THE SPRING RESEARCH AND MANUFACTURERS' ASSOCIATION

# ON THE MECHANISM RESPONSIBLE FOR THE CHANGES IN ELASTIC AND RELAXATION PROPERTIES WITH STRESS, STRAIN AND TEMPERATURE VARIATIONS IN SPRING MATERIALS

by

P. Gray, B.Sc.

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## SUMMARY AND CONCLUSIONS


From considerations of possible residual stress mechanisms and mechanical property and relaxation data available for spring and other materials, arguments are presented for the mechanisms which may be responsible for the often beneficial changes in elastic and relaxation properties in springs after low temperature heat treatment or hot stressing.

The conclusions drawn are that the data can be most satisfactorily explained by the presence of mobile dislocations which interact with fine precipitates in a manner analogous to dispersion hardening, and which can further interact and ultimately break away from the precipitate by stress-assisted thermal activation to an extent dependent on the temperature of the hot stressing or low temperature heat treatment. By this means the behaviour of elastic and other mechanical properties, together with relaxation behaviour, may be satisfactorily explained.

It has been shown that only metals and alloys with sufficient impurities to provide the necessary density of fine precipitates display improvements in elastic and relaxation properties obtained from suitable stress, strain and temperature changes.

A number of recommendations for future experiments are made in the light of this work.

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## 1. INTRODUCTION

After consideration of possible residual stress mechanisms in the various configurations that spring elements may take, a survey is made of the data on both spring material and other relevant materials, in respect of mechanical and relaxation properties and their behaviour with strain, hot and cold stressing, and low temperature heat treatment (LTHT).

From this background information, possible mechanics for the behaviour observed are put forward and critically assessed on the basis of the known data.

Conclusions are drawn as to the most likely mechanism and recommendations made to test these ideas further.

## 2. APPLICATION OF STRESS AND THE EFFECT OF HEATING IN METALLIC SPRINGS

In most springs - especially coil springs - it is difficult in practice to distinguish completely between the mechanical property changes induced by deformation and those induced by heat treatment. Spring manufacture entails a series of mechanical and heat treatment operations, for example: cold drawing of wire - coiling - low temperature heat treatment - prestressing etc., in which, at some stage, residual stresses of various kinds are introduced. Thus any intermediate low temperature heat treatment is almost invariably, at the present time, conducted in the presence of significant residual stress in some part of the spring, even if no external stress is applied at the time of the heat treatment. Low temperature heat treatment (LTHT) is generally applied at temperatures below the recovery range of the alloy, which means that most of the residual stresses are retained. Hence, any effects measured as a result of LTHT must be explainable within the framework of some theory of hot stressing or vice-versa. Moreover, the effect on mechanical properties should be qualitatively similar, once the directionality of applied or residual stresses is accounted for, even though the effects of LTHT are likely to be quantitatively less, the residual stresses being perhaps lower or less extensive than in the case of externally applied stresses. This latter point is illustrated by some very early work on hot prestressing by the SRAMA (then The Coil Spring Federation) by Barson and Pope[1]. In Figs 1 and 2, which are taken from this paper, it is evident that hot torsional stressing of a hardened and tempered steel spring wire has an effect on proof stress and flow stress that is similar to stressing separately at room temperature and then applying the LTHT. Hot stressing is, however, somewhat more efficient, particularly in improving the elastic range.

It must be concluded that any adequate explanation for the behaviour of springs as a result of hot and cold prestressing should also encompass the behaviour reported on various spring (and other) materials after LTHT at different stages of production.

## 3. RESIDUAL STRESS

As mentioned previously, directional residual stress can be introduced during the manufacture of springs in a variety of ways: by cold drawing (or other forms of cold work); straightening; coiling; by heat treatment; by prestressing; and by shot peening. Fig 3 shows the likely macroscopic residual stress distributions in some of these cases.

In order to outline further effects of these residual stresses, consideration will be given to their generation in bending or torsional stressing.

## 3.1 Idealised Residual Stress Situations

A simplified version of residual stress generation may be illustrated using the models shown in Fig 4 (a) and (b). The simplest examples of residual stress would occur in Fig 4(a), in which a beam is considered to be completely elastic up to any stress level except for a thin surface layer on the top and bottom of the beam, which, although having the same elastic modulus as the rest of the beam, has both a tensile and a compressive yield point at a certain stress beyond which plastic flow will occur within the two outer layers. An analogous model for torsion can be visualised, as in Fig 4(b), as a perfectly elastic bar with two thin surface strips wound helically round the surface of the bar at 45$^{\circ}$ to the long axis and at 90$^{\circ}$ to each other.

If a load is applied downwards to the end of the beam in Fig 4 (a), then the top layer will be extended in tension and the bottom layer compressed in its longest dimension. An analogous effect occurs in Fig 4(b) by twisting. Depending on the direction of twist, one of the surface strips will be extended, whilst a contraction will be induced in the other if the composite is to remain whole. Thus, bending and torsion can be regarded as analogous in some ways.

One way in which, it could be argued, there is a major difference, is that in the case of torsion, the surface layer carries both the tensile strain in one direction and the compressive strain in the perpendicular direction, as is represented by the interactions of the two strips in Fig 4 and also shown in Fig 3(c). In bending, however, the tensile layer is completely separate from the compressive layer. Nevertheless, even in the bending model of Fig. 4(b), where for example, the upper layer is effectively acted on by a tensile force in the horizontal direction which (assuming plain strain) produces a linear strain $E_x$, in that direction, a contraction is also produced in the vertical direction, Ey, i.e. through the thickness of the layer where $E_y = -\nu E_x$ (Poisson's ratio, $\nu$, is 0.25 for a perfectly isotropic elastic material, but for most metals the value is closer to 0.33).

Moreover, as practical materials are polycrystalline, complex gradients of expansion and contraction will be set up within each grain, dependent on the individual orientation and the occurrence of microstrains at low stresses in favourably oriented grains[2-4]. This will also require redistribution of elastic strain to maintain structural continuity.

It is evident then that, on a microscopic scale, the differences in stress-strain behaviour of bending and torsional elements are not so different in practice. Thus to illustrate subsequent considerations, the bending model of Fig 4(a) will tend to be used here, although the argument will invariably also apply to torsional straining.

Fig 5 illustrates the bending of the beam in Fig 4(a), assuming elastic behaviour up to the stress at the yield point y after which plastic flow can occur for no further increase in stress up to the necessary strain value at B. On removal of the load, the stress-strain curve will follow the line from B through C and the elastic nature of the majority of the beam will ensure that the beam will return to an almost straight position of zero strain at point D on Fig 5(b) which, within this surface layer, corresponds to a compressive stress present because the layer has been induced to extend by plastic flow and has now to revert elastically to its original length.

A real metallic component would normally experience work hardening which would narrow the resulting stress-strain loop by an amount depending on the intensity of work hardening as shown in Fig 6. Nevertheless, compressive residual stress in a flat strip, after loading to produce sufficient plastic flow, can still be large and, given sufficient strain, can be as high as the compressive yield stress of the material.

The bottom layer in Fig 5 will behave in the same way as the upper layer, except that the applied stress will be compressive, y being the compressive yield stress and the residual stress at D in Fig 5(b) being a tensile one. Thus the set of graphs given in Fig 5 for the upper layer will be the same for an equivalent point in the lower elastic-plastic layer if they are inverted, so that the tensile direction for the upper layer is the compression direction for the lower layer and vice versa.

In addition, the fact that the central elastic portion exerts a force on the surface layer means that an equal and opposite force is exerted on the central elastic region. That is, in the deflected condition, Fig 5(a), elastic energy is stored in the elastic central region, part of the energy being lost in the surface layer by plastic flow. As the flexure is reduced the stored elastic energy in the central region is also reduced. However, at some point, storage of elastic energy in the upper surface is required to provide the compressive residual stress. Reduction of flexure will continue until the energy required for the increasing compressive residual stress in the surface layer balances the reducing stored elastic energy in the upper half of the central region, thus leaving a compressive residual stress in the surface layer and a tensile residual stress in the upper half of the central region (and vice versa in the lower half of the bar). In a real bar, where the plastic flow is not necessarily confined to one surface layer but can be thought of as the sum (or integral) of the effect of a number of such layers, then, as may be seen from Fig. 3(b), the resulting internal stress distribution is qualitatively similar to that described above.

Inside the beam of Fig. 5(b) there are therefore balanced permanent stresses, which imply a permanent macroscopic strain of flexure of the beam. In this case only a thin surface layer contains all the compressive residual stress; therefore, the force corresponding to the residual stress is small because of the small cross sectional area of the surface strip. The residual strain is therefore very near to zero in the graph of Fig 5(b), with negligible permanent flexure of the beam. In a real beam, plastic flow can occur over a substantial proportion of the cross-section and hence some significant permanent flexure or set of the beam occurs.

If the surface layer of the beam in Fig 5(b) had the substantial thickness as drawn, then the beam would be permanently flexed on removal of the load; on the graph this would correspond to movement from point B down towards point D, coming to rest somewhere on the line between points C and D. The larger the cross-sectional area of the plastic layer for a given amount of plastic flow, then the nearer to C will be the point of rest. It can be seen that the larger the permanent set, the smaller the residual stress for a given amount of plastic strain in the surface layer(s).

### 3.2 The Bauschinger Effect[5]

In practice, the distribution and extent of residual stresses in a given component are not usually known. In order to study the mechanical properties of, for example, the beam in Fig 5(b), the load required to bend the beam through given amounts of deflection can easily be measured. To to this, axes are first drawn for zero applied stress and strain (flexure). However it is evident here that the axes for applied stress-strain are different from the total stress-strain axes if the history of the specimen were known. In particular, the strain axis in the total stress-strain co-ordinate system, i.e. zz, is displaced to z'z' when only the applied stress is considered; if the residual stress condition of the beam were unknown, this would be the only possible starting point.

Thus in the (xx,z'z') co-ordinate system it appears that if the beam again is flexed downwards from zero applied stress at point D, then the material will have an unnaturally high yield stress at point B (compared with stress required from 0-Y) or if loaded from point D upwards in the opposite direction (Fig 5c), then the material will have an apparently low yield stress at y'.

This phenomenon is similar to the Bauschinger Effect which occurs in tensile testing; here, however, because of the large range of ~~residual~~ and applied stresses through ~~the~~ section of a flexurally stressed beam, the effect is likely to be larger, and, as given above, more easy to explain.

As shown schematically in Fig 7, a similar effect to bending occurs in tensile/compressive loading, where the initial yield stress of the material in tension is A. If the same material were tested in compression, the yield strength would be approximately the same. Again, consider that a new specimen is loaded in tension past the tensile yield stress to C along the path OAC. If the specimen is then unloaded, it will follow the path CD. If a now compressive stress is applied, plastic flow will begin at the stress corresponding to point E, which is lower than the original compressive yield stress of the material. A similar effect is found if, initially, a compressive stress is applied, followed by a tensile one. Fig 8 is an example of the tensile and compressive behaviour of a low carbon steel from a recent paper[(6)] showing an obvious Bauschinger Effect.

This depression of the elastic and flow properties with reversed loading could again be described in similar terms to the bending model in Fig 5, if it is assumed, as in Fig 7, that the specimen is largely elastic through most of the body up to any applied stress level except for a region, or regions, which are elastic with the same modulus as the rest of the specimen up to a yield stress greater than that at which plastic flow can occur (region B Fig 7). In an analogous way to Fig 5, residual stresses can be induced in the material by loading above the stress corresponding to A (Fig 7). Stress-strain behaviour essentially similar to that in the bending model of Fig 5 is obtained with a similar displacement of yield points, dependent on the internal residual stresses as a consequence of the internal plastic deformation.

If the Bauschinger Effect is to be explained by this method, regions of different yield stress within the body of the specimens are required to take the place of the plastic layers in bending which are easily explained here as a consequence of the gross variation of applied stress from zero at the centre of the beam to a maximum at the surface.

In tensile and compressive loading the macroscopic applied stress is uniform and it is necessary to consider much smaller regions, that is grains or even smaller volumes. It is possible that certain crystals (grains) may deform plastically more than others, depending on the orientation of the natural slip direction of the crystal lattice with the direction of the applied stress. When the load (say, a tensile one) is released, these crystals compress elastically the less strained crystals and thus induce earlier yielding upon loading in compression.

## 3.3 Dislocation Mechanisms

Whilst the difference in deformation behaviour of grains may account for part of the Bauschinger Effect, this phenomenon also occurs with single crystals[8], as shown in Fig 9. This effect can only be explained in terms of line defects or dislocations in metals and alloys, which inevitably links the Bauschinger Effect at this level with work-hardening. One of the earliest dislocation concepts put forward to explain strain-(or work-) hardening was the idea that dislocations pile up on slip planes at barriers in the crystal. The pile-ups produce a back stress which opposes the applied stress on the slip plane. The existence of a back stress was demonstrated experimentally, as shown in Fig 9, by shear tests on single crystals of zinc. Zinc crystals are ideal for crystal-plasticity experiments because they slip only on the crystal plane having the highest atomic density and hence complications due to mobile dislocations in other planes are easily avoided. It should be noted that on reloading in the direction opposite to the original slip direction, the crystal yields at a lower shear stress than when it was first loaded. This occurs because the back stress developed as a result of dislocations piling up at barriers during the first loading cycle aids dislocation movement when the direction of slip is reversed. Furthermore, when the slip direction is reversed, dislocations of opposite sign could be created at the same source that produced the dislocations responsible for strain in the first slip direction. Since dislocations of opposite sign attract and annihilate each other, the net effect would be a further softening of the lattice.

While all metals exhibit a Bauschinger Effect, it may not always be of the magnitude shown for zinc crystals. Moreover, Orowan[9] has pointed out that, if the Bauschinger Effect is due solely to the effect of back stresses, the flow curve after reversal of strain ought always to be softer than the flow curve for the original direction of strain. However, not all metals show a permanent softening after strain reversal. Therefore, Orowan considers that the Bauschinger Effect can be explained by the same mechanism which he proposed for dispersion hardening (see Fig 10). Obstacles to dislocation motion are considered to be other dislocations, inclusions, precipitate particles, etc. The stress required to move a dislocation through these obstacles is inversely proportional to the distance between the obstacles. For a given shear stress, a dislocation line will move over the slip plane until it meets a row of obstacles that are strong enough to resist shearing and close enough to resist the dislocation loop from squeezing between them. When the load is removed, the dislocation line will not move appreciably unless there are very high back stresses. However, when the direction of loading is reversed, the dislocation line can move an appreciable distance at a low shear stress because the obstacles to the rear of the dislocation are not so strong or so closely packed as those immediately in front of the dislocation.

Microscopic precipitate particles and foreign atoms can serve as barriers but for pure single crstals, the alternative barriers are other dislocations intersecting the active slip plane.

The mobile dislocation can cut through the intersecting dislocation provided that it has sufficient enegery to form a "jog" or offset in the dislocation line (Fig 10); if not, the dislocation segment may be joined together to form an attractive junction. A dislocation requires a specific additional energy to break an attractive junction and to form a jog in either the mobile or the intersecting dislocations, or both, depending on the types and orientations of the dislocation involved. Dislocations can be categorised into two types, namely "edge" and "screw" dislocations. Jogs in pure edge dislocations do not affect the subsequent motion of the dislocation: jogs in screw dislocations, however, impair the motion of the dislocations to the extent that movement of the jog requires thermal activation and consequently the movement of the screw dislocation will become temperature dependent[10].

Considerations have up to this point been confined to the residual stress behaviour in torsion, bending and uniaxial tension and compression. However, it may be seen from Fig 3 that residual compressive stresses are introduced into the surfaces of components by such processes as shot peening. It is to be expected that these compressive stresses and the subsequent behaviour of the dislocations generated by these processes will produce the same effects described above. The generation of vast quantities of dislocations by cold working and their interaction are responsible for the substantial increase in mechanical properties in most commercial metals and alloys. It has been shown[7] that, for most of the common metallic materials, with even as low as 3.5% cold work, the dislocations tend to cluster into rather well defined tangles which are arranged into the walls of cells, the cells themselves containing relatively few dislocations. The evidence indicates that there exists a relatively long range back stress arising from the dislocation in the cell walls[7]. These stresses oppose and may almost balance the applied stresses, thus reducing the net force on a moving dislocation, thereby accounting for the work strengthening. This long range stress thus takes over the role formerly applied to dislocation pile-ups as discussed previously. In this case the structural configuration and the hardening associated with it can only be removed by heating in the recovery region, thus converting the cells into subgrains and ultimately new grains with a simultaneous reduction in the internal back stresses and hardening associated with the cell structure.

