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THE ROLE OF DUCTILITY IN HOT WORKING

by

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CoA Note Mat. No. 14

November, 1967

THE COLLEGE OF AERONAUTICS

DEPARTMENT OF MATERIALS

The role of ductility in hot working

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S U M M A R Y

Strength and ductility are the important characteristics which govern the hot working properties of a material. This paper describes test methods for measuring hot workability and then discusses deformation and fracture mechanisms in hot working. Particular emphasis is laid on the correlation of strength and ductility data in simple materials in terms of dependence on strain rate and temperature. Suitable correlations enable the identification of basic parameters controlling deformation and fracture processes. With complex materials, these correlations cannot be applied due to the occurrence of precipitation reactions and the presence of inclusions and second phases having markedly different strength and ductility characteristics.

Paper presented at Seminar on 'Ductility - Limitations, Exploitation, Utilization' organised by the American Society for Metals in Cleveland, Ohio on October 14th and 15th, 1967.

Contents

The two characteristics which determine the forming properties of a material are its resistance to plastic flow (strength) and its ductility. The first determines the size of the equipment needed for the forming operation while the second determines the maximum allowable deformation without risk of fracture. Often the equipment available for hot working is sufficiently overdesigned that it is capable of dealing with most materials from a strength viewpoint and the unknown quantity is the ductility. In practice the limit of ductility is often simply determined by operatives who are governed by economic incentives with respect to throughput of material through say a breaking-down mill. With low cost material, such experience may prove sufficient when dealing with routine hot working procedures but, with high cost materials, such as stainless steels and high temperature alloys, suitable data must be available to decide on the optimum working schedules from a viewpoint of strength, ductility and structure.

The main features of hot working are that extremely large strains are applied to materials at high rates of strain at temperatures above about $0.7T_m$ where T_m is the melting point in degrees Kelvin. Strength and ductility under these conditions are markedly dependent on both temperature and rate of straining, while ductility is intimately related to the deformation processes which govern plastic flow. In this paper, we first discuss test methods for studying hot workability and then consider deformation and fracture mechanisms in hot working in relation to test data and structural observations.

Methods for Studying Hot Workability

The most reliable method is clearly to process material under plant conditions where both the variables inherent in the material, i.e. composition, size, shape, ingot structure, and the variables inherent in the process, i.e. rate of strain, nature of stress system, temperature, are simultaneously covered. However such a method is expensive and the obvious advantages of a laboratory test, e.g. ease of checking different casts, ease of determining optimum conditions for new materials, close control of variables and possibility of relating structure and properties, have led to the development of a number of tests, some of which are simulative in nature and others of which are designed for basic studies

An important feature of hot workability tests is that they should give information on strength and ductility over the ranges of rates of strain, strain and temperature used in practice. Thus, in some processes, e.g. extrusion, the total strain is applied in one operation at roughly constant temperature while in other processes, e.g. forging, the total strain is applied in a number of operations at decreasing temperatures. A good example of these ranges in one process is hot rolling of sheet steel (1). As shown in Fig. 1, the total deformation can be applied either in one operation by a planetary mill or in a series of operations either in a continuous mill or in a reversing mill.

While test data for control purposes have been gathered by a variety of methods, there has been relatively little attempt to relate these data to deformation and fracture mechanisms. Such basic research requires data over ranges of working conditions which are not utilized in practice and more sophisticated test methods have been developed. Thus in the ideal hot working experiment the specimen is deformed uniformly at constant true strain rate and temperature with continuous measurement of stress required for deformation. Rapid cooling is often used to facilitate examination of the hot worked structure and the effect of continuous working operations is simulated by cycles of deformation and rests at temperature.

Various reviews of experimental techniques for studying hot workability are available (2-7) and only a brief discussion is given here of test methods with particular emphasis on the results obtained for strength and ductility. For convenience, we consider three basic methods of stressing - tension, compression and torsion.

Tension Tests. While some workers have used ductility in tensile tests at normal strain rates as a control technique in practical hot working, the most useful technique is clearly to examine strength and ductility in tensile tests at high rates of strain. Fig. 2 shows results obtained on an 18%Cr-12%Ni steel in the as-cast and as-forged condition in an impact tensile test at a rate comparable to that of rolling or press forging (8). Such results clearly bring out the difference between the strength and ductility of the two conditions and can be used to determine optimum hot working ranges.

