A steel with respect to the effect of variation in austenitizing conditions on austenite yield stress is not clear. It may be connected with the very low carbon content of the iron-20 wt% nickel alloy since the low carbon content of the alloy allows easy dislocation movement because of only a small interaction effect between carbon atoms and dislocations. Another factor could be a higher stacking fault energy in the iron-20 wt% nickel alloy as compared with the alloy steel. The stacking fault energy of austenite influences the annealing-out process of dislocations by affecting the readiness of dislocations to cross-slip and climb. Consequently the systems with higher stacking fault energy will reach an equilibrium dislocation density sooner than the specimens with lower stacking fault energy.

The linear relationships obtained in the yield stress vs. temperature graphs (Fig.33) for various austenitizing conditions for En 30A steel, were extrapolated to 300°C. The yield stress values thus determined at 300°C were used in plotting the reciprocal of yield stress vs. total extension for stresses less than the austenite yield stress. According to equation ($_i$), such a plot should yield a straight line and the data shown in Fig.50 are in agreement with the predicted linear relationship. In the case of the 13.100 lb/in^2 curve, the deviation in the linear graph is likely to be due to effect of the stress being larger than the yield stress of austenite obtained after an austenitizing treatment at 1200°C. From the slopes of the lines, together with the appropriate values of volume change and stress, the value of the constant 'A' may be evaluated with the help of equation (^{i}). For each of the values of applied stress the calculated value of 'A' is 0.5, and is in

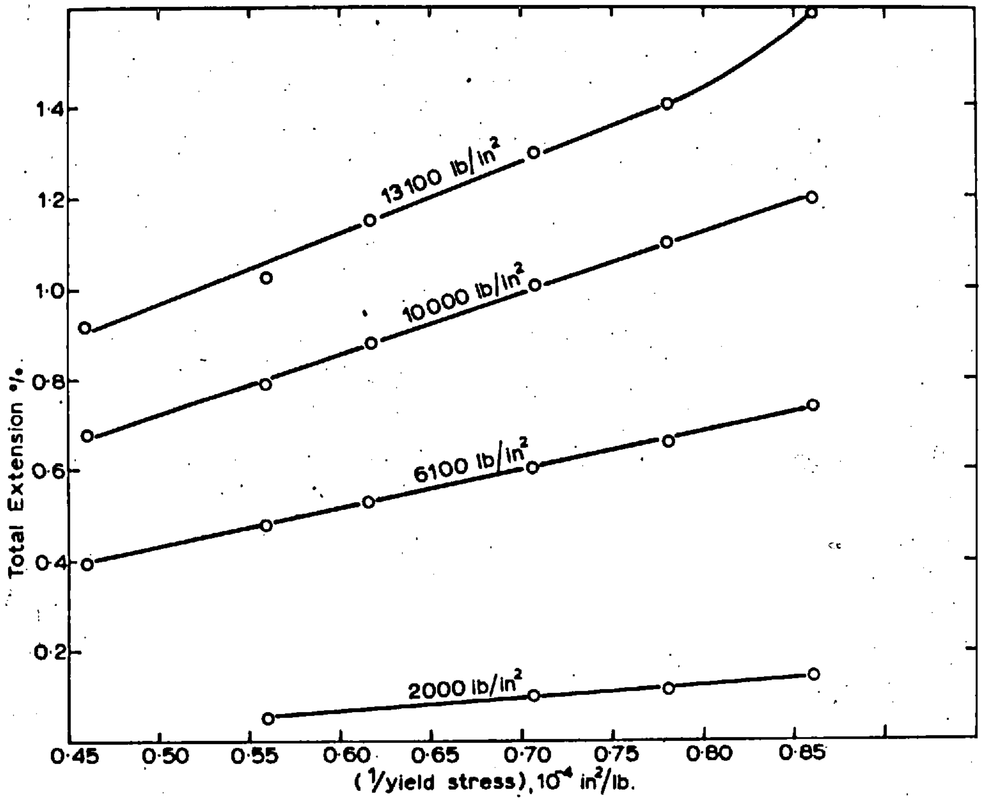

Fig.50. Total extension vs reciprocal of austenite yield stress for En 30A steel.

reasonable agreement with that derived by Johnson and Greenwood of 2/3, considering the uncertainties involved in the assumptions for the derivation of the equation (i).

With reference to the iron-20 wt% nickel alloy, comparison may also be made between the values of total extension obtained through equation (i) and that obtained by the experiments for different austenitizing conditions. For example, the austenitizing temperatures selected were 650°C and 700°C and the stress applied was 8000 lb/in^2. The values of total extension obtained for these conditions by the equation (i) were 0.6% and 1.4% and the experimental values for the same condition were 0.55% and 1.3% respectively, which show a good agreement. In the case of the iron-20 wt% nickel alloy, to find the values of austenite yield stress at the transformation temperature, the yield stress vs. temperature graph was extrapolated to 250°C which is the M_s temperature for this alloy.

The applied tensile stresses of 8000 lb/in^2 and 23,000 lb/in^2 were found to be below and above the austenite yield stress respectively, for any austenitizing conditions employed, and consequently the latter stress condition cannot be utilized for the equation (i). The austenitizing temperature range, for which transformation plasticity is essentially constant, corresponds to the range where the austenite yield stress is unaffected by austenitizing temperature. This is in accordance with the predictions of equation (i) that the transformation plasticity is directly dependent upon the austenite yield stress. It is of interest to note that the data for the 23,000 lb/in^2 stress as well as for the 8000 lb/in^2 stress, show the constancy of transformation plasticity over the range 700°-1000°C, even though equation (i) does not hold for the former stress which is greater than the austenite yield stress.

It may also be significant to note that the application of stress can affect the transformation kinetics, viz., the rate of martensite formation, and under the application of large stresses adiabatic heating can occur and thus the amount of plasticity that occurs can be affected.