## 3.4 Types of Residual Stress

To summarise, therefore, there are at least five possible mechanisms, distributed over regions of the metallic solid, ranging from the macroscopic to the microscopic scale, which may contribute to the depression of elastic properties of the body as a whole after some form of plastic deformation.

These are, in order of the "size" of the volume occupied by the internally strained region responsible for the residual stress, from the largest to the smallest:

1. Residual macro-stress - introduced by inhomogenous applied stress, i.e. flexure, torsion, cold work, shot peening, heat treatment etc.

2. Residual in-grain stresses - introduced by preferential plastic straining within grains with preferentially oriented slip systems formed by uniaxial tensile and compressive stress or any applied stress.

3. Back stresses - from pile-ups of dislocations against any obstacles to dislocation movement.

4. Back stresses - from cell walls in cold worked metals and alloys.

5. Asymmetry in dislocation motion due to prior dislocation interactions[9].

## 3.5 Evidence for Reduction in Elastic Properties after Various Methods of Plastic Straining

The assymetry in elastic properties after prestressing in bending or torsion is very well known in the spring industry. Fig 11 shows the effect for reversed torsion of hard drawn carbon steel wire[11]. It can be seen that subsequent repeated torsional cycling removes the depressive effect on the elastic limit by the additional work hardening induced. Also well known are the problems encountered with the reduction of elastic properties with straightening and coiling wire. Wire drawing and cold work generally can also be shown to reduce the elastic and proof stress properties, although this is usually greatly masked by the increase in general strength by work hardening. A particular example is given in Fig 12 for a stable austenitic stainless[12] steel; this shows a definite depression of the proportional limit, and to a lesser extent the proof stress, by application of limited amounts of cold work. These properties increase again with greater amounts of cold work as a result of increasing work hardening effects. One assumes here, and in general, that residual stresses still exert a depressive effect on elastic properties at high degrees of cold work but are masked by the general increase in strength properties by work hardening.

## 4. REVIEW OF EXISTING DATA ON PRESTRESSING, LTHT AND HOT STRESSING OF SPRING AND OTHER MATERIALS

Using as a background the discussions of Section 3, which briefly outlined the roles of various mechanisms responsible for the modification of elastic properties resulting from the various kinds of plastic deformation encountered in spring-making practice, consideration will be given to possible alternative explanations for the behaviour, based on the available data, of different metallic spring materials in various pre-stress and heat treatment conditions. Other metals and alloys not normally considered as spring materials will be considered where this adds support to the other data.

### 4.1 Stress-Strain Behaviour after Torsional Prestressing and Testing in Both Directions

Before reviewing the data generally it will be useful to consider briefly work done some years ago by Pope and Andrew[13]. The effects of prestressing in torsion of hardened and tempered carbon spring steel are shown in Fig 13. The graphs show the 1% proof stress and proportional limits, after twisting to produce a series of levels of prestress, represented by permanent shear strain measured at the surface, in both the directions of twist and the reverse direction, and both before and after LTHT.

It is evident that the total (positive plus negative) range of 1% proof strength is not just displaced in the direction of prestressing but the range itself is reduced by about 25%. LTHT restores the proof strength range for all levels of strain and also reduces somewhat the displacement of the range, since the proof strength is increased more in the reverse direction than in the strained directions. The authors also calculated the amount of plastic strain induced in the opposite direction by the residual stress as the load was removed. This effect is expected to reduce the residual stress which would have otherwise been introduced.

If the LTHT has been applied whilst still under load (i.e. hot stressed), then further recovery of elastic properties, leading to less reverse plastic strain, and therefore higher residual stresses and higher elastic properties in the direction of straining, would have occurred. This was shown to be the case in a later paper[1] and is demonstrated in Figs 1 and 2.

Fig 13 shows that the proportional limit is increased by 33% in the prestressed direction and, according to Pope and Andrew, by 260% in the reverse direction - in terms of actual torque values, however, the proportional limit is increased by the same amount, about 40 Nm, in both directions. This indicates that recovery of elastic properties has taken place within the stress range displaced by the Bauschinger Effects, but the displacement in properties and, therefore the bulk of the internal residual stresses, remains.

The bulk of the internal residual stresses can thus only be removed by recovery which also leads to the loss of work hardening and therefore a general softening of the material. Another, separate mechanism must therefore be involved in the changes in elastic properties with LTHT to explain why the elastic range is depressed with straining but then recovers with LTHT (or hot stressing), the bulk of the internal stress generated at the same time during straining being a purely additive effect.

## 4.2 Effect of Stress, Strain and Temperature on Elastic Properties

The information in Fig 14, which shows the proportional limit in torsion of wires of patented cold drawn carbon spring steel[14, 15], cold worked stainless steel[16] and brass[17], is representative of the behaviour to be expected after a prior LTHT at the temperatures indicated. It can be seen that there is a very distinctive peak in the carbon steel wires at 200-250°C. Brass has a less marked peak but still exhibits a relatively large uplift in elastic properties at about 250°C. The cold worked stainless steel shows a steady increase in proportional limit up to 350-450°C, followed by a sudden drop in elastic properties at higher temperatures. Fig 15 indicates that similar effects to patented cold drawn carbon steel, in respect of improved proof properties[18], can be expected after cold prestressing in torsion followed by LTHT for hardened and tempered silicon manganese steel (250A53)[19].

A wider view can be taken, enabling the comparisons to be extended to include metals and alloys other than spring materials if tensile elastic properties are considered. Fig. 16 gives examples of the elastic properties in tension of patented hard drawn carbon steel[20], cold worked stainless steels[12, 21] and brass[17]. Comparison of the tensile data given in Fig 16 with the torsional data of the previous two figures shows that the behaviour is generally similar, the peak for each material being in the same position. A striking feature common to Figs 14-16 is that, although all the materials reported respond well to LTHT, the carbon steels respond markedly better than either the stainless steels or 65/35 brass.

The elastic properties of patented cold drawn carbon steel wire may be raised even more if the wire is stressed to a level which will give 1-2% extension at the same temperatures used for LTHT. Fig 17 shows that by this means the elastic properties can be raised in a 5 mm diameter wire[22] to the same level as a similar material of 2.65 mm diameter, given the same LTHT without stressing.

Investigations on iron containing only small quantities of impurities such as carbon and nitrogen have shown that the increase in yield stress that occurs with LTHT after prior deformation (Fig. 18(a)) does not occur if the iron is first annealed in hydrogen[23], as shown by Fig. 18(b). Whether the absence of a peak in Fig. 18(b) is due to the reduction of the carbon content from 0.0019% to 0.0013% or by the hydrogen treatment as suggested by Smith et al[23], or by some other cause, is debatable.

Further examples of the effects of elevated temperatures on the mechanical properties of carbon steels (this time in the as hot rolled condition), whether deformation precedes heat treatment or both are carried out simultaneously, are given in Fig. 19, which shows these effects at three levels of carbon content[24], from 0.09 - 0.77%. It can be seen that the maximum increase in proof stress which occurs at 250 - 300$^{o}$C for a prior reduction of area of 12 - 15% followed by an LTHT, is only small at the 0.09% carbon level and becomes progressively larger with higher carbon content. If, however, the deformation is performed at temperature, the increase in proof stress is great at all temperatures, with a shift of the peak to higher temperature at the highest carbon content. At 0.77% carbon, the maximum proof stress occurs at 350$^{o}$C. A similar study of two levels of nitrogen content in carbon steel as illustrated in Fig. 20, appears to indicate that nitrogen has some effect in increasing the yield stress generally; however, the peaks, where present, are much lower in these figures, which may be a feature of the increased deformation, 18 - 30% reduction in area, of these steels.

Thus Figs. 18 -20 illustrate that the large elevation in mechanical properties with straining and LTHT or hot-straining is a general phenomenon with carbon steels. The next two figures, Figs. 21 and 22, show the effects of hot stressing on 18/9 stainless steel and an aluminium alloy[25] which, together with the data on a wide range of other materials in Table I, again shows that improvements in elastic and proof stress properties by LTHT after straining, and even more by hot stressing, are possible in a wide range of materials and alloys

4.3 Effect of Stress, Strain and Temperature on Subsequent Relaxation

Where the operating temperature of springs is above ambient, typically for IC engine valve springs, the minimisation of loss of load or deflection with service life is of considerable importance. There is evidence to suggest that the phenomena which are responsible for improvement of elastic properties after deformation and LTHT or hot stressing or straining may also produce significant improvements in relaxation behaviour[22]. Obviously, hot prestressing or LTHT (in the presence of residual stresses) at a similar temperature to the service temperature of the spring, will improve subsequent relaxation properties, since the early part of the relaxation will have occurred during processing. Moreover, since a large proportion of the relaxation occurs in these early stages the subsequent amount of relaxation occurring in service will be reduced correspondingly. Thus, the effects noted for hot prestressing in the lower temperature range, as reported for example in SRAMA Report 248 by Heyes[26], may be exclusively accounted for by the above explanation.

However, the more dramatic reductions in relaxation induced by stressing at temperatures much higher than the temperature of the subsequent relaxation tests reported in SRAMA Reports 215 and 234 by Gray[27,28] for Cr-V and Cr-Si wire and earlier by Evans and Bhattacharya[22] for hot stressed patented cold drawn wire, require another explanation. Fig. 23 shows the relaxation behaviour for shot peened Cr-V and Cr-Si steel springs at 150$^{\circ}$C after hot prestressing at temperatures up to 400$^{\circ}$C from ref.(28). It can be seen that a sharp reduction in relaxation occurs, with a maximum in the change of relaxation at a hot prestressing temperature of about 200$^{\circ}$C. Fig. 23 also shows for Cr-V steel the effect on relaxation of prior cold prestressing and LTHT at the indicated temperature[28], which shows again that LTHT has a similar effect to hot prestressing. Graves[29] has compiled data on the effect of LTHT on relaxation for a number of materials, from which it is evident that many steels show a sharp reduction in relaxation with LTHT. Carbon and low alloy steels appear to show more response to LTHT than some other materials such as stainless steel.

The data presented in Fig. 24 are taken from a recent SRAMA report, No. 266[20], concerned with patented cold drawn carbon steel spring wire; some of Bird's[20] data are replotted on the same graph for comparison. Maximum tensile stress and proof stress data, together with a relaxation curve and the curve for "wind up" of the springs, are given. Wind up is an extra coiling of the springs due to removal of the residual stresses which originated during cold coiling of the spring. It is re-plotted here in the reverse direction to emphasise the similarity between the change in this effect and the reduction of both the proof stress and the relaxation in the same temperature range (200-350$^{o}$C).

The alignment of these curves suggests that these effects may be motivated by the same basic metallurgical changes. Evans and Bhattacharya[22] found an almost identically shaped relaxation curve for similar (patented cold drawn carbon steel) wire in room temperature tensile loading of the wire after hot stressing in tension. Their curve is shown in Fig. 25. They also found similar effects for conductor purity aluminium wires but found that improvements in relaxation with hot stressing were less marked and almost absent in pure iron and OHFC copper.

## 5. EXPLANATIONS FOR THE EFFECTS OF STRESS, STRAIN AND TEMPERATURE ON THE MECHANICAL AND RELAXATION BEHAVIOUR OF SPRING (AND OTHER) MATERIALS

### 5.1 Dislocation-Precipitate Interaction and Elastic Properties

It is possible to consider as the simplest explanations for the increase in elastic properties with heat in a strained or prestrained metallic material mechanisms based solely on dislocations and their mutual interaction or interaction with other obstacles such as precipitate particles.

Part of the Bauschinger Effect may be explained in terms of the dispersion hardening model of Orowan[9], summarised towards the end of Section 3 and also in Fig. 10.

In this case it is more difficult for the dislocation segments to move in the prestrained direction than in the reverse direction. Before prestraining, the dislocation represented by Line 1 in Fig. 10(c) has the same stress strain behaviour in whichever direction the stress is applied. However, after prestraining to the right in the diagram, the dislocation will be in position 2. When the stress-strain determination is made, therefore, the axes are xx, y'y' rather than xx, yy and it can be seen that the strain response to an applied stress will be markedly different depending on whether the stress is applied in the prestrained direction or the reverse direction, according to the position of the obstacles on the right.

If, in the prestrained condition - i.e. dislocation position 2 - there is an increase in temperature, then the obstacles at ambient temperature may well not be so serious, since there are several thermally activatable processes which may occur with the aid of applied or internal residual stress, and which may change the situation.

The dislocation segment may "cross-slip", utilising other slip planes than the usual one; it may "climb", which involves displacements of small dislocation segments in planes other than slip planes; or, if the obstacle is another dislocation, then the breakage of attractive junctions and jog formation may occur. Screw dislocation containing jogs also moves more easily at higher temperatures. All these processes aid the dislocation to surmount or bypass obstacles, but all require thermal activation energy, which is more easily available the higher the temperature. The effect of increased temperature is to "randomise" the position of the dislocation by allowing it to surmount or bypass some of the obstacles; the position of the dislocation therefore tends to some random position, as typified by position 1 in Fig. 10(c), where applications of stress in any direction after the specimen has returned to ambient temperature will now produce the same stress-strain response. That is, this mechanism allows recovery of elastic properties without any bulk loss of internal residual stress or work hardening etc.

In general, however, the dislocation will not be completely randomised at the elevated temperature because certain, more persistent obstacles may well still be insurmountable. These will prevent the dislocation escaping altogether, and thus will be retained to take part in any subsequent creep or relaxation effects, unless the temperature has been sufficiently high to allow thermal activation for all obstacles, leading perhaps to some softening of the material and general loss of elastic properties.

With this mechanism the optimisation of elastic or relaxation properties may require a different LTHT, optimisation of relaxation occurring in general at a higher LTHT temperature than for optimum elastic properties.

In order to explain the data adequately, two further phenomena need to be accounted for:

(a) Some materials, particularly carbon steels, show large increases in tensile strength as well as elastic properties (Figs. 19,20,24). Although other materials, for example cold drawn brass wire[17], show only very small increases in tensile strength and others, for example, stable 7% Mn - 8% Ni austenitic steel[12], show no change in strength and only a small change in proof stress even though the proportional limit is much increased.

(b) Relatively pure iron (0.0019 - 0.0013% carbon), by suitable prior heat treatment in hydrogen, can be made to show no increase in yield stress after straining and LTHT[23] (Fig. 18). Similarly, pure gold, silver and copper also show no increase with LTHT after straining[34]. For example, with high purity, oxygen-free copper, the limit of proportionality and the 0.01% proof stress are unaltered by low temperature heat treatments after 5 and 15% tensile strain. On the other hand, with tough pitch copper containing 0.33% arsenic and 0.10% silver[35], the limit of proportionality after 5% tensile strain has been shown to be 18.5 $N/mm^2$, this being raised to 86.5 $N/mm^2$ by a treatment of 2 hours at 300°C. Again, a copper alloy containing 1% tin and 0.02% silicon, and with 50% cold work, had its limit of proportionality increased from 61.8 $N/mm^2$ to 308.9 $N/mm^2$.

The above phenomena can be explained if the obstacles to dislocation motion are precipiate particles rather than just other dislocations. The particles could sometimes also cause extra hardening by further precipitation at the LTHT or stressing temperature, notably in carbon steels.

Internal friction measurements have shown, for example, that the solubility of carbon in iron is reduced from 0.010 wt % at 650°C to about 0.0006% at 300°C and about $2 \times 10^{-7}$% at 20°C[30] and nitrogen from 0.10% at 650°C to about $3 \times 10^{-5}$% at 20°C[31]. It has been shown [30,32] that numerous fine precipitates can be formed at 20°C if sufficient time is allowed for the process. Several days at room temperature are required to allow a reasonable proportion of the precipitation to take place[30,33] and several months for the precipitation to be anywhere near completion. Increasing the temperature to 100 - 200°C or more, accelerates the process.

So it is generally quite possible to establish a dislocation-precipitate interaction model for the effects of strain and LTHT, if it is assumed that most metals and alloys contain impurities which are in the form of extremely fine precipitates either at the time of straining, or which form during LTHT or hot stressing. It is possible that the higher stress in hot stressing will lead to more dislocation-precipitate interactions which will increase the subsequent pinning and hardening at ambient temperature. Moreover, even if this does not occur, as pointed out previously[13], increased residual stress and the Bauschinger Effect will occur with hot stressing, since thermal pinning of the dislocations with an applied stress means less plastic reverse strain on subsequent release of the stress.