The work on En 30A steel wire specimens showed that the total amount of deformation that occurred when stresses larger than the austenite yield stress were maintained as a specimen cools through M_s, increased at an increasing rate with increasing amount of stress (Fig.27), and the effect of stress was particularly marked in the case of higher austenitizing temperatures employed (viz. above 1,100°C). This can be attributed to the following factor:

It is well established that the application of stress can affect the transformation kinetics. When a uniaxial stress is applied, the work done on or by transformation (as the acting forces are carried through transformation displacements) contributes algebraically to the free energy change, thus accelerating the transformation rate. On the other hand, application of large stresses (above the yield stress of austenite) will plastically deform the matrix. Under this condition, adiabatic heating could occur ~~because of~~ due to the increased amount of heat liberation (~~because of~~ resulting from the increased rate of transformation) which will superimpose with the heat generated due to the plastic deformation and thus change the temperature range in which plasticity occurs. The amount of adiabatic heating increases as the stress is increased or as the yield stress of austenite is decreased by a high temperature austenitizing treatment. An increase in temperature means a reduction in yield stress and since the amount of adiabatic heating increases with stress or with increase in austenitizing temperature, the plastic deformation will also increase as the stress is increased or as the specimen is austenitized at higher temperatures.

Another interesting point to be noted is the change of M_s temperature with austenite yield stress. When the M_s values for the various treatments were plotted against the corresponding yield stress at M_s, a straight line was obtained for both materials investigated; Figures 35 and 36 show the results for the En 30A steel and iron-20 wt% nickel alloy respectively. The above results indicate that the force restraining the propagation of an embryo may be directly dependent on the yield stress of austenite. The idea that the shear strength of the austenite lattice may influence martensite formation has been used by previous workers(104).

It is believed that the actual strength of austenite may have a direct contribution to the non-chemical free energy difference between austenite and martensite in the embryo theory of Knapp and Dehlinger(105), since the strain energy associated with the formation of a martensite plate will be increased by an increase in yield stress. This is justified because the strain associated with the formation of a plate exceeds the yield point of austenite; the only assumption which has to be made here is that the mean strain resulting from the formation of the plate is independent of its size.

It is also thought that the presence of dislocations in the austenite might cause deviation from the optimum embryo shape, and the value of non-chemical free energy difference between austenite and martensite would therefore be greater(105). This in turn will increase the driving force necessary for the formation of the martensite and so the M_s temperature may be expected to be lowered. The interaction of carbon atoms with dislocations may also be important in restraining embryo propagation.

i) Short time treatment:

The results for transformation plasticity obtained from treatments involving a very rapid heating to the austenite range and immediate quenching showed less deformation than for a conventional long time treatment. The difference in the extent of transformation plasticity between short-time and normal austenitizing treatments in the En 30A steel at 850°C may also be explained on the basis of the difference in yield stress of the austenite produced by these two treatments. The austenite, produced by rapid heating of martensite to austenite and quenching immediately after holding for a very short time (one second) at relatively low austenitizing temperatures, has a much higher yield stress than the austenite produced by normal austenitizing treatments. It is suggested that the density of dislocations and accompanying defect structures in austenite must be very high, because the very short austenitizing times employed are insufficient for the dislocations to interact and thereby annihilate one another. The austenite yield stress at 300°C following a normal austenitizing treatment was about 21,000 lb/in^2 while a value of about

37,000 lb/in^2 was obtained following a one second austenitizing treatment at 850°C. These values of yield stress may be utilized in calculating the total extension at, say, 21,000 lb/in^2 with the help of equation (i). For the two indicated treatments the calculated values of total extension are 2% and 1.1% respectively. The measured values of the same data as shown in Fig.30 are in close agreement with the calculated values. The plot of the total extension vs. applied stress for the short-time treatment deviates from linearity at between 30-35,000 lb/in^2, and this value appears to be somewhat less than the actual measured value of the austenite yield stress (viz. about 37,000 lb/in^2), so that the agreement with the value predicted by equation (i) is not so close in this case.

The observed decrease in the difference of total extension between short time and normal austenitizing treatments with the increase of austenitizing temperature (Fig.31) can be attributed to the following factor. It was suggested that the annealing-out of the dislocations will occur relatively easily at higher austenitizing temperatures; thus an essentially constant state would be reached in shorter times. This in turn will decrease the difference between the dislocation densities obtained by the two types of austenitizing treatments. Once a constant state is reached, longer times of holding at the temperature will not alter the constancy of the dislocation density and as a consequence of this, a constant value of yield stress will be obtained. Thus after a certain level no difference in the total extension percentage during transformation between the two types of treatment will be observed.

ii) Slow heating through A_{c1}-A_{c3} range:

The transformation plasticity results of Fig.32, showing a greater amount of transformation plasticity in specimens slowly heated during austenitizing as compared with rapidly heated specimens, may also be attributed to a difference in austenite yield stress for these two types of treatment.

The formation of austenite either direct from martensite, or from a ferrite-carbide aggregate, must inherently introduce defects into the structure of austenite. The faster ~~is~~ the heating rate at which austenite is formed, the greater might be the defects introduced in the structure because of the sudden volume change involved.

The yield stress measurements on the specimens slowly heated through A_{c1}-A_{c3} range showed that the austenite was relatively weaker than the specimens quickly heated (100°C/sec.). The extrapolated value of austenite yield stress at 300°C following a slow heating experiment was about 22,000 lb/in^2 as compared with a value of 27,000 lb/in^2 following a normal austenitizing treatment of three minutes at 800°C. The above mentioned slow heating experiment consisted of normally heating the specimen (100°C/sec.) to a temperature of 690°C and then slowly heating from 690°C to 730°C in 60 minutes, and finally normally heating (with the initial rate) to 800°C and holding for three minutes.

Grain size measurements showed that varying the heating rate through the A_{c1}-A_{c3} range for the En 30A steel did not significantly change the austenite grain size. The measured grain diameters were 1.08×10^{-2} mm for slow heating, and 1.04×10^{-2}mm for the normal heating experiment. Therefore it may be concluded that the reduction in the austenite yield stress is due to a reduction in the dislocation density. Thus the actual rate of heating through the critical range appears to affect the yield stress by influencing the concentration of defects formed in austenite.