Some workers[21,25] have based their interpretation of the increasesin elastic proof and yield properties with hot stressing on enhanced precipitation caused by the stress itself, i.e. a pressure induced transformation effect. There is some argument whether conventionally applied stresses are capable of sufficient, if any, effect in this respect[36]. However, it has been shownmore recently, for example[33], that isostatic pressures of 3,500 $N/mm^2$ can cause a much higher density of smaller precipitates in a Fe-0.019% C alloy when aged at $300^0C$ compared with similar samples heated for the same time and temperatures at atmospheric pressure. Directionally applied stresses may also alter the orientation of precipitates during precipitation, which may influence subsequent mechanical properties. Fig. 26 shows the effect of precipitation of nitrides in a single crystal of iron with and without an applied stress; although the stress did not alter the quantity of nitride precipitation, the effect of alignment of the precipitate particles with the applied stress direction considerably modified the resulting stress-strain curve[37]. However, although such effects may contribute to enhancement of elastic and other properties in some instances, such stress-ageing effects are probably not generally required to explain the documented differences in properties for hot-stressing compared with straining followed by LTHT, which can also be explained by other effects as outlined previously.

## 5.2 Dislocation-Solute Atom Interaction Phenomena and Elastic Properties

An alternative, and perhaps at present more popular, theory of the cause of the increase in elastic properties of prestrained material with LTHT is connected with the phenomena of the recoverable yield point in low carbon steels and is known as strain-ageing. Strain-ageing manifests itself chiefly by an increase in yield or flow stress with ageing after or during straining, ageing after straining being classed as "static" and ageing during straining as "dynamic" strain-ageing. Other properties may also change during strain-ageing: return of a sharp point; rise in the ductile/brittle transition temperature and tensile strength; reduction in ductility; decrease in Snoek internal-fraction peak etc.

The theory argued by Cottrell[38,39,40] and now widely quoted, is that interstitial carbon and nitrogen atoms still in solution segregate to the dislocations in the iron crystals and there "anchor" them. However, this theory involves the conclusion that the increase of yield stress arising from this cause reaches a maximum when the quantity of dissolved carbon and nitrogen is sufficient to anchor all the dislocations. It does not of itself readily explain the rise in yield stress that occurs when the quantity of interstitial solute is greater than this very low value, nor the further rise that occurs when the conditions are such as to favour precipitation of the solute. Further illustration of this point is given in Fig. 27[30], which constitutes stress-strain curves of iron single crystals containing 0.002% carbon. Curve A was obtained immediately after the crystal had been cooled rather slowly from 890$^{\circ}$C, and Curve B was obtained from a similar crystal tested after 2 years at room temperature. The time was sufficient to precipitate any metastable carbon that remained in solution after the cooling from 890$^{\circ}$C, and the ageing resulted in a substantial increase of strength. From this, and, as previously mentioned, from internal friction tests the equilibrium solute concentration of carbon, which is 0.000 0002% it seems unlikely that this hardening effect is due to solute interactions.

Instead of the vaguely defined "atmospheres" of dissolved carbon or hydrogen atoms in the neighbourhood of a dislocation which constitutes ~~solute~~ interaction, an alternative is that dislocations are anchored by finely precipitated particles, since there is evidence that precipitated particles grow at dislocations[41].

However, there is also further evidence that, at room temperature, precipitation within the matrix (at vacancy-carbon defects) is preferred but at higher temperatures this tendency is less favoured. The abrupt change in the mode of precipitation is a well recognised characteristic of ageing in alpha iron. It has been shown[42,33] that, as the ageing temperature drops below $60^{0}C$, precipitation occurs primarily at matrix sites, while above this temperature, carbides form preferentially on dislocations.

Most commercial heat treatment schedules would not allow anywhere near sufficient time for complete precipitation to occur during cooling after, say, patenting or conventional hardening and tempering. Fig. 28, from the data given in Ref.(30), indicates how much and how long holding times need to be to allow significant precipitation to occur in iron with an equilibrium quantity of carbon originally in solutions at $400^{o}C$.

The limited amount of precipitation which does take place at temperatures somewhat above room temperature will tend to be coarse - perhaps too coarse for dislocation interactions or pinning - and the fine matrix precipitation which characterises room temperature precipitation will then be nucleated at room temperature[32], even if subsequent working followed by heating is necessary, first to speed up nucleation and then to allow the precipitates themselves to grow [33].

## 5.3 Relaxation

As previously pointed out, the close similarity in temperature range and the decrease of proof stress and residual stress (characterised by "wind up" of coil springs) and drop in relaxation with LTHT of patented cold drawn carbon spring steel in Fig, 24, leads one to suspect that the same underlying phenomenon is responsible for all these changes. The application of stress at the LTHT temperature (Figs. 23 and 25) gives a similar and more pronounced drop in relaxation over the same prior treatment temperature range as for LTHT (the relaxation tests themselves being conducted in the range $20^{o}C$-$175^{o}C$).

Can this relaxation behaviour be explained on the basis of the dislocation interaction or perhaps solute interaction theories already described? In fact, according to Fig.24, relaxation drops rapidly as the dislocation pinning mechanism, whatever that might be, breaks down, i.e. the proof stress falls as the relaxation falls. In other words, the dislocations responsible for holding the internal residual stresses break away from their pinning sites due to stress-assisted thermal activation.

These dislocations, whose presence is necessary for creep and relaxation to occur in metals, are consequently allowed to escape to grain or cell boundaries. Thus, as shown in Fig. 24, a simultaneous loss of elasticity occurs - due to removal of some of the residual stresses responsible for part of the Bauschinger Effect - and a reduction in relaxation due to loss of the potentially active dislocations to the grain or cell boundaries. The additional assumption here is that, in cold worked metals and alloys, the dislocations in the tangles which constitute the cell walls(30) do not, or are less likely to, contribute to creep and relaxation processes than are the more mobile dislocations, which are themselves likely to interact only with precipitates and solutes rather than mutual interactions within the cell walls. The fact(10) that screw dislocations are pinned by interpenetration or interaction with other dislocations due to the formation of jogs, whereas the interpenetrations of an edge dislocation with another dislocation (any jogs formed here being mobile) does not impair its motion, implies that cell walls are constituted mainly from screw dislocations whereas edge dislocations require precipitates to pin them if sufficient thermal or stress energy is present at room temperature for jog formation or to resist solute or vacancy pinning effects.

The formation of grids of dislocations in copper(43) and iron(44) is predicted by theoretical models(30) for cell walls which support the idea that such walls are composed mainly of screw dislocations.

Additionally, the idea that the dislocation population is composed of immobile and a relatively small proportion of mobile dislocations has been treated elsewhere[45]. The discussion is again based on plastic flow, in which the strain rate is controlled by the rate at which the mobile dislocation segments surmount local obstacles by means of stress-assisted thermal activation, this being the controlling factor for plastic flow. The mobile dislocation segments are thus those which, although temporarily pinned, are thermally activatable, whereas others will be non-activatable and will thus remain immobile. A model based upon a network of such linked dislocations, which is shown to be consistent with plastic flow behaviour in a polycrystalline nickel chromium steel, predicts that the fraction of mobile dislocations becomes a relatively small proportion of the total number of dislocations with increased straining and reaches an almost constant level after an initial transient. Other experimental evidence on alpha iron[46] supports the idea that the density or quantity of mobile dislocations is approximately constant during plastic flow.

It may thus be supposed that only a fraction of the total dislocation content is mobile, as defined previously. This view may account for the sudden and simultaneous removal of residual stresses and reduction in relaxation behaviour with LTHT or hot prestressing which at a critical temperature allows thermally activated unpinning of the mobile dislocations from precipitates or other obstacles to their motion.

At somewhat higher temperatures the cell walls themselves are reorganised in a recovery process which leads to immediate softening of the material. This has an adverse effect on relaxation because the load bearing capabilities are then reduced and the relaxation then increases again rapidly. This is indeed what is seen to be happening in Figs 24 and 25.

The hardened and tempered steels are stable in this respect up to their tempering temperature, generally about 450$^{\circ}$C, and so, as illustrated in Fig 23, there is no increase in relaxation again up to at least 400$^{o}$C.

The fact that the rapid drop in relaxation with prior LTHT and hot stressing occurs at about the same temperature indicates that it is the removal of mobile dislocations which is the important requirement rather than, say, merely a removal of the transient part of the relaxation process by a prior creep interval imposed during LTHT or hot stressing, since during LTHT the residual stresses at any given point in the specimen are acting in the opposite direction to that of the applied stress in hot stressing - certainly in a coil spring.

If the density of mobile dislocations were to remain the same, it would be expected, during LTHT, that relaxation would increase, since presumably the dislocations would need to move back to where they were before LTHT, as well as then proceeding to generate the normal quota of relaxation with an applied load. Because this does not in fact happen it must be assumed that it is the loss of mobile dislocations during heating that robs the relaxation process of its means of operation.

It is certainly well known in the spring industry that in some cases relaxation can be reduced if prestressing is minimised, i.e. if the number of mobile dislocations is in this way kept as small as possible.

The reason that hot prestressing, as compared to LTHT, leads to even lower values of relaxation is evident from the foregoing interpretations of the microstructural behaviour. Clearly, hot prestressing enables the advantageous residual stresses to be maintained whilst at the same time removing the cause of subsequent relaxation - the mobile or thermally activatable dislocations. Upon removal of the load after hot prestressing the beneficial internal residual stresses are stable, in the sense that they do not depend for their existence on thermally activatable dislocations. When reloaded in a subsequent relaxation test or in service the total stress is reduced by the stable residual stresses which of course helps to reduce subsequent relaxation effects as does of course the low level of thermally activatable dislocations. With straining followed by LTHT, however, although a similar amount of thermally activatable dislocations may have been removed (or perhaps less, since there is less stress-assistance), their removal will have resulted in some loss of internal residual stress which, upon subsequent loading for relaxation will mean that the total stress at any point within the specimen or component will be higher and consequently greater relaxation will ensue. The relaxation behaviour is thus not inconsistent with a matrix precipitate dislocation interaction view of elastic property improvement with heating - a solute interaction interpretation would perhaps be less able to account for the rather steep drop in relaxation in the range 150 - 350°C, since solute interaction effects are expected to persist up to much higher temperatures[47]. For example, carbon and nitrogen are still expected to segregate strongly to dislocations at temperatures up to at least 670°C in iron, although of course they are gradually less effective as the temperature is increased. The relatively sudden drop in relaxation and residual stress at about 300°C is thus difficult to reconcile with the solute interaction phenomenon.

A dynamic strain ageing mechanism favoured by Evans and Battacharya[22] would produce quite the opposite effect on the mobile dislocation density since the continuous pinning by solute atoms[47,48] and creation of new dislocations at temperatures in the region of 200 - 300°C does not solve the problem of removal of mobile dislocations leading to less relaxation since new dislocations are continuously being created by this mechanism. Moreover Evans and Battacharya failed to find evidence in their electron micrographs for increases in dislocation density in carbon steel by this mechanism. They also found that conductor purity aluminium wire (containing small amounts of impurities) also showed a substantial improvement in relaxation after hot stressing compared with straining followed by LTHT, whereas relaxation in pure iron and OHFC copper with low interstitial impurity content was not affected, whether hot stressing or straining followed by LTHT was used. This again supports the contention that precipitation is an essential requirement to explain elastic property and relaxation behaviour in metals and alloys.

Any dislocation precipitation would also be expected to be still effective at 300°C, or at least relatively insensitive to temperatures up to 300°C, when according to Fig. 28 there will be some driving force for part of the precipitate particles to redissolve - however, this process could perhaps be expected to be relatively slow, being presumably diffusion-dependent. Breakaway of dislocations in a situation where the precipitate is distributed along much of the length of the dislocation - a situation which is quite likely since growth occurs preferentially along the dislocation - seems difficult to imagine as being thermally activatable over the range of temperatures up to 300°C.

## 6 CONCLUSIONS

6.1 The observed improvements in elastic and sometimes other mechanical properties, and also in relaxation, in most metals and alloys, including spring materials, subject to either straining followed by low temperature heating or by hot stressing, may be most satisfactorily explained by the presence of mobile dislocations which interact with a fine precipitate in a manner analogous to dispersion hardening, and which can further interact and ultimately break away from the precipitation by stress-assisted thermal activation to an extent dependent on the temperature of subsequent hot stressing or LTHT.

6.2 The fine precipitation necessary for this mechanism may be provided by small quantities of interstitial or other impurities. It has been shown that only metals and alloys containing sufficient impurities to provide the necessary fine precipitate density display improvements in elastic and relaxation properties to be obtained from suitable stress strain and temperature changes.

## 7 RECOMMENDATIONS FOR FUTURE WORK

7.1 Further elucidation of the mechanism responsible for the effects which are observed with hot stressing and treating of formed springs could be obtained if suitable experiments were devised.

For instance, suitable applications of stress and strain followed by LTHT, if repeated in cycles of straining and LTHT, could distinguish whether matrix precipitation or dislocation precipitation were operative in a particular material. If it were matrix precipitation, the first LTHT would produce the most benefit to elastic properties and subsequent LTHT's would provide progressively less benefit, levelling out to a constant improvement as precipitation eventually became complete. If dislocation precipitation were the operative mode then only the first LTHT would be really beneficial and subsequent LTHT's would eventually provide no further improvement in the elastic limit because the precipitation would have been "spent" on pinning dislocations in earlier strain and LTHT cycles. If solute interaction were the cause, then each straining operation would destroy the effects of the previous LTHT and the same actual increase in the elastic property would then occur at each LTHT stage. Thus each cycle of strain and LTHT would produce essentially the same results, except, of course, for the increases in strength due to the extra work hardening at each strain stage.

This kind of experiment could usefully be done using coil springs where strain could be applied by extending the springs to required lengths and tests make by accurate load-extension measurements to determine comparative elastic ranges. This technique avoids the use of extensometers which would be required if the material were tested as wire, with the attendant problem of deciding where the elastic or proportional limit is, since this depends on the inherent accuracy of the extensometer. The use of coil springs enormously amplifies length changes during testing and consequently enables accurate, if only comparative, stress-strain data to be obtained.

7.2 In Section 5.3 of this report, the point is made that hot prestressing causes a stable and presumably high level of residual stress to be retained in a spring whilst removing the more harmful (in terms of relaxation) dislocations. It is evident here that the relaxation under an applied load will be lower because the stable residual stresses oppose the externally applied stress. It may then be that the total stress at any point within the specimen will actually be much higher when there is no external load on the spring because the internal residual stress is not cancelled by an external stress. Thus the relaxation under applied load may well be lower than the relaxation which occurs when a similar sample is tested at the same temperature without an externally applied load.

Tests have not been made to check this point in the hot stressing work done so far apart from the work of Graves[49] on nickel-based alloys which showed "negative" relaxations with low applied loads in relaxation tests after a hot stressing operation.

For applications which require that relaxation in service should be as low as possible, not only under load but when unloaded, (e.g. dynamic applications which may work under conditions of low or no applied initial load, as well as a higher final working load) it is suggested that future investigations on hot prestressing and LTHT take this into account and appropriate checks for this effect should be carried out where relevant.

7.3 The effects of the amount of strain and residual stress content of a spring should be investigated to decide the optimum amount and method of prestressing for a given service requirement in respect of stress range and relaxation requirement.

For example: compression springs are normally prestressed several times in compression to increase the elastic range in that direction. However, if prestressing were performed both in compression and extension (once or more than once) to the same extent, then this would remove some of the residual stress and Bauschinger Effect whilst still giving the required elastic range in compression. What would be the effect on relaxation, static or dynamic (or other metallurgical properties i.e. fatigue) of the removal of this residual stress, in effect, by conversion to work hardening (see Fig. 11)? What would be the effect of a subsequent LTHT or hot stressing operation on these properties?

In such a way, both a greater understanding of the mechanisms involved and optimisation of production techniques for springs for particular applications may ensue.

7.4 If subsequent tests follow the trend of past experiments, i.e. the same mechanisms are operative, then relaxation in cold worked 18/8 stainless steel should suffer a relatively sharp drop with prior LTHT or hot stressing temperatures in the region of 450 - 500°C and above (if these temperatures are not so high that recovery is taking place). The reason for this is that, at 400 - 450°C, the elastic properties show a sharp decline which, if consistent with the mechanism put forward in this report, will be due to the thermal breakaway of mobile dislocations from matrix precipitates.

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TABLE I SOME FURTHER EXAMPLES OF ALLOYS WHICH SHOW INCREASES IN TENSILE MECHANICAL PROPERTIES WITH STRAIN, LTHT AND HOT STRESSING

| Alloy | Condition | Stress $N/mm^2$ | Temp. ($^oC$) | Time (hours) | Proportional Limit ($N/mm^2$) | 0.2% Proof Stress ($N/mm^2$) | Ultimate Tensile Strength ($N/mm^2$) |
|---|---|---|---|---|---|---|---|
| 70-30 brass[21] cold rolled 52% | as rolled | - | - | - | 193 | 476 | 586 |
| | LTHT | 0 | 175 | 4 | 324 | 497 | 598 |
| | hot stressed | 372 | 175 | 4 | 490 | 599 | 612 |
| 5%Sn- [21] bronze cold rolled 60% | as rolled | - | - | - | 305 | 596 | 602 |
| | LTHT | 0 | 175 | 4 | 359 | - | 563 |
| | hot Stressed | 517 | 175 | 4 | 517 | 593 | 599 |
| 99%Cu-1%Cd[21] cold rolled 60% | as rolled | - | - | - | 207 | 481 | 483 |
| | LTHT | 0 | 225 | 4 | 248 | - | 443 |
| | hot stressed | 447 | 225 | 4 | 460 | 495 | 495 |
| 18% Beryllium copper wire[36] | 1. Solution treated and straightened | - | - | - | 365 | 448 | 631 |
| | 2. (aged) | 0 | 315 | 2 | 1040 | 1379 | 1426 |
| | 3. Cold stressed after 2 | 1365 | - | - | 1367 | 1411 | 1563 |
| | 4. Hot stressed after 2 | 1365 | 200 | 2 | 1401 | 1425 | 1438 |
| | 5. Drawn 50% straightened and aged at 315$^oC$ | - | - | - | 1125 | 1405 | 1414 |
| | 6. As 5 cold stressed | 1379 | | | 1156 | 1449 | 1467 |
| | 7. As 5 hot stressed | 1379 | 200 | 2 | 1262 | 1445 | 1462 |
| 18Cr-9Ni[21] stainless steel 38% cold worked | 1. As cold worked | - | - | - | 499 | 1231 | 1332 |
| | 2. Cold stressed | 1034 | 25 | 4 | 827 | 1218 | 1230 |
| | 3. Hot stressed (only) | " | 400 | 4 | 1062 | 1420 | 1440 |
| | 4. LTHT (only) | 0 | 400 | 4 | 665 | 1241 | 1358 |

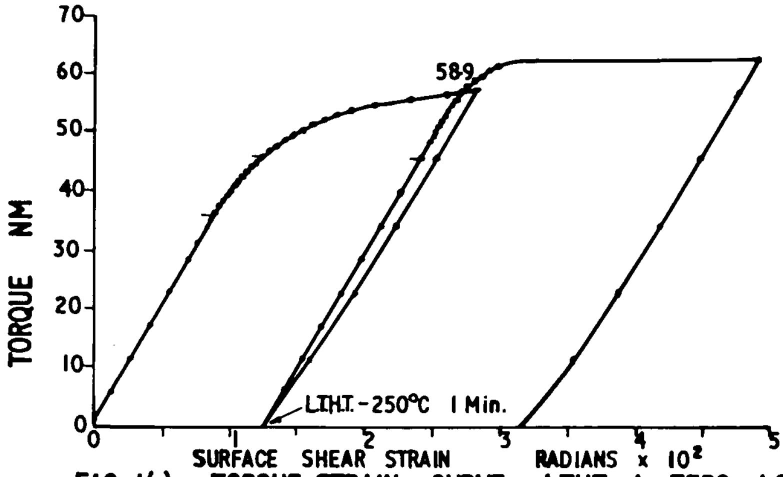


FIG. 1(a) TORQUE-STRAIN CURVE L.T.H.T. A ZERO LOAD FOR 12·5 mm dia. HARDENED AND TEMPERED CARBON SPRING STEEL.(1)

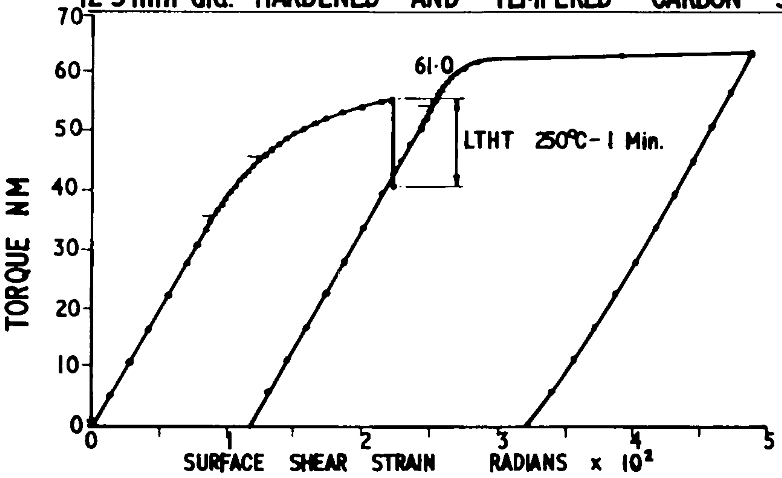


FIG. 1 (b) TORQUE - STRAIN CURVE : L.T.H.T. ON LOAD FOR 12·5 mm dia. HARDENED AND TEMPERED CARBON SPRING STEEL.(1)

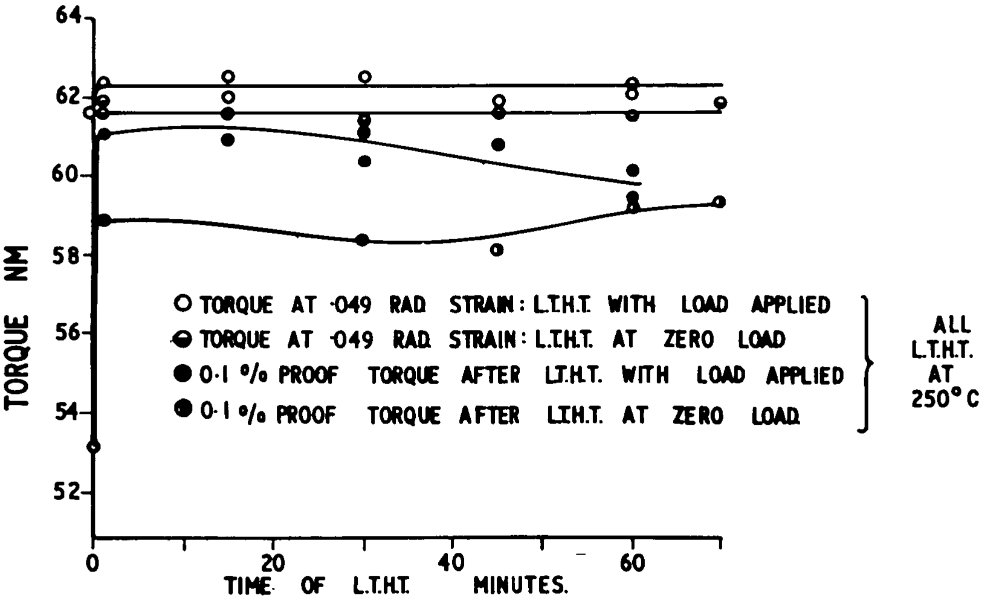


FIG. 2 EFFECT OF TIME OF L.T.H.T. UPON 0·1% PROOF TORQUE FOR 12·5 mm dia. HARDENED AND TEMPERED CARBON SPRING STEEL. (1)

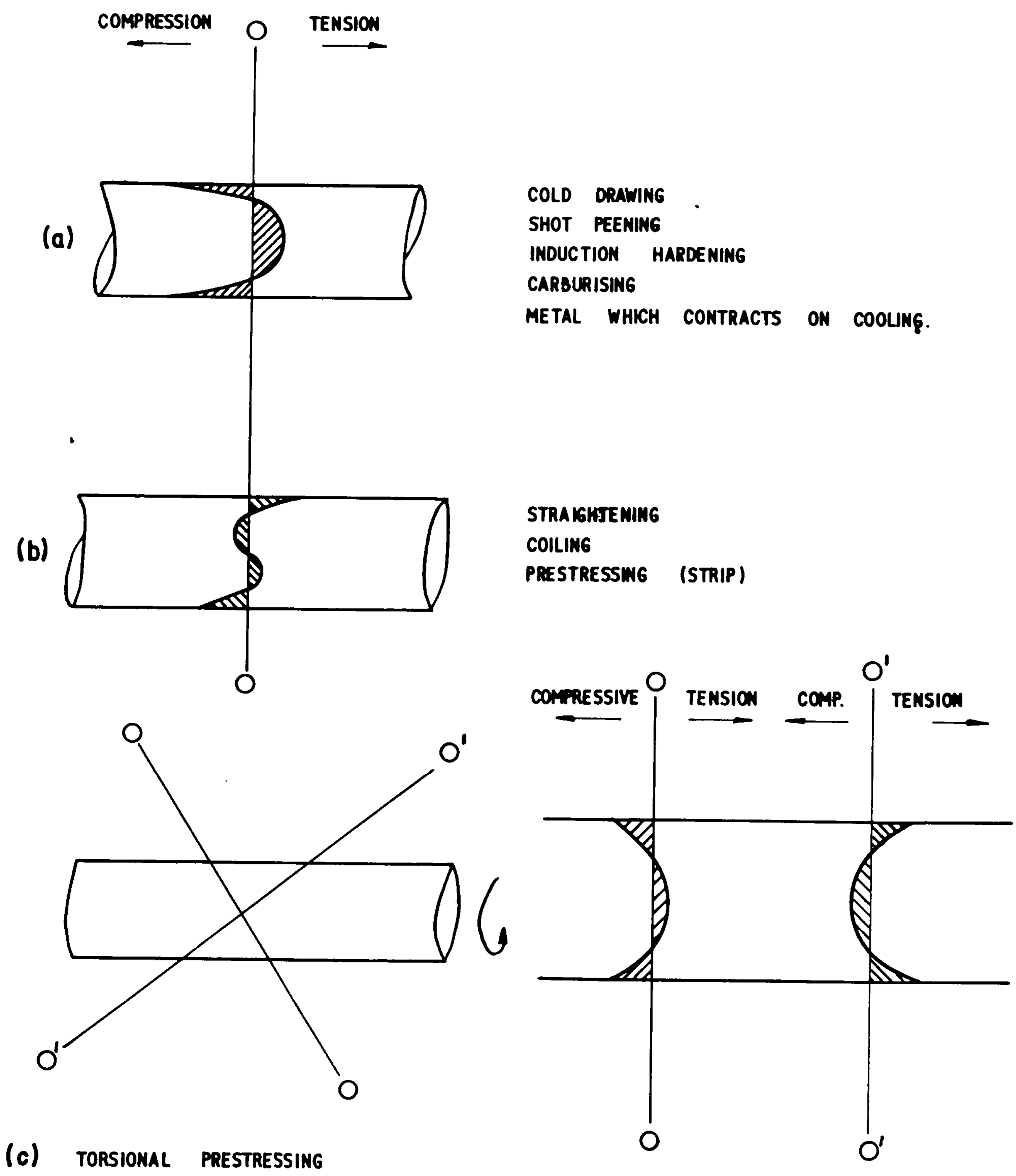


FIG. 3 RESIDUAL STRESS PATTERNS DUE TO VARIOUS PROCESSING OPERATIONS.

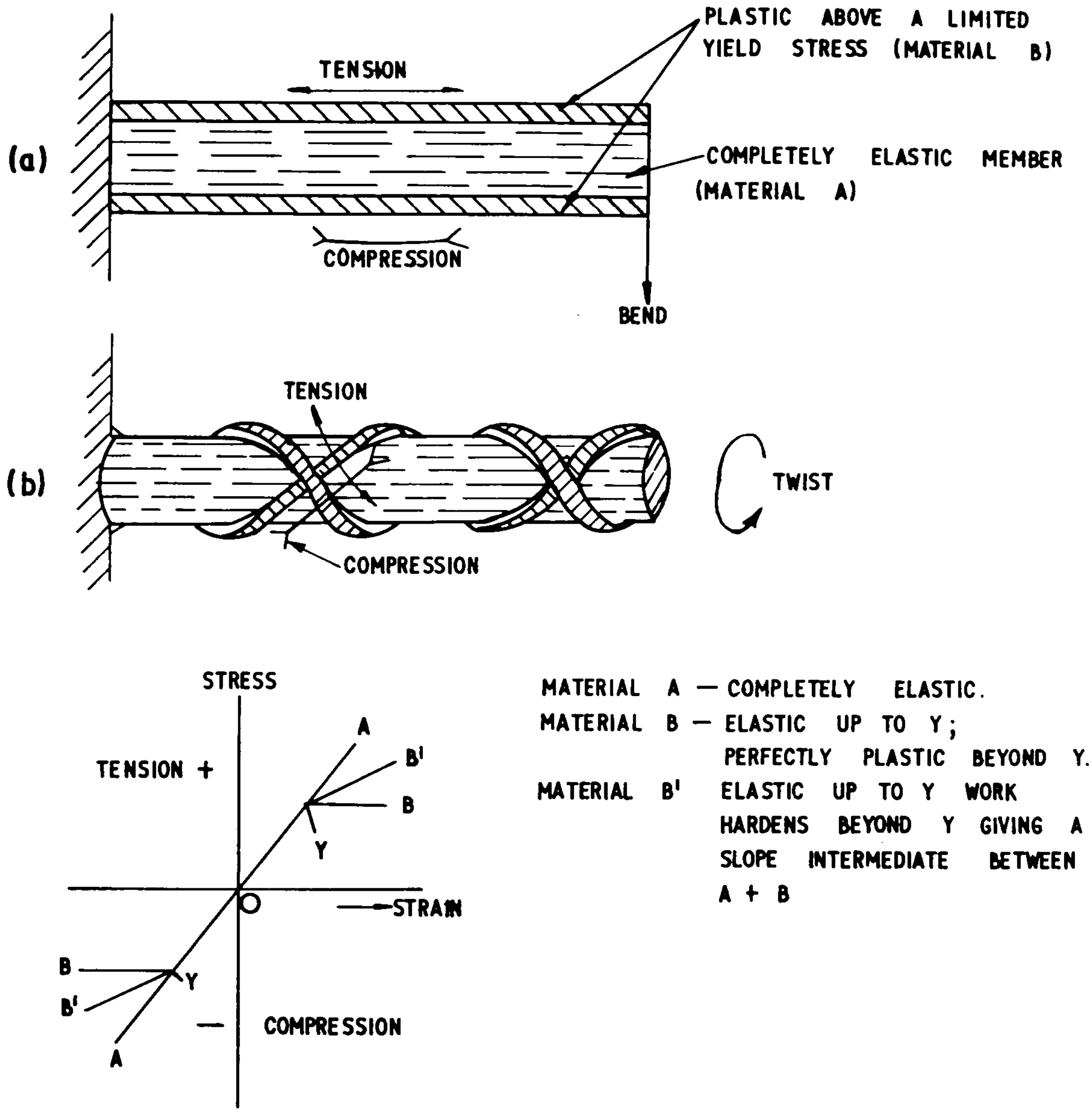


FIG. 4 SIMPLIFIED CONSTRUCTIONS FOR (a) BENDING AND (b) TORSION TO ILLUSTRATE THE EFFECTS OF MACROSCOPIC RESIDUAL STRESS.