Using the values of yield stress for slow and normal heating rates, equation (1) may be applied to calculate the magnitude of total extension for checking it with the experimental value. For example, for a stress of 22,000 lb/in^2 the calculated value of total extension for normal and slow heating would be approximately 1.6% and 2% respectively. Reference to Fig.32

shows that these values agree closely with the experimental data.

iii) Double austenitizing treatments:

The results of double austenitizing treatments show that the maximum austenitizing temperature controls the transformation plasticity. This suggests that the yield strength of austenite is not altered by holding the specimen at the second holding temperature. This is in accord with the defect hypothesis. During the high temperature austenitizing treatment defects are annealed out from the austenite lattice, and lowering of the temperature does not re-introduce these defects. Thus the yield stress of austenite would depend upon the condition set by the higher temperature treatment.

iv) Effect of austenite stabilization:

The magnitude of the transformation plasticity occurring on cooling through M_s is reduced by the application of stresses larger than the austenite yield stress above the M_s temperature. This is shown in Figs.25A and 25B. The reduction in the extent of transformation plasticity is attributed to the increased value of yield stress due to the following two causes: These are, first, the plastic deformation taking place above M_s and, second, the formation of a small amount of stress-induced martensite. An increase in the yield stress of austenite will be expected as a function of austenite deformation; (strain hardening of austenite) larger deformations contributing to higher yield stresses. The formation of stress-induced martensite between M_s and M_d will also increase the yield stress by increasing the obstructions to the movement of slip dislocations as they are themselves barriers. The effect of the existence of a relatively small amount of martensite in reducing transformation plasticity is known and is illustrated by curve 'd' of Fig.25B. The reduction in transformation plasticity from curve 'd' to curve 'e' of Fig.25B is therefore due to the strain hardening of austenite. As expected, larger deformations will result in smaller transformation plasticity, which is in accord with the strength hypothesis.

The stabilization of austenite by thermal treatment (at 302°C, 300°C and 295°C) has the same effect as the stabilization produced by mechanical treatment on the transformation plasticity. One suggestion that has been made is that isothermal ageing increases the stability of partially transformed austenite, due to the formation of carbon clusters, which increase the resistance to the movement of dislocations and this in turn raises the yield stress of austenite. Thus, the carbon might be responsible for the stabilization in steel. This is consistent with the preliminary estimate of the activation energy for stabilization. The value of about

30,000 cal. per mole which is found by Morgan and Ko(106) is of the same order of magnitude as the value of the activation energy for the diffusion of carbon in austenite.

The carbon grouping might occur either at dislocations by forming Cottrell atmospheres or anywhere within the matrix. If it occurs at dislocations by forming Cottrell atmospheres, the time required for stabilization to take place should be that required for carbon atoms to move a distance greater than that between carbon atoms but smaller than that between dislocations. Using the approximation $t=X^2/D$; where 't' is the time required for stabilization, 'X' is the length of the diffusion path and 'D' is the diffusion coefficient of carbon in γ-iron, a rough estimate can be made of the length of diffusion path at the temperature of around 300°C(106). Around 300°C $D \cong 10^{-13}$ cm^2/sec(107), the time required for stabilization is of the order of 100 seconds, therefore $X = \sqrt{Dt}$ or $X = \sqrt{10^{-13} \times 10^2} = 10^{-5}$ to 10^{-6} cm; which seems of the right order of magnitude required by the mechanism.

The results of Fig.26 are consistent with the above mentioned theory of thermal stabilization. A reduction in transformation plasticity will occur only with an increase in yield stress of austenite since the other parameters involved in the equation (i) do not change by isothermal holding. The sharp fall in transformation plasticity after three minutes of isothermal holding time is believed to be associated with the formation of bainite during the period of isothermal holding.

v) <u>The structural factors and the mechanism involved in transformation plasticity</u>:

With regard to the complementary approach involving considerations of structure and mechanisms, two particular aspects may be noted; these are, first, the dislocation structure of the austenite and its relevance to the plastic flow that occurs during transformation, and, second, the preferred orientation of martensite plates formed under stress.

In terms of dislocation theory, it has been suggested(35) that the energy to move the dislocations out from their sources and to force arrays against the grain boundaries may be supplied by the phase transformation, and only a small additional stress might then be needed to activate slip in preferred directions in the adjacent grains. This explanation suffers from the disadvantage that it may be applied only to alloys of low stacking fault energy, as such dislocation pile-ups would not be expected in alloys of high stacking fault energy(108). An alternative proposal is based on the well-established observation that a high concentration of dislocations exists in austenite adjacent to martensite plates$^{(92)}$. The deformation produced when austenite transforms to martensite has been revealed as slip lines in the surrounding austenite. The thin foil technique provided an ideal method for revealing this deformation directly. It has been found that the deformation is greatest in the vicinity of the martensite plates themselves, and it is shown that these dislocations have been generated by the distortion which accompanies transformation. Besides deforming the surrounding austenite, the formation of a martensite plate has been noted to result in the deformation (internal twinning on a fine scale) of the martensite plate itself. The deformation of austenite during martensite formation may be related to the stressed region postulated in the macroplasticity approach by assuming that under the influence of an applied stress during the transformation, dislocation movement could readily occur in favourably oriented slip systems, thus producing the large extension during transformation. Slip causing transformation plasticity is unlikely to occur in martensite plates because the critical shear stress for dislocation movement is much greater than in austenite, especially since the plates are internally twinned. The preferred slip in the austenite round a martensite plate would then account for the observed extension. As the amount of martensite present in the structure increases with falling temperature, the austenite becomes more and more deformed, and by the interaction of increased amount of dislocations, the movement of slip dislocations becomes very limited, and hence the yield strength of the residual austenite will be increased; a strengthening effect will also be operative due to the martensite plates acting as barriers for dislocation movement. These effects, together with the reduction of the amount of untransformed austenite as the temperature decreases, provide an explanation of the cessation of detectable transformation plasticity at a temperature some 40°C below M_s.

It is possible that the formation of martensite plates with a preferred orientation may be a contributing factor to the occurrence of transformation plasticity. As described above, the formation of martensite plates with a preferred orientation was observed in the present work, and may be interpreted on the basis of Le Chatelier's principle; which, in effect, states that the mode of transformation chosen is such that the applied stress does the maximum amount of work, and it is assumed that the applied stress acts to propagate the 'growth' of the product plates.

With the aid of Le Chatelier's principle, the direction of the first formed stress-induced martensite plate can be determined. The most general type of the (invariant plane strain) transformation strain will have a normal ($\mathcal{E}$) as well as tangential (γ) component(109). The work done on or by the transformation due to the action of applied stress will be a maximum when the following relation is a maximum.

$$U = \tau_r \gamma + \sigma_r \mathcal{E} \quad \ldots\ldots\ldots \quad (1)$$

where 'τ_r' is the resolved shear stress along a potential habit plane, and 'σ_r' is the resolved normal stress perpendicular to the habit plane. Under an applied tensile stress, the numerical value of the normal stress 'σ_r' is positive, and it will be negative when the applied stress is compressive. 'τ_r' is always positive, because the many habit permutations virtually permit shearing in either sense.