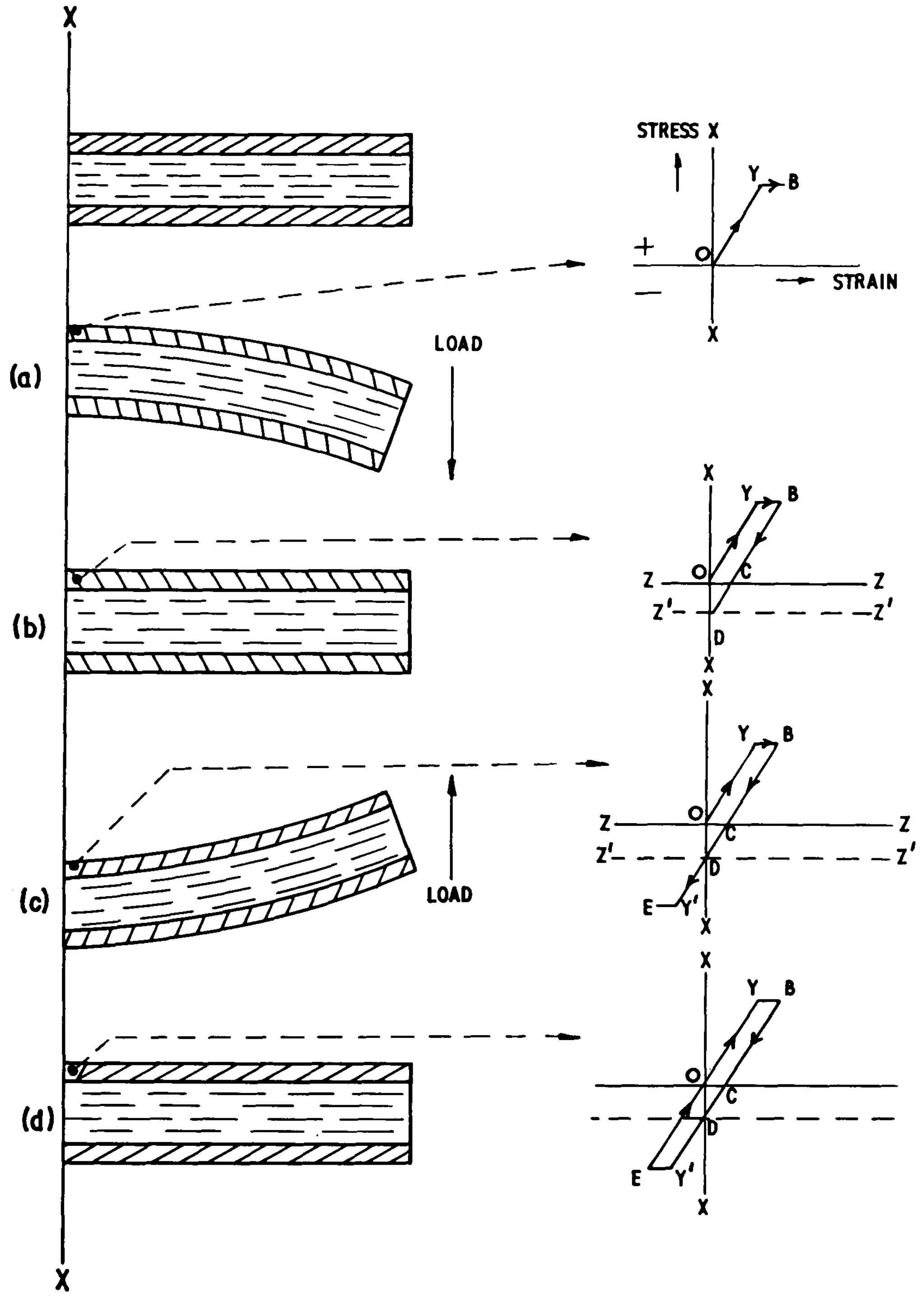


FIG. 5 STRESS – STRAIN BEHAVIOUR IN BENDING USING THE MODEL IN FIG. 4 (a) IN AN ELASTIC – PLASTIC SURFACE LAYER.

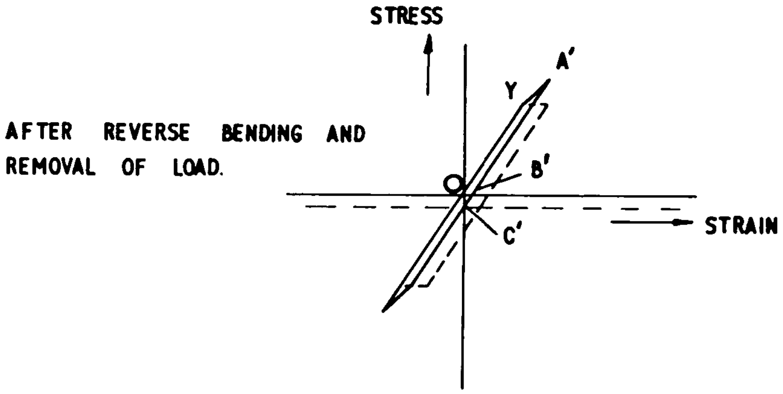


FIG. 6 STRESS - STRAIN BEHAVIOUR USING THE MODEL IN FIG. 4 (a) IN AN ELASTIC - PLASTIC SURFACE LAYER WITH WORK HARDENING COMPARED WITH BEHAVIOUR IN THE ABSENCE OF WORK HARDENING.

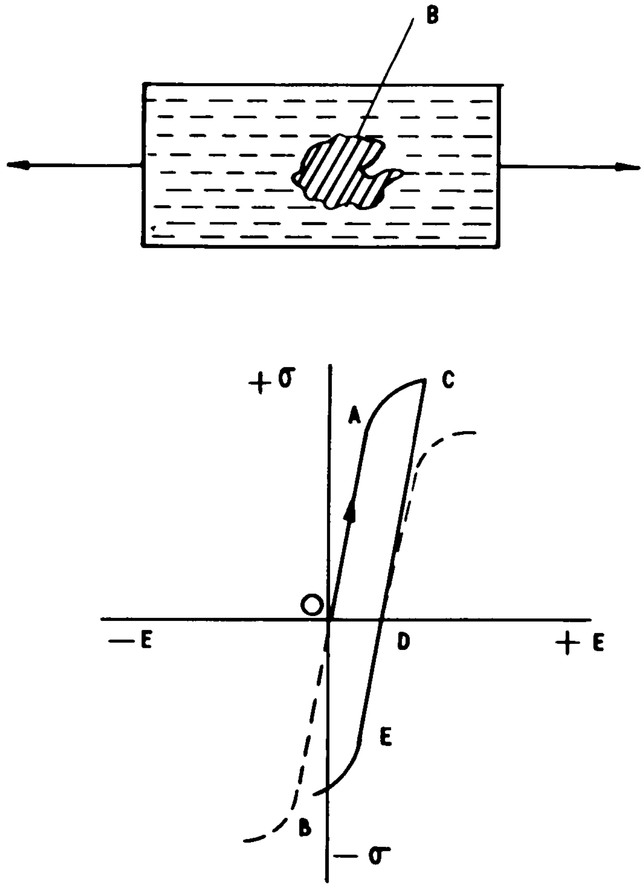


FIG. 7 BAUSCHINGER EFFECT IN TENSILE AND COMPRESSIVE LOADING.

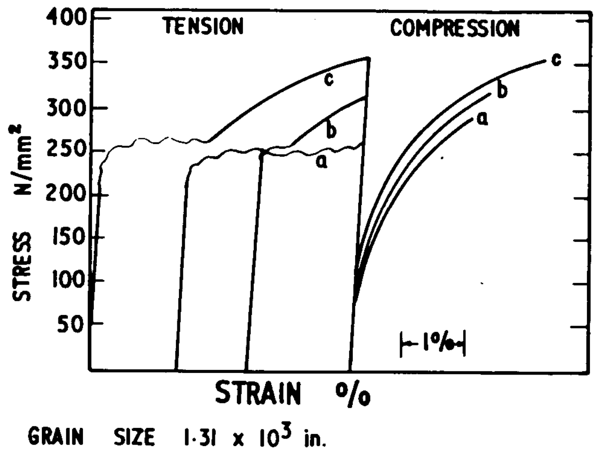


FIG. 8 TENSION AND COMPRESSION BEHAVIOUR OF STEEL.[6]

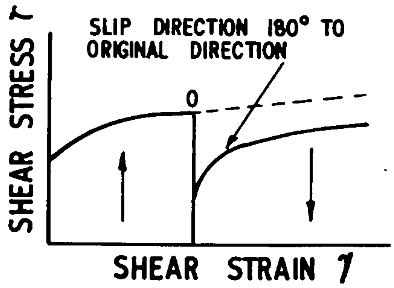


FIG. 9 EFFECT OF COMPLETE REVERSAL OF SLIP DIRECTION ON STRESS - STRAIN CURVE.[8]

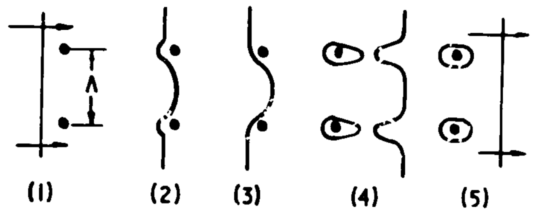


FIG. 10 (a) SCHEMATIC DRAWING OF STAGES IN PASSAGE OF A DISLOCATION BETWEEN WIDELY SEPARATED OBSTACLES - BASED ON OROWAN'S MECHANISM OF DISPERSION HARDENING.[9]

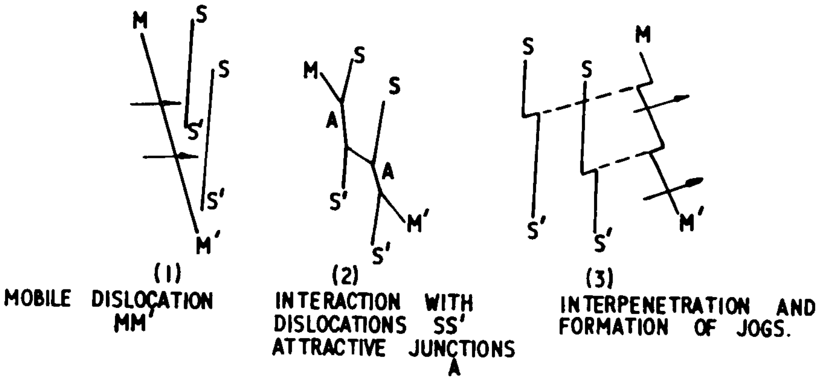


FIG. 10 (b) OTHER DISLOCATIONS AS OBSTACLES TO MOBILE DISLOCATIONS.

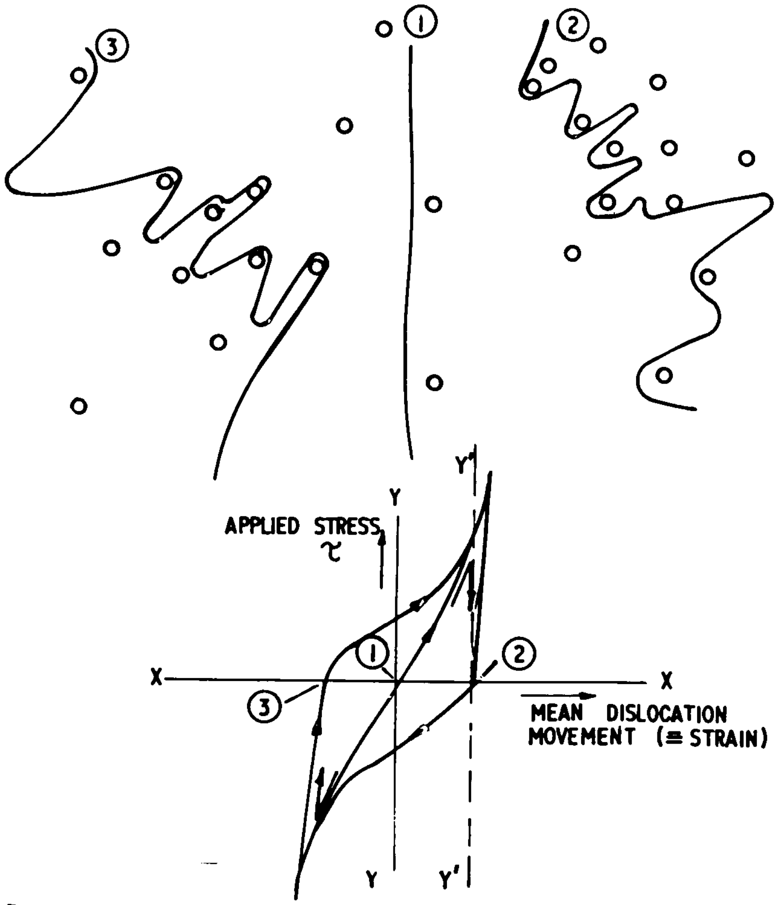


FIG. 10 (c) DISLOCATION SEGMENT IS STOPPED AT A DENSE CLUSTER OF OBSTACLES. ② ON STRESS REVERSAL IT MOVES BACK AND IS STOPPED IN POSITION ③ BY THE NEAREST CLUSTER OF EQUAL DENSITY.[9]

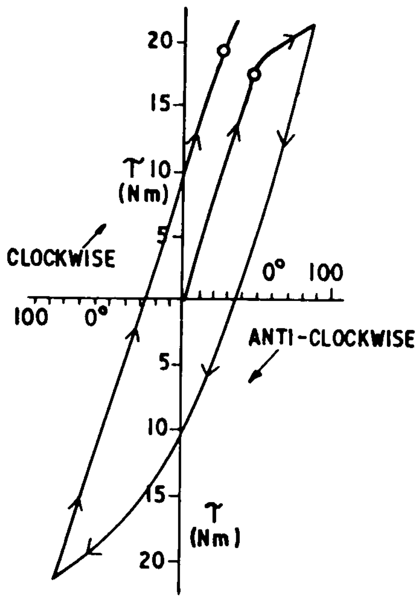


FIG. 11 TORSIONAL LOADING AND SUBSEQUENT REVERSED LOADINGS OF A HARD DRAWN SPRING STEEL. 4·67mm DIAMETER WIRE. [11]

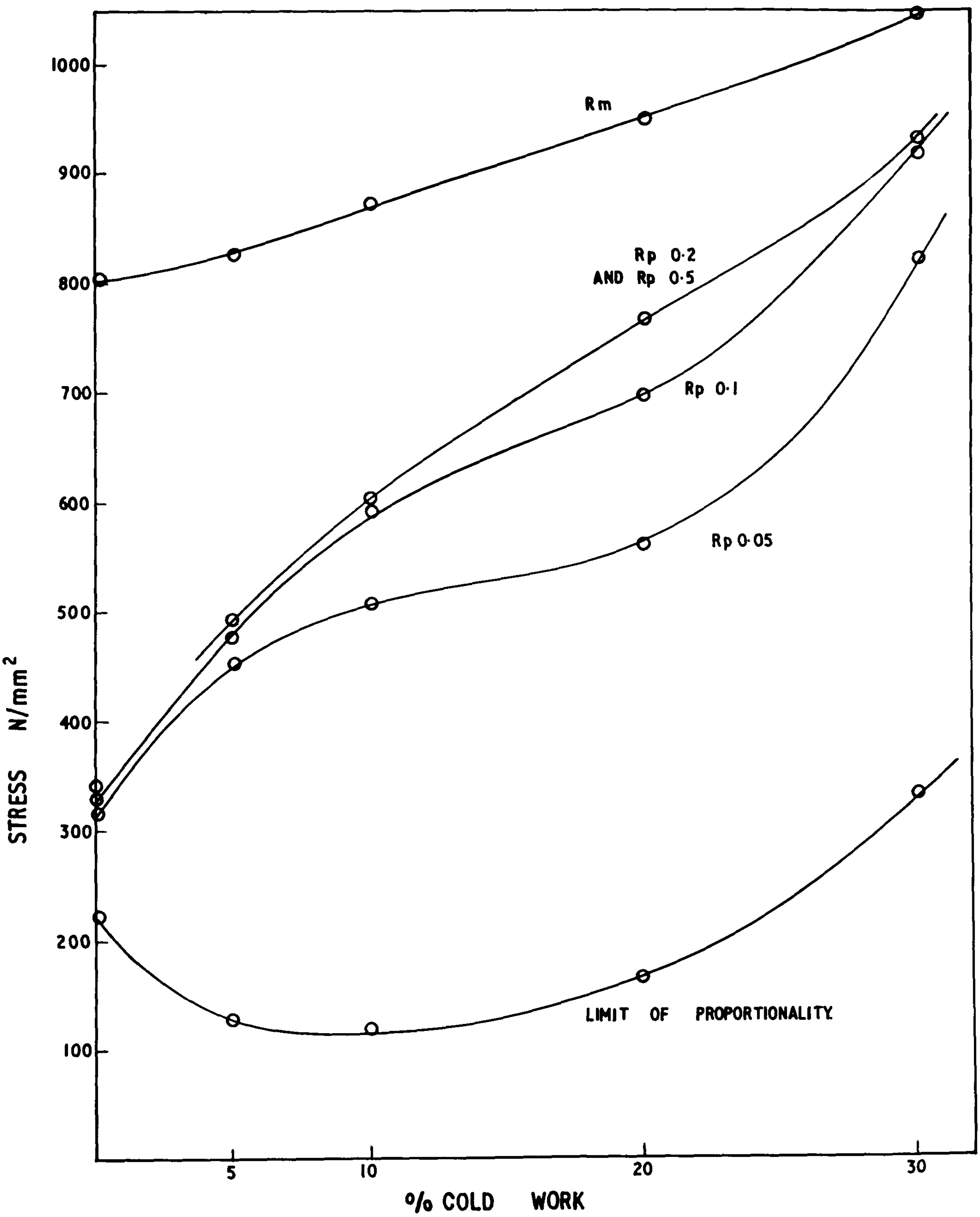


FIG. 12 EFFECT OF COLD WORK ON THE MECHANICAL PROPERTIES OF A STABLE AUSTENITIC STEEL ( 0·62 % C, 0·35% Si, 7·32% Mn, 3·86% Cr, 8·55 % Ni, 0·021 % N ) [12]