To find the direction of the theoretical potential habit planes, in order that the above relation will assume its maximum possible value, the following procedure has been adapted:

Let θ be the angle between the normal to the potential habit plane and the direction of applied stress (the specimen axis).

In Fig.51 a macroscopic region of transformation of cross-sectional area 'A' is shown carrying an axial tensile load 'W'. The tensile load 'W' can be resolved into two components; a component 'N' perpendicular to the potential habit plane, and a component 'S' parallel to the potential habit plane. Let 'σ_r' and 'τ_r' be the stresses due to the 'N' and 'S' respectively and let $\sigma = W/A$ where 'A' is the area of section 'AB'. Therefore,

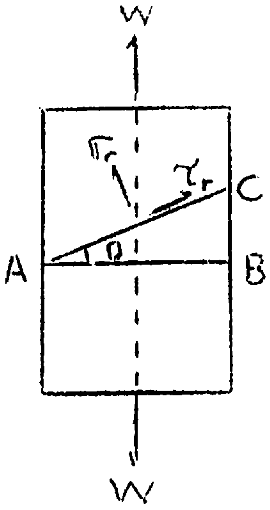

Fig. 51. The plane AC is the potential habit plane.

Area of section 'AC' $= A/\cos\theta$

Also $N = W \cos\theta$

and $S = W \sin\theta$

so, $$\sigma_r = \frac{N}{\text{area at AC}} = \frac{W \cos\theta}{A/\cos\theta} = \frac{W}{A}\cos^2\theta$$

$$= \sigma \cos^2\theta$$

and $$\tau_r = \frac{S}{\text{area at AC}} = \frac{W.\text{Sin}\ \theta}{A/\cos\theta} = \frac{W}{A}\ \text{Sin}\ \theta\cos\theta$$

$$= \sigma\sin\theta\cos\theta$$

'U' (the work done) may now be expressed as a function of the orientation of a transforming martensite plate.

$$U = \varepsilon\sigma\cos^2\theta + \gamma\sigma\sin\theta\cos\theta \quad \text{or}$$

$$U = \varepsilon\sigma\tfrac{1}{2}(1 + \cos 2\theta) + \gamma\sigma\tfrac{1}{2}\sin 2\theta$$

Since the main concern is given to the plates that form first (at M_s) under the influence of applied stress, it is necessary to find the particular orientation which yields a maximum value of 'U'.

$$\frac{du}{d\theta} = -\varepsilon\sigma\sin 2\theta + \gamma\sigma\cos 2\theta = 0$$

$$\frac{\sin 2\theta}{\cos 2\theta} = \tan 2\theta = \frac{\gamma}{\varepsilon}$$

The values of γ and ε are given as 20% and 5% respectively(110) for the iron 30% nickel alloy, but the dilatometric measurements in this investigation indicated that the unit bulk expansion was closer to 4% rather than to 5%. This is a more exact value of ε, because this value must correspond to the volume change as shear strains cannot cause a volume change.

Substituting the know components of the transformation strain into the above equation;

$$\tan 2\theta = \frac{0.20}{0.04} = 5$$

therefore, $\theta = 39.5°$, thus giving the orientation of the first martensitic plates to form on cooling under the applied stress. This value is in very good agreement with the observed one in the present work for preferred orientation (i.e. about 40°).

It is well established that elastic tensile stresses favour the formation of martensite plates and the plates which form initially are expected to be those most favourably oriented with respect to the applied stress system to relieve the applied stress. The tendency to a 'directional' effect in the martensite formation would mean that the transformation volume change might lead to extension of the specimen under quite low stresses. However, the present results indicate that this could not be a major factor controlling the mechanism of transformation plasticity. For instance, no significant preferred orientation was observed for a stress of 8,000 lb/in^2 despite the fact that transformation plasticity did occur. Furthermore, no significant preferred orientation occurred below about 295°C even though transformation plasticity continued for some 20-30°C below this temperature. Further evidence in support of the fact is that austenitizing conditions affect the extent of transformation plasticity; no experiments were made to see whether the preferred orientation effects varied with austenitizing conditions, but it seems unlikely that there would be much difference. The effect of austenitizing conditions in fact seems more logically explained in terms of the yield strength of the austenite.

The discussion to the problem may be summarized in stating that the approach by way of macroplasticity gives by far the most reasonable agreement between the experimental and theoretical values. The observations suggest that the transformation plasticity is dominated by three factors, namely the applied stress, the austenite yield stress and the transformation volume change accompanying phase transformation. In terms of structural considerations, various possible mechanisms may be invoked to explain the observed phenomena; it is suggested, however, that the main mechanism is the high concentration of dislocations in austenite adjoining martensite plates, and the movement of these dislocations in favourably oriented slip systems during transformation.

VI. Transformation Plasticity during Tempering:

Some experiments have been made on the En 30A steel to study plastic deformation during tempering of martensite. The quench hardened steel contained austenite in very small amounts. No retained austenite could be detected by metallography, but in the X-ray diffraction photographs the f.c.c. lines, characteristic of austenite, were always seen though very faintly. The observed plasticity during tempering results from the combined effects of the decomposition of the austenite and carbide precipitation from martensite. This occurs in three stages:

1) Precipitation of coherent close-packed hexagonal ε- carbide and the formation of a low carbon martensite.

2) Decomposition of retained austenite.

3) Nucleation of cementite after ε - carbide goes back to solution. With further tempering, cementite gradually coarsens and in doing so, eventually spheroidizes to form a dispersion of spheroidal Fe_3C in ferrite.

The presence of chromium in the En 30A steel is expected to have a retarding effect on the growth of the cementite over a wide temperature range, e.g. 100-500°C. It is assumed that the chromium in solid solution in the tempered ferrite retards the solution of cementite by combining with available vacancies, and possibly by influencing the rate of carbon diffusion at the lower tempering temperatures.

The underlying reactions here are time dependent and occur either isothermally or on heating. Moreover, the three stages do not occur entirely consecutively; the third stage overlaps the second and both may partially overlap the first.

These descriptions of structural changes which occur during the three stages of tempering are supported by the electron microscope investigations of Trotter and McLean(85), and Lement et al(86,87).

In the present investigation no work has been done to separate the plastic deformation occurring during each stage of tempering; rather a general combined effect was investigated. Moreover, no work has been done in the temperature range (e.g. 500°C) where the replacement of cementite (formed at lower temperatures) with alloy carbides occurs.