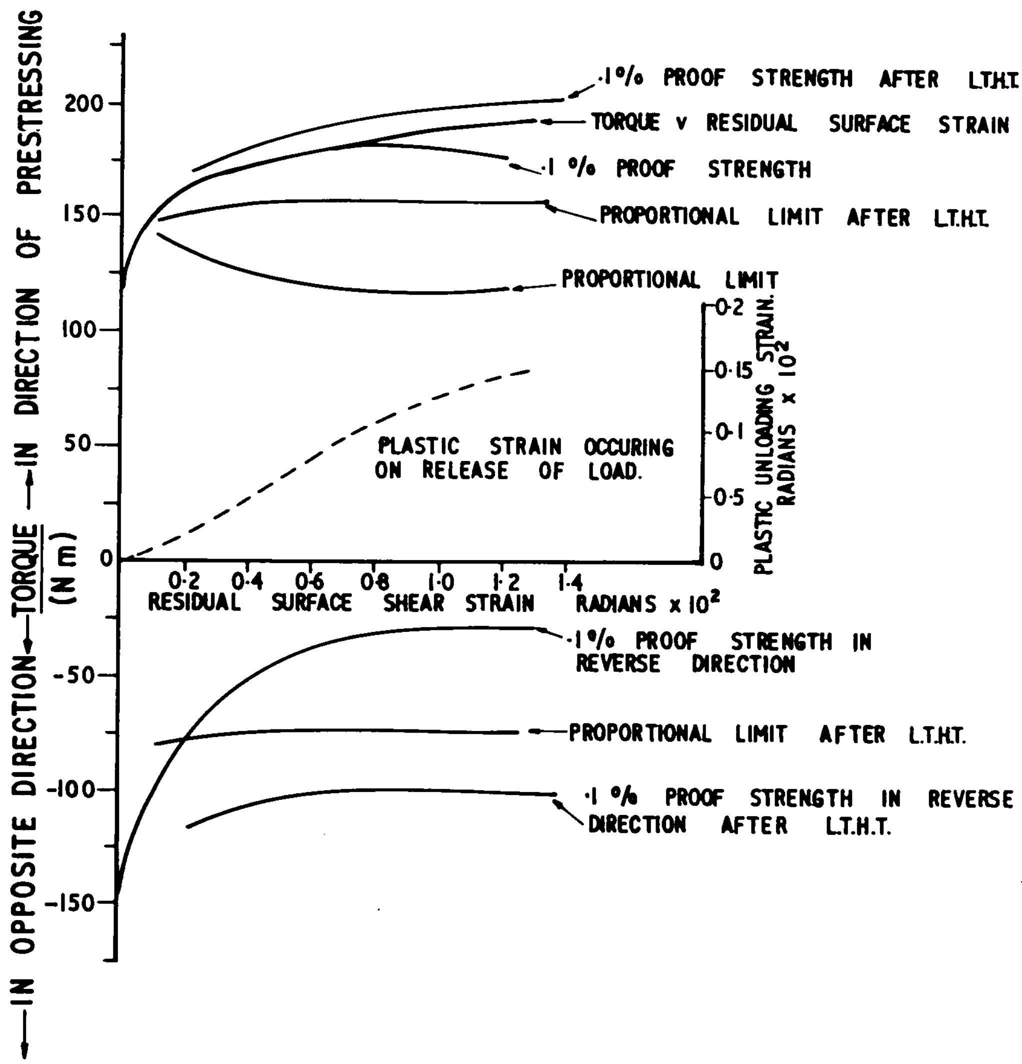


FIG. 13 MECHANICAL PROPERTIES IN TORSION AFTER PRESTRESSING BY VARIOUS AMOUNTS AND BOTH BEFORE AND AFTER L.T.H.T. FOR 9·53 mm DIAMETER HARDENED AND TEMPERED CARBON SPRING STEEL BAR.[13]

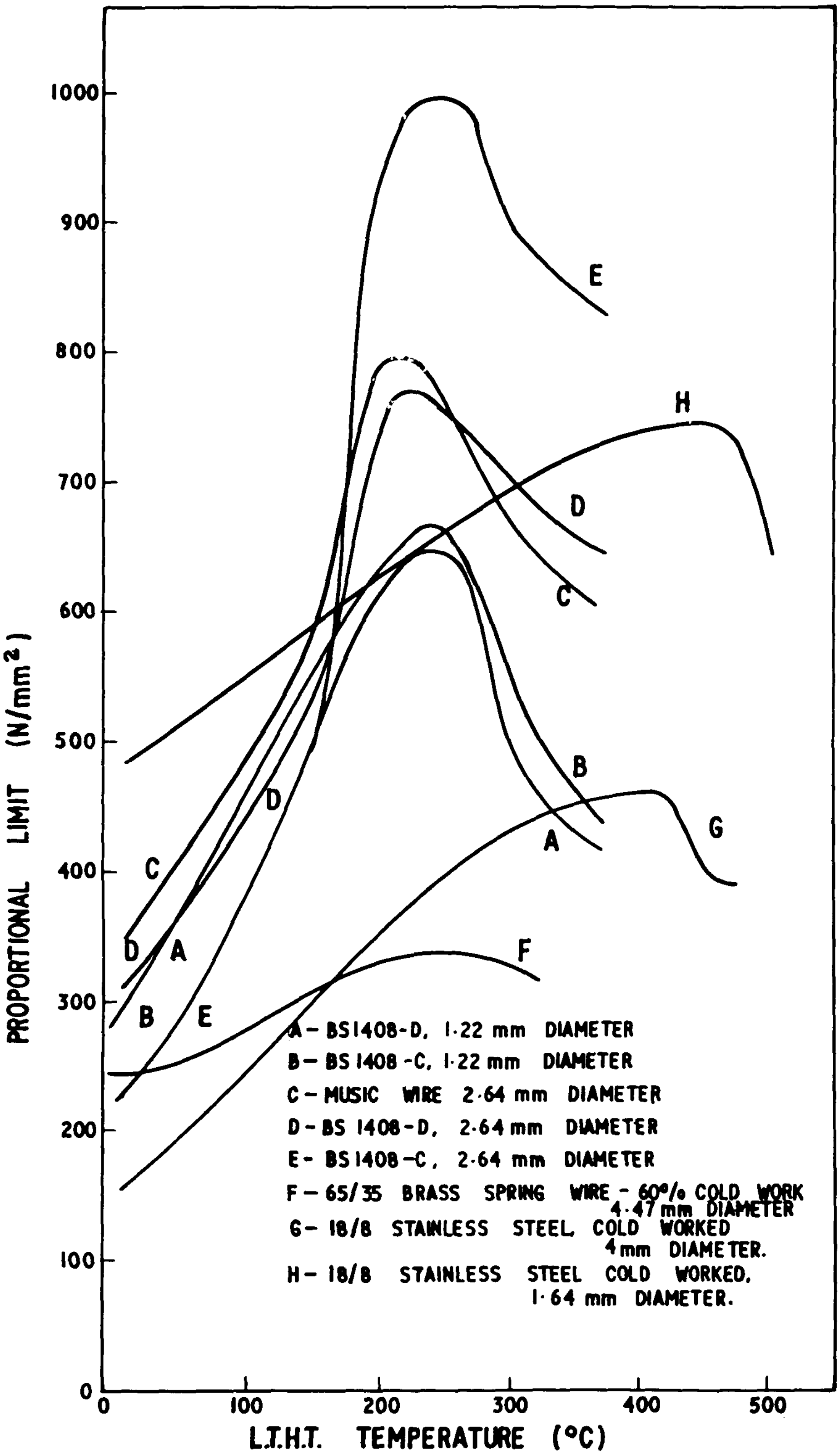


FIG. 14 VARIATION IN PROPORTIONAL LIMIT IN TORSION WITH L.T.H.T. FOR VARIOUS COLD DRAWN SPRING WIRES [14–17] (ALL L.T.H.T.'s FOR 30 min. EXCEPT STAINLESS STEEL WHICH ARE FOR 2 HOURS)

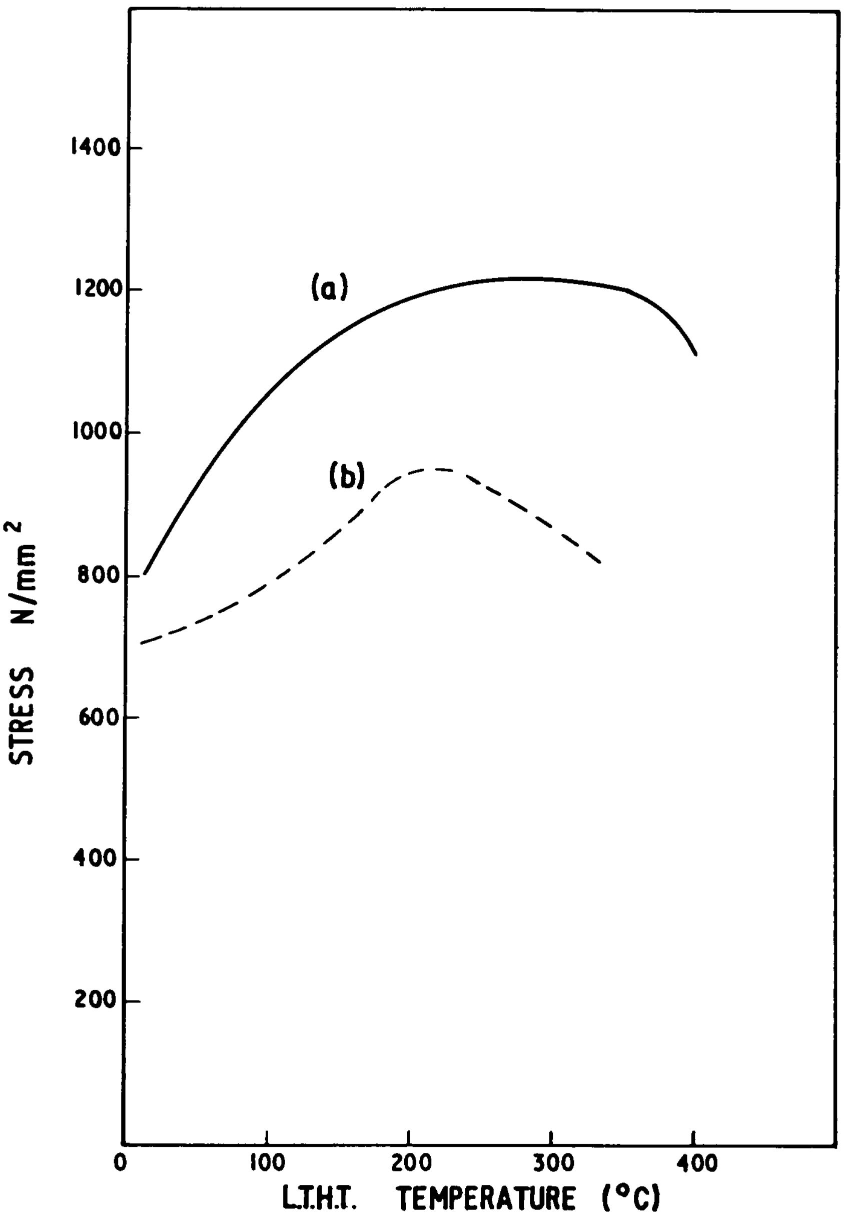


FIG. 15 VARIATION OF 0·1 % PROOF STRESS IN TORSION FOR (a) 12·7 mm DIAMETER HARDENED AND TEMPERED SILICON-MANGANESE STEEL WITH L.T.H.T. AFTER PRESTRESSING TO 0·013 RADIANS RESIDUAL SURFACE STRAIN [19] ALSO 0·1 % PROOF STRESS IN TORSION FOR (b) BS 1408 M - RANGE 3mm DIAMETER WIRE WITH LT.H.T. FOR COMPARISON. [18]

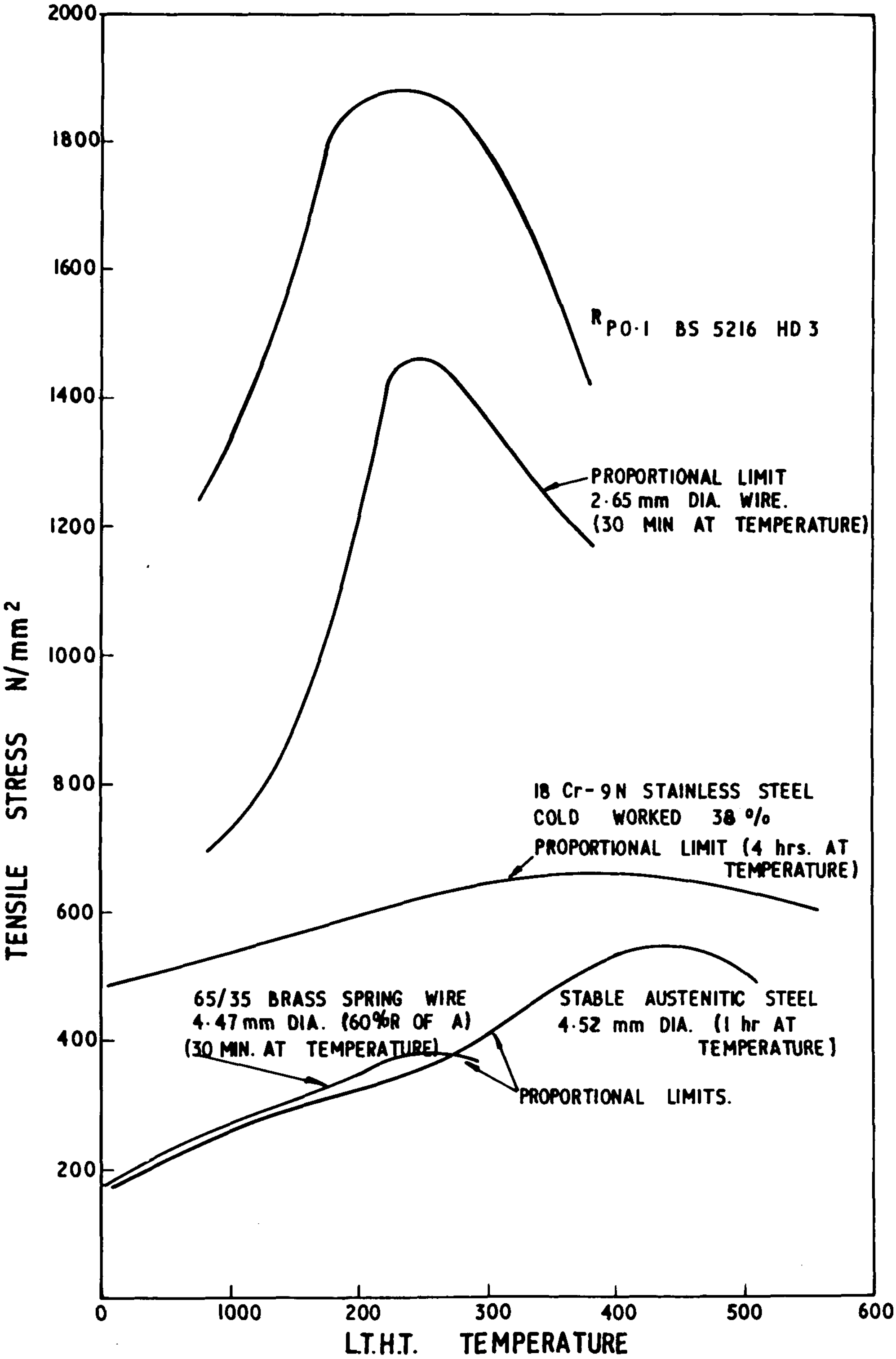


FIG. 16 VARIATION OF PROPORTIONAL LIMIT IN TENSION WITH L.T.H.T. TEMPERATURE FOR SOME SPRING AND POTENTIAL SPRING MATERIALS. (12, 17, 20, 21.)