The results on tempering may be interpreted in a similar way as for martensite formation(in the En 30A steel or the iron-20 wt% nickel alloy). The following mechanism is suggested for the increase in plasticity found during tempering:

It is known that the specific volumes of carbides are always higher than those of the matrix solid solutions of ferrite. Thus, as precipitation of carbides occurs, stresses may be produced which would lead to the generation of dislocations and the formation of interstitials. This feature may be related to the stressed region postulated in the macroplasticity approach, as discussed above.

The ductility during tempering might be controlled by several means. By controlling the mean free distance of dislocation movement, the size of the dislocation network exerts a strong influence on the total extension that occurs during tempering; thus changing the morphology of martensite might affect the ductility. A short time austenitizing treatment will produce very small martensite plates (1 to 10 microns in length) with rather irregular shape and internal twinning(84). Thus such a treatment might be expected to reduce the ductility, and in fact some effects were found with the very limited amount of work done in this particular field. Results on tempering of martensite produced by short time treatment are not presented here because there is still very much to be done to find the influence of the morphology of martensite upon the plasticity during tempering. Even by increasing the carbon content (which will change the substructure in the martensite grains of the steel) ductility during tempering might be affected.

The plasticity during the tempering of steel might be reduced if the growth of the carbides might be suppressed. If the surface energy of the carbide matrix interface is high the ratio of the concentration of solute elements adjacent to a particle to the equilibrium concentration will be large; consequently large concentration differences will be set up, diffusion will be facilitated and this will lead to rapid particle growth. The surface energy will depend on the nature of the precipitate. If it is partially coherent with the matrix the surface energy will be low; if it is incoherent with the matrix the interfacial energy will be high. It is possible to change the particle matrix interfacial energy by adding some alloying elements which may be soluble in either the carbide, the matrix or both. The addition of alloying elements may change the lattice parameters of the carbide and the matrix and the degree of mismatch across the interface will be lowered and consequently the surface energy will be decreased. This has been discussed by Smith and Nutting(111) and Baker and Nutting(112), in ferritic steels. The rapid growth of V_4C_3 (when the V/C ratio exceeded 4) was explained to be due to the excess vanadium in solid solution beyond that required to form V_4C_3 which changed the lattice parameters of the ferrite. This caused an increase in the degree of mismatch across the interface. However, by the addition of molybdenum, which is soluble in both ferrite and the V_4C_3, the mismatch and thus the growth rate of V_4C_3 may be decreased. So, to minimize the plasticity during tempering the carbide growth rate should be low, thus the free energy of the precipitated phase should be low.

The observed slight increase in hardness of samples stressed during tempering as opposed to samples tempered without stress may be assumed to be due to the carbides precipitated in the former case being finer and greater in amount. An increased amount of carbide precipitation is justified, because the dislocations would provide sites for carbide nucleation. Furthermore, for the tempering time employed the recovery of the deformation caused during the martensite transformation could not have been completed. However, longer times of tempering or higher tempering temperatures might be expected to result in lower hardness values for stress tempered specimens than plain tempered ones. This is because the dislocations may facilitate diffusion and this in turn will accelerate the growth of carbides. No metallographic work has been done on this subject and there is still much to be learnt about the influence of stress, tempering temperature and time upon the morphology of the precipitated carbide phase.

Conclusions:

It is claimed that the findings of the present work have thrown light on the underlying mechanism of transformation plasticity, by providing new experimental data and by confirming the findings of previous workers on other materials.

The following conclusions have been drawn from the results obtained in the present research:

1. During the transformation of austenite to martensite, plastic deformation occurs under stresses much less than the normal yield stress of austenite; e.g. for the En 30A steel, deformation was detected with a stress of only 1,000 lb/in^2.

2. In both the alloys investigated, for a given temperature of stress application, the amount of deformation increased with increase in applied stress, and a linear relation existed between extension and stress up to a stress corresponding to the austenite yield stress.

3. Transformation plasticity was detected only in a range of about 40°C below M_s in which range about 70% martensite formed. Within this temperature range,

the extent of transformation plasticity decreased with decreasing temperature of stress application; when the stresses were applied below this range, no transformation plasticity was detected.

4. Plastic deformation of the austenite above M_s reduced the amount of transformation plasticity that occurred during subsequent cooling through M_s.

5. Thermal stabilization of austenite in En 30A steel reduced the amount of transformation plasticity that occurred during cooling through the martensitic range.

6. The extent of transformation plasticity in both the alloys investigated varied with the conditions of austenitizing. The lowest values of plasticity were associated with short-time austenitizing treatments and the extent of plasticity increased considerably with increase in austenitizing temperature or time. Values of transformation plasticity following rapid heating during austenitizing were lower than those obtained when slow heating was used for austenitizing.
The iron-20 wt% nickel alloy, however, differed from the En 30A steel in that in the former alloy a change

in austenitizing temperature in the range 700°-1,000°C did not cause a change in the amount of transformation plasticity.

7. In both the alloys the yield stress of the austenite was found to vary with austenitizing conditions; relatively high strengths were associated with short-time treatments at relatively low temperatures, and lower strengths with higher austenitizing temperatures. Again the iron-20 wt% nickel alloy differed from the En 30A steel in that in the former alloy a change in austenitizing temperature in the range 700°-1,000°C did not cause a change in the austenite yield stress.

8. The results can be interpreted on the basis of macroplasticity considerations and are in good quantitative agreement with a relationship that predicts that the extent of transformation plasticity is directly proportional to the magnitude of the applied stress, and to the transformation volume change, and inversely proportional to the austenite yield stress.

9. Martensite plates were found to form with a preferred orientation in the initial stages of transformation, under some conditions of stress.

10. In the En 30A steel, plastic deformation occurred during tempering of martensite under stresses much less than the martensite yield stress, and for a given tempering temperature and time there was found to be a linear relationship existing between the extension and the magnitude of the applied stress.

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REFERENCES

1. Saveur, A., Iron Age, 1924, 113, 581.
2. Scheil, E., Z. Anorg. Chem., 1932, 207, 21.
3. Wassermann, G., Arch. Eisenhüttenw., 1933, 6, 347; 1937, 10, 321; 1937, 11, 89.
4. Wassermann, G. and Horbogen, E., Z. Metallk., 1956, 47, 427.
5. Bibring, H. and Sebileau, F., C. R. Acad. Sci., 1957, 244, 149.
6. Buckley, S. N., Hading, A. C. and Waldron, M. B., J. Institute of Metals, 1959, 87, 150.
7. Lehr, P., C. R. Acad. Sci. Paris, 1956, 242, 1172.
8. Lozinsky, M. G., Acta Met., 1961, 9, 689.
9. Stobo, J. J., J. Nucl. Mat., 1960, 2, 97.
10. Clark, C. L., and Russ, J. J., Trans. A.I.M.E., 1946, 167, 736.
11. Ihring, H. K., Iron Age, 1944, 153, 170.
12. Ihring, H. K., Trans A.I.M.E., 1946, 167, 749.
13. Saveur, A., Trans. A.I.M.E., 1946, 167, 749.
14. Gueussier, A. and Castro, Rev. Met., 1958, 55, 1023.

15. Brandsma, W. F. and Lips, E. M. H., Z. Metallk., 1936, 28, 381.

16. Greenwood, J. N. and Worner, H. K., J. Inst. Met., 1939, 64, 131.

17. Andrade, C., Nature, 1948, 162, 410.

18. Thornton, P. H. and Chan, R. W., J. Inst. Met., 1960-61, 89, 455.

19. Hardwick, D., Sellars, C. M. and Tegart, W. J. McG., J. Inst. Met., 1961-62, 90, 21.

20. Gifkins, R. C., J. Inst. Met., 1958-59, 87, 255.

21. Rohnhanau, W., Z. Metallk., 1932, 24, 127.

22. Schmid, V. E. and Wassermann, G., Z. Metallk., 1931, 23, 242/43.

23. Wassermann, G., Z. Metallk., 1932, 24, 127/31.

24. Sauveur, A., Tr. A.I.M.E., 1924, 70, 3.

25. Michael, G. E. and Izina, S. S., Acta Met., 1959, 7, 709.

26. Lagerberg, G. and Joseffson, A., Acta Met., 1955, 3, 236.

27. Cuff, F. B. and Grant, N. J., J. Inst. Met., 1959, 87, 278.

28. Guarniery, J. G. and Kanter, J. J., Tr. A.S.M., 1948, 40, 1147.

29. de Jong, M. and Rathenau, G. W., Acta Met., 1959, 7, 246.

30. Bibring, H. and Sebilleau, F., C. R. Acad. Sci. Paris, 1957, 244, 1496.

31. de Jong, M. and Rathenau, G. W., Acta Met., 1957, 5, 679.

32. Rathenau, G. W. and Boas, G., Métaux, Corrosion Indust., 1954, 29, 139.

33. de Jong, M. and Rathenau, G. W., Acta Met., 1961, 9, 714.

34. de Jong, M., Thesis, University of Amsterdam, 1960.

35. Porter, L. F. and Rosenthal, P. C., Acta Met., 1959, 7, 504.

36. Karskii, N. E. and Sobolov, T. I., Zavod. Lab., 1949, 15, 1355.

37. Kayushnikov, P. Ya., Advanced Practice of Production; Heat Treatment of Metals, Oborongiz, 1946, No 1.

38. Kayushnikov, P. Ya., Distortion of Steel in Quenching and Methods of Eliminating it, Moscow, Off. Sci. and Tech. Propaganda, 1957.

39. Goffard, J. W. and Wheeler, R. G., Hot Indentation Testing of Magnesium and Other Selected Material, Trans. Met. Soc. A.I.M.E., 1959, 215, 902.

40. Prensyakov, A. A. and Chervyakova, V. V., Izv. Akad. Nauk, SSSR, Otdel. Tekh. Nauk, 1960, 3, 92.

41. Prensyakov, A. A. and Chervyakova, V. V., Fiz. Met. i Metallovedenie, 1959, 8, No 1, 114.

42. Prensyakov, A. A. and Chervyakova, V. V., Proc. Inst. Nuclear Physics, Acad. Sci. Kazakh SSSR, 1959, 2, 3.

43. Prensyakov, A. A. and Chervyakova, V. V., Proc. Inst. Nuclear Physics, Acad. Sci. Kazakh SSSR, 1959, 2, 30.

44. Ibid., p. 101.

45. Prensyakov, A. A. and Starikova, G. V., Izv. Akad. Nauk SSSR, Otdel. Tekh. Nauk, 1959, 1, 75.

46. Prensyakov, A. A. and Dnysemaliyev, U. K., Proc. Inst. Nuclear Physics, Acad. Sci. Kazakh SSSR, 1959, 2, 74.

47. Prensyakov, A. A. and Starikova, G. V., Izv. Akad. Nauk SSSR, Otdel. Tekh. Nauk, 1960, 123.

48. Prensyakov, A. A. and Chervyakova, V. V., Kaz, SSSR, Fylym Akad. Khabarshysy, Vest. Akad. Nauk Kaz. SSSR, 1958, 12, 76-80.

49. Prensyakov, A. A. and Chervyakova, V. V., Izv. Akad. Nauk SSSR, Otdel. Tekh. Nauk, 1958, 3, 120.

50. Bockvar, A. A., Izv. Akad. Nauk SSSR, Otdel. Tekh. Nauk, 1947, 1364.

51. Osipov, K. A. and Miroshkina, E. M., Trudy Inst. Met., Akad. Nauk SSSR, 1958, 3, 160.

52. Konobeyevskii, S. T., Zh. Expt. Teor. Fiz., 1943, 13, No 6.

53. Osipov, K. A., Izv. Akad. Nauk SSSR, Otdel. Tekh. Nauk, 1949, No 9.

54. Bockvar, A. A., Various mechanisms of plasticity in metallic alloys, Izv. Akad. Nauk SSSR, Otdel. Tekh. Nauk, 1948, 649.

55. Guy, R. G. and Pavlick, J. E., Tran. A.I.M.E., 1961, 221, 802.

56. Bilby, B. A. and Pickering, F. B., Precipitation process in steels, spec. rep. No 54, Iron Steel Inst., London, 1959, 313.

57. Vorobev, V. G., Metalloved. i Obrabot. Met., 1958, 2, 11.

58. Boas, W., Physics of Metals and Alloys, Wiley, New York, 1949, 121.

59. Johnson, R. H. and Greenwood, G. W., N.P.L. Symposium No 15 HMSO, 1963, 205.

60. Johnson, R. H. and Greenwood, G. W., Nature, 1962, 195, 138.

61. Chang, L. C. and Read, T. A., Trans. A.I.M.E., 1959, 191, 47.

62. Burkart, M. W. and Read, T. A., Trans. A.I.M.E., 1953, 197, 1516.

63. Kochendörfer, A. and Otto, G., Arch. Eisenhüttenw., 1958, 29, 709.

64. Bhattacharyya, S. and Kehl, G. L., Trans. A.S.M., 1955, 47, 351.

65. Nobarro, F. R. N., Symposium on Internal Stresses in Metals and Alloys, Institute of Metals, London, 1948, 237.

66. Cottrell, A. H., Theory of Dislocations, Progress in Metal Physics, Interscience, New York, 1950, 1, Chapter 2.

67. Eshelby, J. D., Frank, F. C. and Nobarro, F. R. N., The Equilibrium of Linear Arrays of Dislocations, Phil. Mag., 1951, 42, 351.