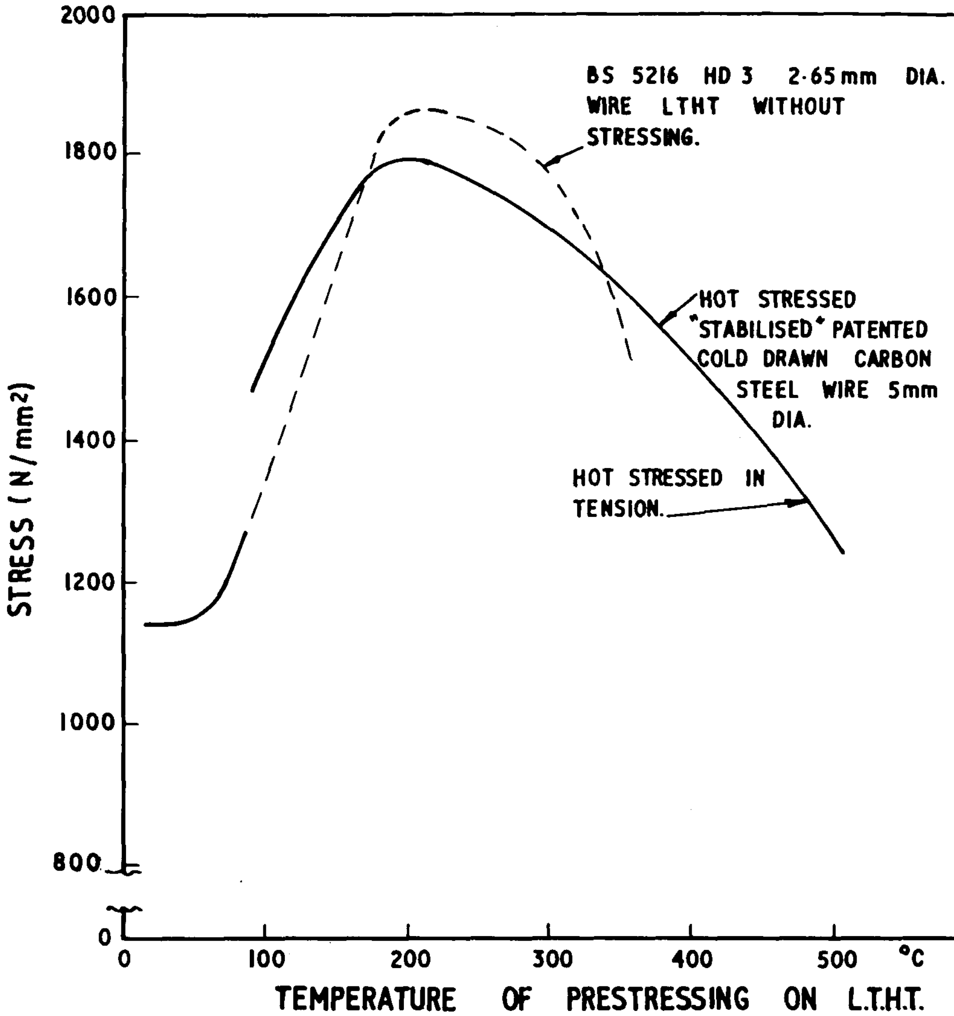


FIG. 17 VARIATION OF 0·1 % PROOF STRESS IN TENSION FOR PATENTED HARD DRAWN WIRE (a) WITH L.T.H.T. FOR 30 min.[20] AND (b) HOT PRESTRESSED IN TENSION.[22]

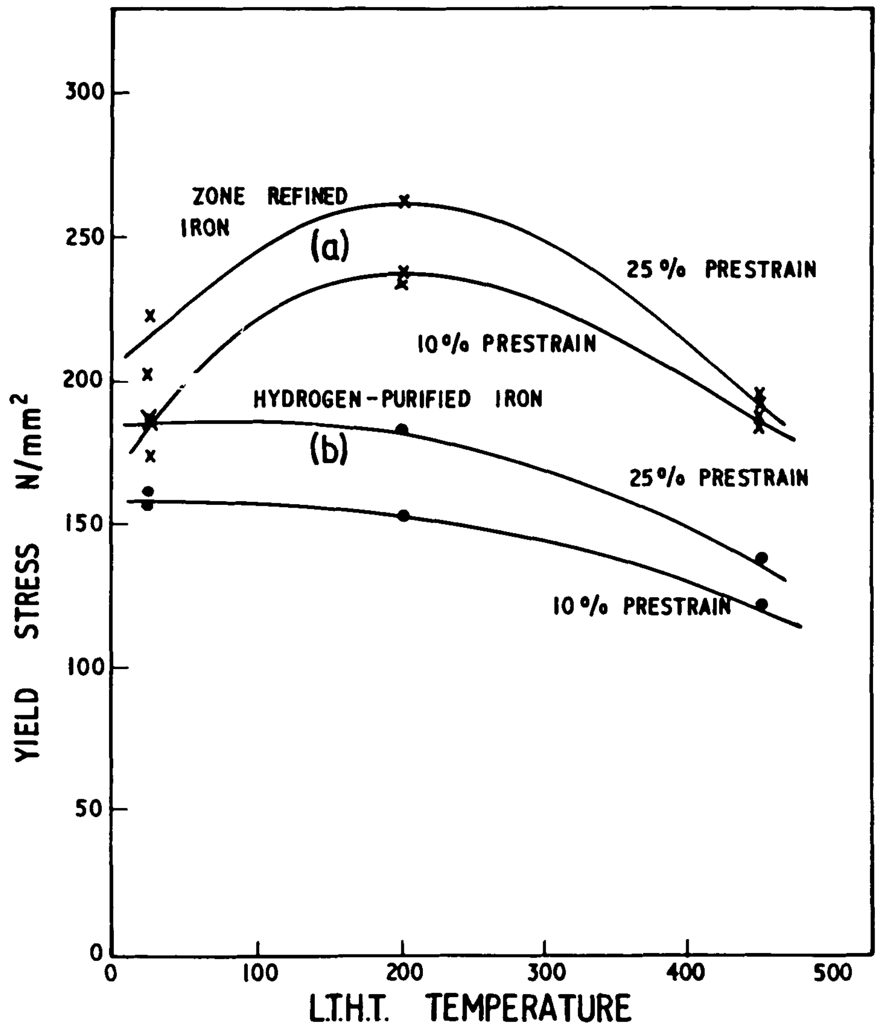


FIG. 18 VARIATION OF TENSILE YIELD STRESS WITH L.T.H.T. AFTER STRAINING OF IRON IN TWO CONDITIONS [23]
(a) ZONE REFINED IRON CONTAINING 0·0019 % CARBON
(b) HYDROGEN PURIFIED IRON, 0·0013 % CARBON

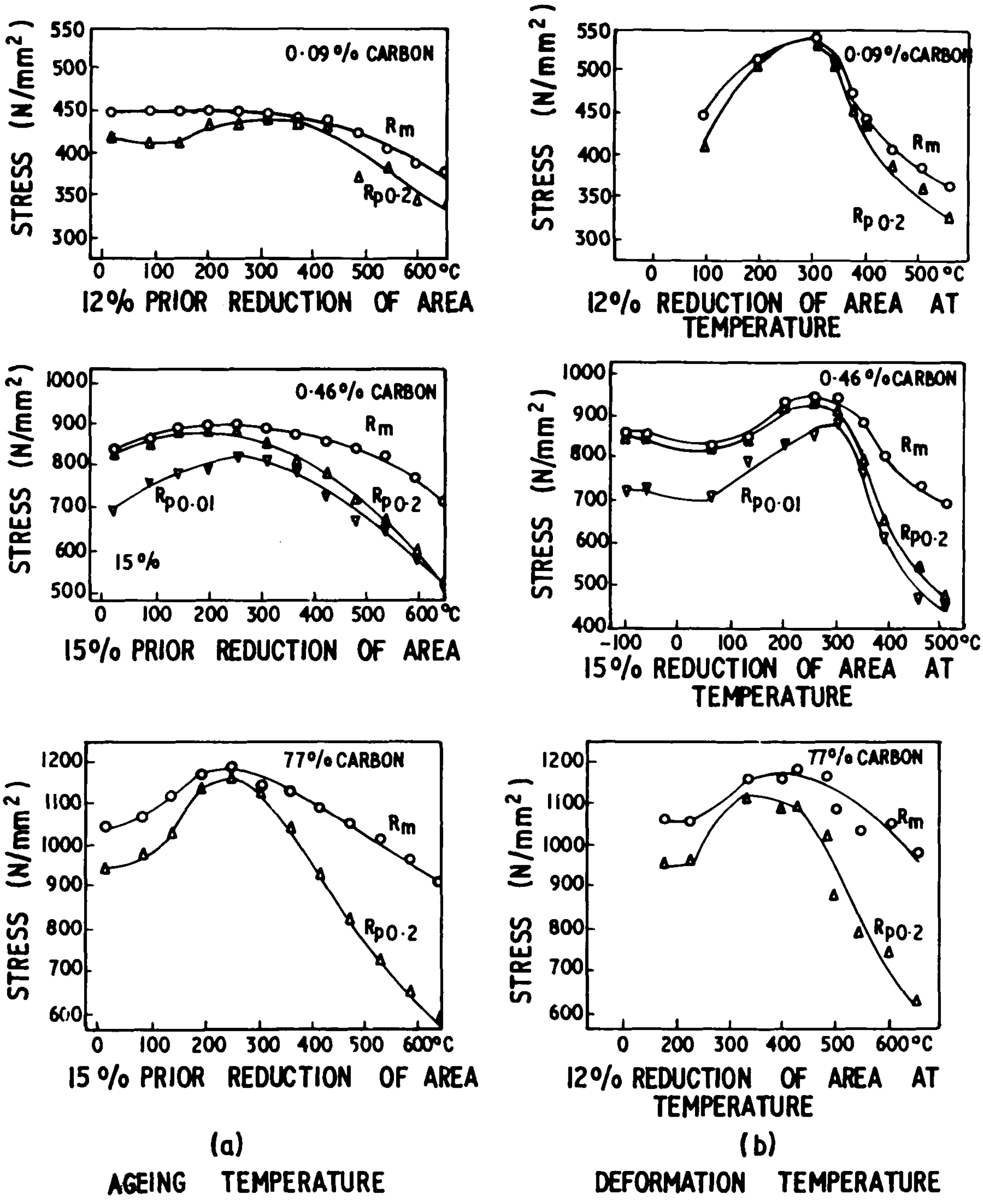


FIG. 19 VARIATION OF PROOF AND MAXIMUM STRESS IN TENSION OF A SERIES OF CARBON STEELS AFTER
(a) COLD WORK FOLLOWED BY L.T.H.T.
(b) HOT WORKING OVER A SERIES OF TEMPERATURES [24]

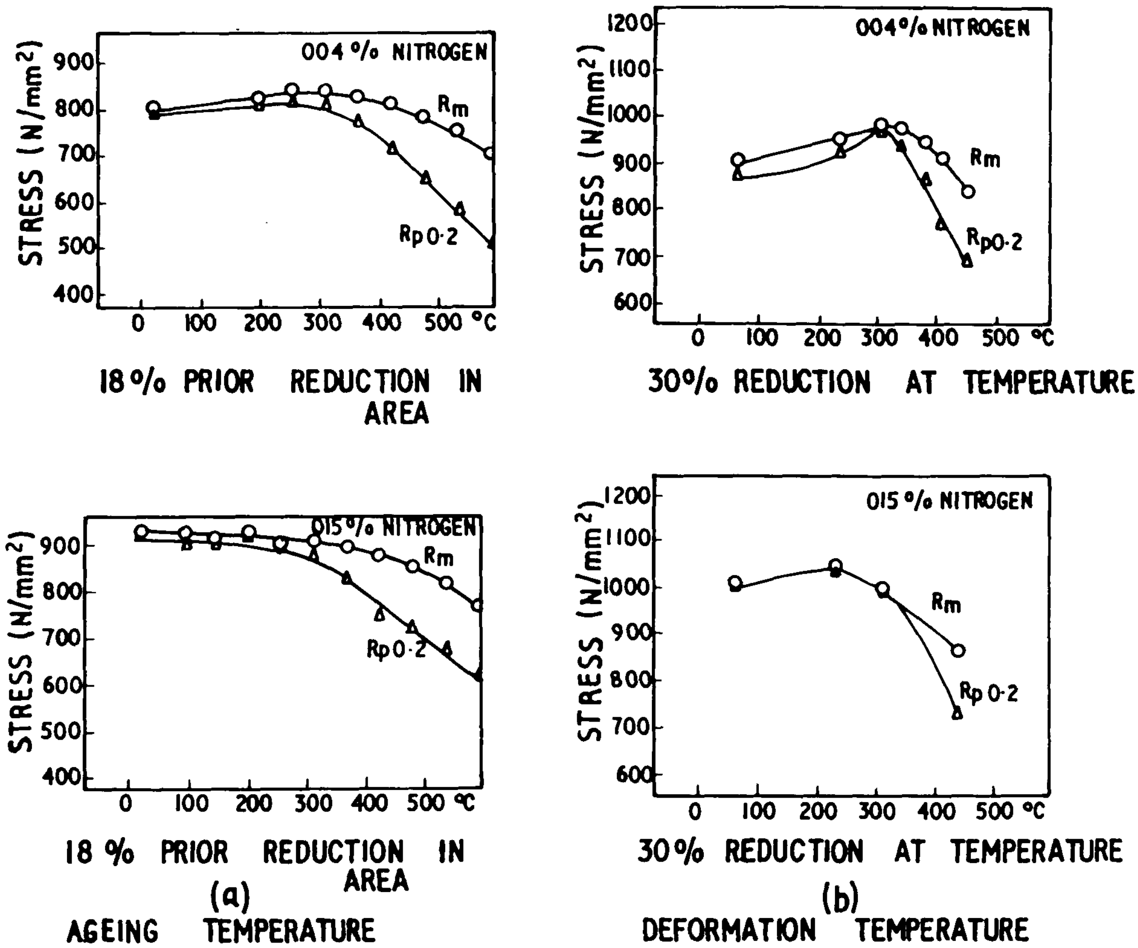


FIG. 20 VARIATION OF TENSILE PROOF AND MAXIMUM STRESS OF 0·46 % CARBON STEELS AT TWO NITROGEN LEVELS AFTER
(a) COLD WORK FOLLOWED BY L.T.H.T.
(b) HOT WORKING OVER A SERIES OF TEMPERATURES. [24]

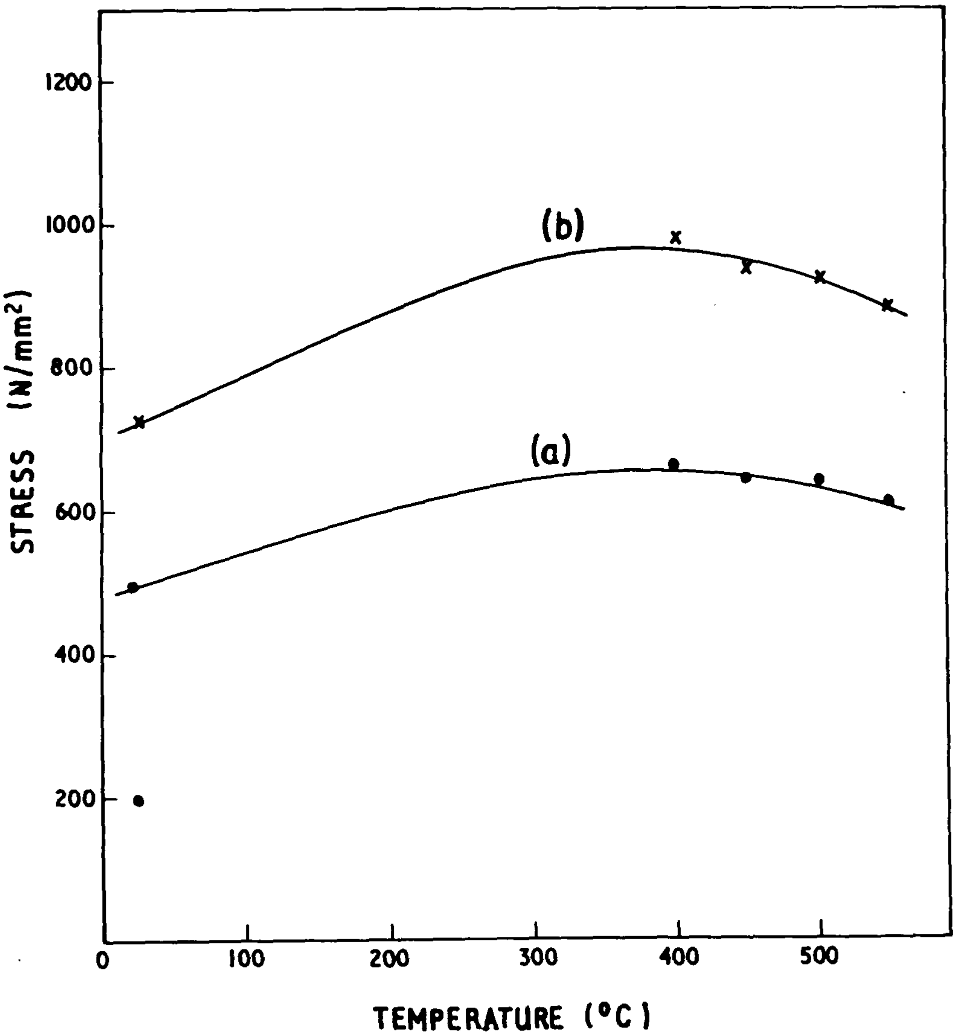


FIG. 21. VARIATIONS OF PROPORTIONAL LIMIT IN TENSION WITH (a) L.T.H.T. TEMPERATURE AND (b) HOT STRESSING TEMPERATURE AT 620 N/mm² (ALL 4 HOUR TREATMENTS) FOR 18 Cr-9 Ni STAINLESS STEEL COLD WORKED 38 % [21]