68. Koehler, J. S., The Production of Large Tensile Stresses by Dislocations, Phys. Rev., 1952, 85, 480.

69. Troiano, A. R. and Greninger, A. B., Metal Progress, 1946, 50, 303.

70. Chevenord, P., "Mechanism of the Carbon Steels", Revue de Metallurgie, Mem, 1919, 16, 17-79.

71. Bain, E. C. and Davenport, E. S., Trans. A.I.M.E., 1930, 90, 117.

72. Oakley, D. B. and Oesterle, J. F., Trans. A.S.M., 1940, 28, 832-852.

73. Griffiths, W. T., Pfeil, L. B. and Allen, N. P., "Intermediate Transformation in Alloy Steels", British Iron and Steel Institute, Special Report, No 24, 369.

74. Parke, R. M. and Herzig, A. J., Metals and Alloys, 1940, 11, 6.

75. Wever, F. and Lange, H., Mitt. Kaiser-Wilhelm-Inst. Eisenforsch., Dusseldorf, 1932, 14, 71.

76. McReynolds, A. W., Jour. Appl. Phys., 1946, 17, 823.

77. Heal, H. T. and Mykura, H., Metal Treatment and Drop Forging, 1950, 17, No 61, 129.

78. Gilliam, E. and Cole, D. G., Iron and Steel, 1953, 26, 471.

79. Wells, M. G. H., "Thesis on the Effect of Applied Tensile Stress and of Austenitizing Conditions on the Formation of Martensite in Steels", University of London, 1961.

80. A.S.M., Metals Handbook, 1948.

81. Koistinen, D. P. and Marburger, R. E., Acta Met., 1959, 7, 59 (Letter).

82. Rolt, R. H., "Gauges and Fine Measurements", Macmillan and Co., London, 1929, 2.

83. Digges, T. G. and Rosenberg, S. J., Trans. A.S.M., 1943, 31.

84. Sastri, A. S., Ph.D. Thesis, University of London, 1964.

85. Trotter, J. and McLean, D.,J.Iron St.Inst.,1949,163,9.

86. Lement, B. S., Averbach, B. L. and Cohen, M., Trans. A.S.M., 1954, 46, 851.

87. Lement, B. S., Averbach, B. L. and Cohen, M., Trans. A.S.M., 1955, 47, 291.

88. Timoshenko, S., Strength of Materials, 1953, Part 2, Chapter 9.

89. Zener, C. and Hollmon, J. H., Trans. A.S.M., 1944, 33, 163.

90. Koster, W., Z. Metallk., 1948, 39, 1.

91. Kelly, P. M., Int. Union of Crystallography 5th International Congress and Symposia, 1960.

92. Kelly, P. M. and Nutting, J., Iron Steel Inst., 1961, 197, 199.

93. Edmonson, B. and Ko, T., Acta Met., 1954, 2, 235.

94. Kurdjumov, G. V., J. Iron Steel Inst., 1960, 195, 26-48.

95. Krauss, G., Jr., Acta Met., 1963, 11, 499.

96. Krauss, G., Jr. and Cohen, M., Trans. A.I.M.E., 1962, <u>224</u>, 1212-1221.

97. Edwards, C. A. and Pfeil, L. B., J. Iron Steel Inst., 1925, <u>112</u>, 79.

98. Edwards, C. A., Jones, H. N. and Walters, B., Ibid., 1939, <u>139</u>, 341.

99. Winlock, J. and Leiter, W. E., A.S.M., 1937, <u>25</u>, 163.

100. Petch, N. J., J. Iron Steel Inst., 1953, <u>174</u>, 25.

101. Stroh, A. N., Phil. Mag., 1955, <u>46</u>, No 7, 968-972.

102. McLean, D., Mechanical properties of metals, 1962, 112.

103. Holden, F. C., Ogden, H. R. and Jaffe, R. I., Second Interior Report Contract No DA-33-014-ORD-1397 Battelle Memorial Inst. to Watertawn Arsenal.

104. Porter, L. F. and Dienes, G. J., Trans. A.I.M.E., 1959, <u>215</u>, 854.

105. Kaufman, L. and Cohen, M., Progress in Metal physics, 1958, <u>7</u>, 165.

106. Morgon, E. R. and Ko, T., Acta Met., 1953, <u>1</u>, 36.

107. Wells, C., Botz, W. and Mehl, R. F., Trans. A.I.M.E., 1950, <u>188</u>, 553.

108. Whelan, M. J., Hirsch, P. B., Horne, R. W. and Bollmann, W., Proc. Roy. Soc., 1957, (A), <u>240</u>, 524.

109. Patel, J. R. and Cohen, M., Acta Met., 1953, <u>1</u>, 531.

110. Machlin, E. S. and Cohen, M., Trans. A.I.M.E., 1951, <u>191</u>, 1019.

111. Smith, E. and Nutting, J., J. Iron Steel Inst., 1957, <u>187</u>, 314.

112. Baker, R. G. and Nutting, J., J. Iron Steel Inst. Special Report No 64, 1959, 1-22.

113. Cullity, B. D., Elements of X-ray diffraction, 1956, 272.

APPENDIX I.

Determination of preferred orientation:

The following is a calculation of the preferred orientation of the martensite plates formed under a stress of 50,000 lb/in^2 for En 30A steel.

The first step is to index the Debye rings by calculating θ, the half angle between incident and reflected beam (113); thus by the use of equation

$$\sin^2\theta = \frac{\lambda^2}{4}\left(\frac{h^2 + k^2}{a^2} + \frac{l^2}{c^2}\right)$$

(113) the rings marked as 1, 2 and 3 were identified as (110), (200) and (211) reflections respectively (Fig.46). The angle α is then measured from a vertical line through the centre of the film to the centre of each strong Debye-arc. The average values of these angles are given below, together with the calculated values of ρ (113).