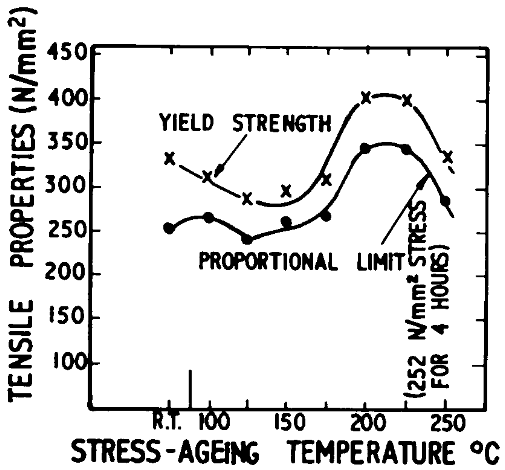


FIG. 22 EFFECT ON PROPORTIONAL LIMIT OF (252 $N/mm^2$) HOT STRESSING ON A HEAT TREATED AL-4% Cu-0·5% Mn-0·5% Mg ALLOY [25]

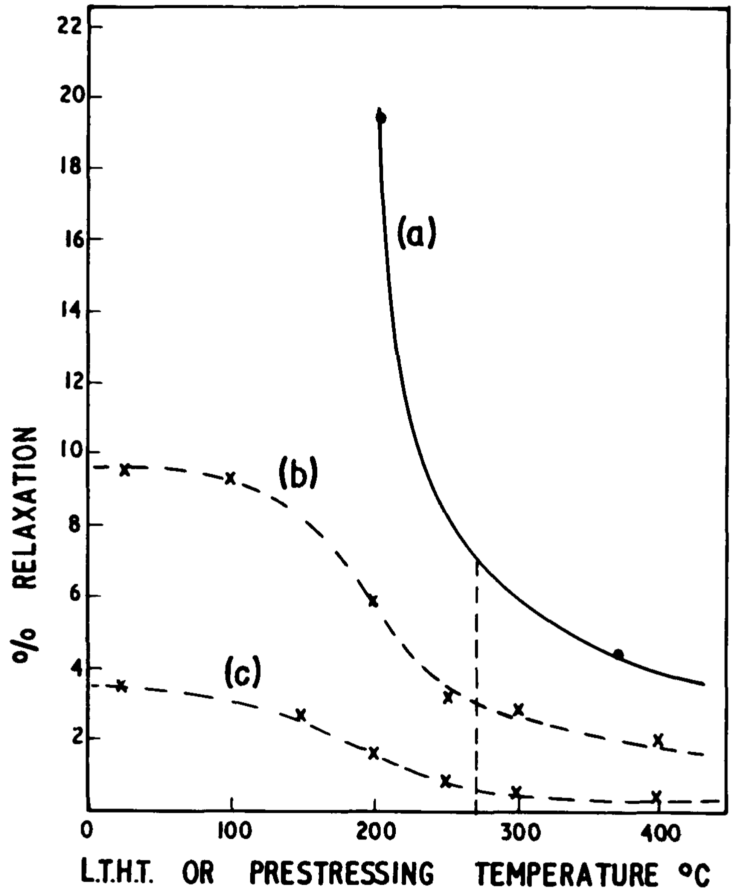


FIG. 23 COMPARISON OF COIL SPRING RELAXATION BEHAVIOUR FOR (a) Cr-V STEEL AT 175°C AND 552 N/mm$^2$ AFTER L.T.H.T.,[29] AND (b) Cr-V STEEL AND (c) Cr-Si STEEL BOTH SHOT PEENED AND TESTED AT 150°C AND 800 N/mm$^2$ FOR 72 HOURS VS TEMPERATURE OF PRESTRESSING [28]

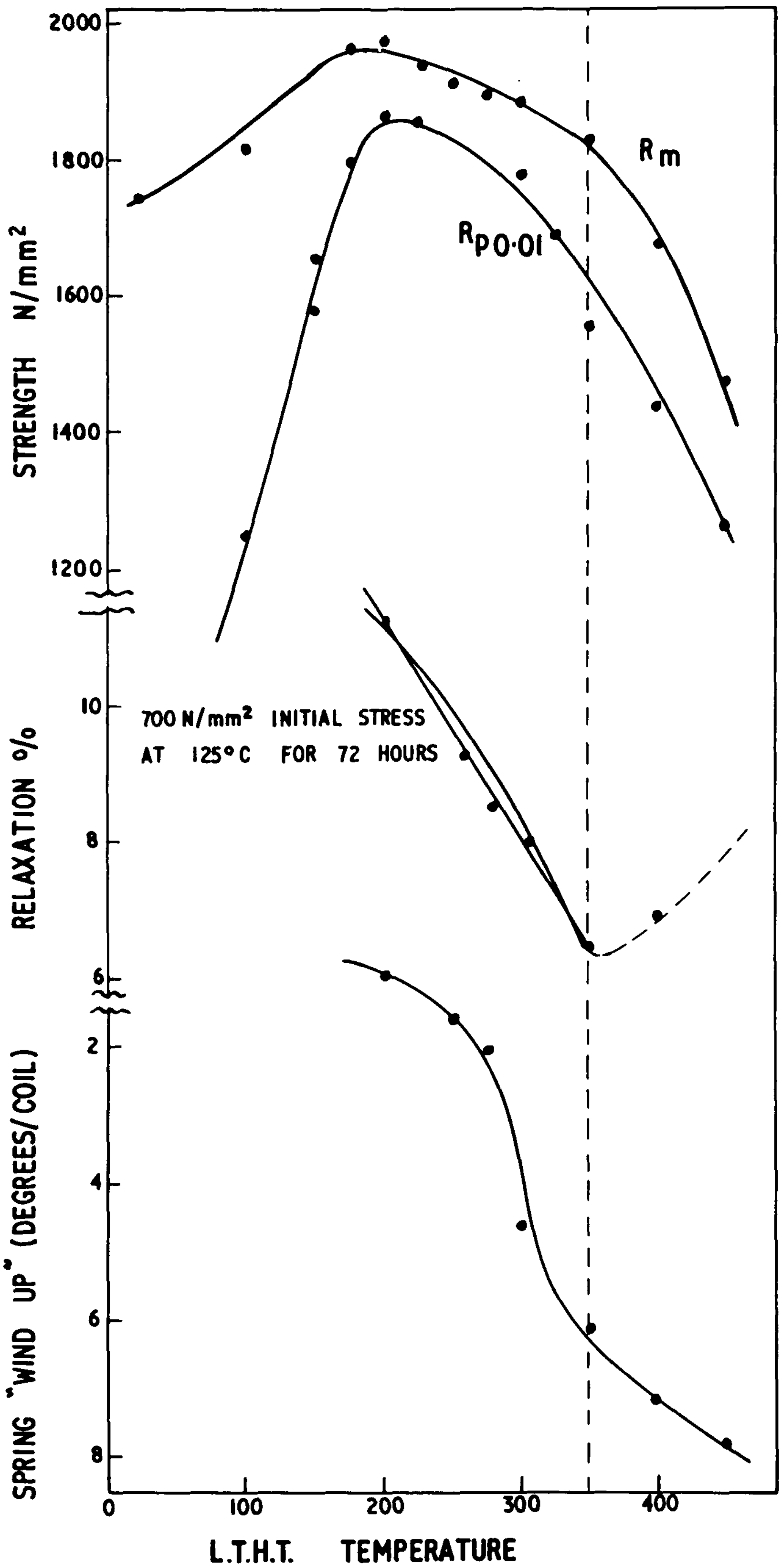


FIG. 24 VARIATION OF TENSILE PROOF AND MAXIMUM STRESS AND COIL SPRING RELAXATION AND "WIND UP" WITH PRIOR L.T.H.T. TEMPERATURE FOR PATENTED COLD DRAWN CARBON STEEL WIRE BS 5216 HD 3 2·65 mm DIAMETER [20]

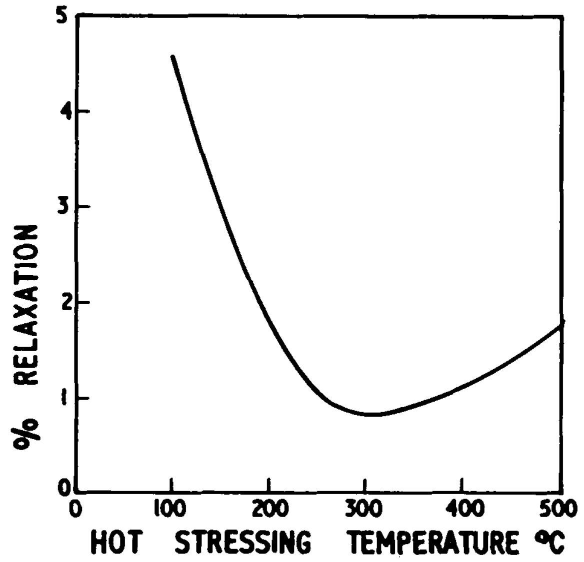


FIG. 25 VARIATION IN TENSILE STRESS RELAXATION BEHAVIOUR AT ROOM TEMPERATURE AND 1130 $N/mm^2$ APPLIED STRESS WITH TEMPERATURE OF PRIOR HOT TENSILE STRESSING TO 1°-2% STRAIN - FOR A PATENTED COLD DRAWN CARBON STEEL WIRE.

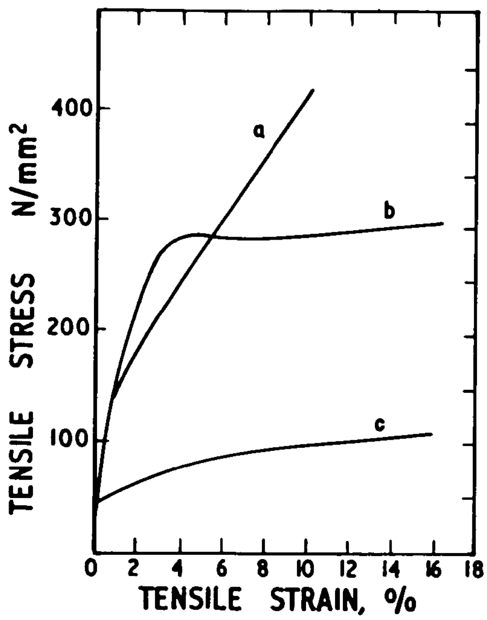


FIG. 26 TENSILE STRESS - STRAIN CURVES OF IRON SINGLE CRYSTALS (a, b) WITH 0·08 wt % NITROGEN CONTENTS (PRECIPITATES OF $Fe_{16}N_2$) AND (c) WITHOUT NITROGEN. (a) WAS AGED AT 100°C AND APPLIED TENSILE STRESS OF 118 $N/mm^2$ FOR 3 HOURS. (b) WAS AGED AT 100°C FOR 9 HOURS WITHOUT APPLIED STRESS.[37]

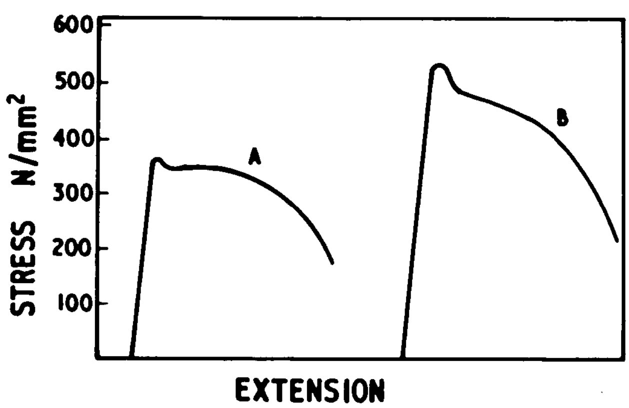


FIG. 27 STRESS/ EXTENSION CURVES AT − 124°C OF UNAGED AND AGED SINGLE CRYSTALS CONTAINING 0·0027 %C: (A) IMMEDIATELY AFTER COOLING; (B) AGED 2 YEARS AT ROOM TEMPERATURE.[30]

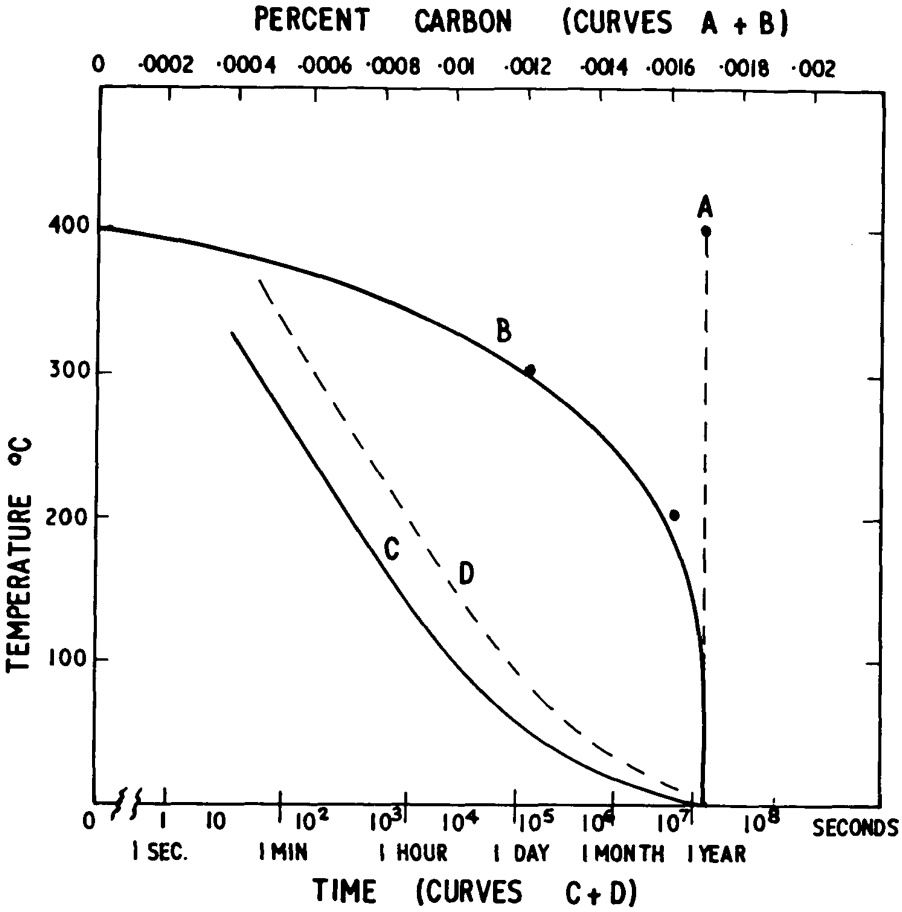


FIG. 28 CURVE A – EQUILIBRIUM SOLID SOLUBILITY OF CARBON IN ALPHA IRON AT 400°C. CURVE B – QUANTITY OF CARBON (A) WHICH WILL PRECIPITATE AT LOWER TEMPERATURES GIVEN SUFFICIENT TIME. CURVE C – TIME TAKEN FOR 50 % OF AVAILABLE QUANTITY OF CARBON (B) TO PRECIPITATE. CURVE D – TIME FOR 95 % OF CARBON TO PRECIPITATE.[30]