(ρ is the angle between the normal of the reflecting plane and fibre axis)

Debye Circle	hkl	α	θ	ρ
Inner	110	55°	10.2°	56°
Outer	200	90°	14.6°	90°

The normals to the (110) and (200) planes therefore make angles of 56° and 90° respectively, with the fibre axis. The indices [u v w] of this axis can be determined by making use of the interplanar angles, thus the nearest direction to [u v w] with simple indices would be [320], since the angle between ⟨110⟩ and ⟨320⟩ is 54° and that between ⟨200⟩ and ⟨320⟩ is 90°, and these values agree with the values of ρ given above within experimental error. The fibre axis of the En 30A steel transformed to martensite under stress is therefore [320]. There is some scatter of the [320] direction about the wire axis, however, inasmuch as the reflections on the film are short arcs rather than sharp spots.

APPENDIX II.

If it is assumed that the deformation during transformation obeys Lévy von Mises criterion(59,60) then

$$\bar{\varepsilon} = \sqrt{\frac{2}{3}\left(\varepsilon_1^2 + \varepsilon_2^2 + \varepsilon_3^2\right)}$$

where $\bar{\varepsilon}$ = equivalent or effective strain

$\varepsilon_1, \varepsilon_2, \varepsilon_3$ = are principal strains

Therefore

$$\bar{\varepsilon} = \sqrt{\frac{2}{3}\left\{\left[\varepsilon - \left(\frac{\Delta V}{V}\right)_x\right]^2 + \left[-\frac{\varepsilon}{2} - \left(\frac{\Delta V}{V}\right)_y\right]^2 + \left[-\frac{\varepsilon}{2} - \left(\frac{\Delta V}{V}\right)_z\right]^2\right\}}$$

where $\left(\frac{\Delta V}{v}\right)_x$, $\left(\frac{\Delta V}{V}\right)_y$, $\left(\frac{\Delta V}{V}\right)_z$ are the principal strains due to the fractional difference between specific volumes

ε, $-\frac{\varepsilon}{2}$, $-\frac{\varepsilon}{2}$ are the principal strain components due to the resultant deformation.

The material yields when $\bar{\varepsilon} = \frac{2}{3}\lambda Y$

where λ = differential parameter

Y = yield stress of the deforming phase

Thus

$$\sqrt{\frac{2}{3}\left[\left[\varepsilon - \left(\frac{\Delta V}{V}\right)_x\right]^2 + \left[-\frac{\varepsilon}{2} - \left(\frac{\Delta V}{V}\right)_y\right]^2 + \left[-\frac{\varepsilon}{2} - \left(\frac{\Delta V}{V}\right)_z\right]^2\right]} = \frac{2}{3}\lambda Y$$

expanding the above equation

$$\frac{3}{2}\varepsilon^2 + \left[\left(\frac{\Delta V}{V}\right)_x + \left(\frac{\Delta V}{V}\right)_y - 2\left(\frac{\Delta V}{V}\right)_z\right]\varepsilon$$

$$+ \left(\frac{\Delta V}{V}\right)_x^2 + \left(\frac{\Delta V}{V}\right)_y^2 + \left(\frac{\Delta V}{V}\right)_z^2 = \frac{2}{3}\lambda^2\gamma^2$$

From ε considerably less than $\frac{\Delta V}{V}$, the above equation becomes

$$\left(\frac{\Delta V}{V}\right)_x^2 + \left(\frac{\Delta V}{V}\right)_y^2 + \left(\frac{\Delta V}{V}\right)_z^2 \sim \frac{2}{3}\lambda^2\gamma^2 \quad \ldots\ldots\ldots \quad (1)$$

and using the following equation

$$\sigma = \frac{3}{2\lambda}\left[\varepsilon - \left(\frac{\Delta V}{V}\right)_z\right]$$

where σ = the applied stress

with equation 1, it follows

$$\varepsilon \approx \sqrt{\frac{2}{3}\left(\frac{\sigma}{\gamma}\right)^2\Big[\left(\frac{\Delta V}{V}\right)_x^2 + \left(\frac{\Delta V}{V}\right)_y^2 + \left(\frac{\Delta V}{V}\right)_z^2} + \left(\frac{\Delta V}{V}\right)_z$$

the resolved components of the transformation volume change are

$$\left(\frac{\Delta V}{V}\right)_x = \frac{2}{3}\left(\frac{\Delta V}{V}\right)$$

$$\left(\frac{\Delta V}{V}\right)_y = -\frac{1}{3}\left(\frac{\Delta V}{V}\right)$$

$$\left(\frac{\Delta V}{V}\right)_z = -\frac{1}{3}\left(\frac{\Delta V}{V}\right)$$

substituting these values into equation 2

$$\varepsilon \cong \sqrt{\frac{2}{3}\left(\frac{\sigma}{Y}\right)^2 \left\{\left[\frac{2}{3}\left(\frac{\Delta V}{V}\right)\right]^2 + \left[-\frac{1}{3}\left(\frac{\Delta V}{V}\right)\right]^2 + \left[-\frac{1}{3}\left(\frac{\Delta V}{V}\right)\right]^2\right\}} + \left(\frac{\Delta V}{V}\right)_z$$

or

$$\varepsilon = \frac{2}{3}\,\frac{\sigma}{Y}\left(\frac{\Delta V}{V}\right) + \left(\frac{\Delta V}{V}\right)_z \quad \text{thus}$$

$$\varepsilon = A\,\frac{\sigma}{Y}\left(\frac{\Delta V}{V}\right) + B\left(\frac{\Delta V}{V}\right) \text{ is obtained.}$$

APPENDIX III.

Metallographic observations suggested that the iron-20 wt% nickel alloy exhibited some banding in the structure. Electron probe microanalysis, conducted previously, indicated no inhomogeneity in the material but recently the microanalysis results communicated by the International Nickel Company (Mond) Limited show that banded structure is the result of incomplete mixing and homogenizing in the original powder metallurgy process. The nickel content in this alloy has been shown to vary from 19.5% to 20.5% and the band width was about 5 microns. It is difficult to estimate the effect of the banding on the results obtained in this investigation but it seems very unlikely that the main conclusions reached would be altered. This is so because the temperature vs. length change curve showed a sharp discontinuity at M_s and because of the excellent reproducibility shown in the M_s temperature and transformation plasticity results even when various cycles of austenitizing treatments were done in succession. The uncertainty in grain size results, mentioned in the present investigation, might be affected by this banding.