However the change in dimensions of the test piece during extension means that the strain rate is continually changing unless a controlled rate of cross-head movement is used. A further change in strain rate occurs with the onset of necking which usually begins at a true strain of less than 0.7 which is in general much less than the total strain one wishes to study. This necking strain is however of the same magnitude as the strain occurring during each pass in hot rolling or press forging and from this viewpoint useful data can be obtained from tensile tests. By controlling the rate of cross-head movement during uniform extension and necking, tensile tests at constant true strain rates have been carried out up to true strains of roughly unity as shown in Fig. 3 for an 0.25%Cr steel at 1100°C (9). The form of these curves will be discussed later.

A major disadvantage of tensile testing for study of fracture mechanisms is that the necked region is unsuitable for structural studies since at high strain rates marked adiabatic heating occurs in a small volume leading to marked temperature rises. This effect may however be useful in practice where problems of 'hot shortness' arise due to inhomogeneous deformation during working.

Compression Tests. Since many hot working processes involve essentially compressive stresses, various workers in recent years have developed compression tests for obtaining strength and ductility data at high strain rates. Perhaps the simplest test for hot ductility is that using a flywheel press with a constant reduction in height of a preheated cylinder in one blow. Examination of the specimens then enables the determination of hot working temperatures in terms of the minimum and/or maximum temperature at which upsetting can be accomplished without cracking. Such data can be related to structural or compositional factors as shown in Fig. 4 where the amount of second phase is shown to have a significant effect on the minimum upsetting temperature of a titanium-tin-aluminium alloy with additions (10).

From the viewpoint of strength data, simple axisymmetric compression tests, whether of the drop hammer or press type, suffer from a similar disadvantage to tensile tests in that again change of dimensions of the test piece during compression leads to a variation in strain rate. While the complication of necking is avoided and thus much larger strains are possible in compression than in tension, the friction between platens and specimens leads to 'barrelling' of the specimen with consequent inhomogeneous deformation. This 'barrelling' usually occurs at a true strain of about 0.7 and thus limits the usefulness of the test for high strains. However by eliminating platen/specimen friction and by controlling the rate of compression in relation to the specimen dimensions as in the so-called 'cam plastometer', compression tests at a constant true strain rate can be carried out up to high values of true strains. Fig. 5 shows stress-strain curves for aluminium at various temperatures at a constant true strain rate of 4.4/sec (11).

It is worth noting that plane strain compression has also been used in hot working studies to obtain true strains up to about 5 at a constant true strain rate. However, frictional problems also arise in this type of compression and can limit the usefulness of the technique. An advantage is that the load requirements are smaller than in axisymmetric compression where the increase of area leads to a continual increase in the load required for deformation.

Torsion Tests. A major advantage of torsion for studying hot workability is that a constant high strain rate can be readily maintained up to fracture by twisting at constant high speed since the specimen dimensions can be maintained constant. Since necking and barrelling are not limiting factors, large strains can be applied prior to fracture thus simulating more closely the conditions in practical hot working operations. Ductility can be readily measured by the number of revolutions to failure and results at various temperatures can be used to define optimum hot working conditions as shown in Fig. 6 (12).

However, fracture in torsion does not take place under simple shear since dimensional changes occur at large plastic strains and their restraint by imposing a fixed geometry on the specimen leads to axial tension or compression (13). While the average longitudinal stresses are relatively small compared to the applied shear stress, they are important in determining measured ductility since externally applied longitudinal tension or compression can alter the values obtained for materials exhibiting relatively low ductilities (14, 15). Dragan (14) has recently suggested a method for correcting measured ductilities of steels to obtain a 'true' ductility but the assumptions involved need closer study before the method is applied more widely. In fact the success of hot torsion tests in predicting optimum temperatures for hot piercing of steels probably lies in the complex stress system imposed at fracture.

The torque measured during hot torsion can be converted to shear stress and thence to true stress so that stress-strain curves can be obtained up to very high strains at constant true strain rates. In such calculations, the surface shear stress and shear strain are used since the outer layers make the major contribution to the measured torque. The agreement of tension, compression and torsion data indicates that this procedure is an acceptable one. Clearly, however, stress, strain and strain rate vary with radial position and care must be taken in the interpretation of the structures of samples taken from solid specimens. The use of tubular specimens eliminates these problems but buckling limits the attainment of large strains prior to failure.

The temperature dependence of strength and ductility at a constant strain rate is shown by the stress-strain curves for aluminium and copper derived from hot torsion data (Fig. 7) (16). Comparison of the aluminium curves with those of Fig. 5 indicates the much higher strains that can be obtained in torsion than in compression. Fig. 7 shows that the nature of the stress-strain relation changes at high strains so that a steady stress is required for deformation independent of strain. This behaviour is maintained over a wide range of strain rates as shown in Fig. 8 for an 0.25%Cr steel (17). At low strain rates the curves show oscillations similar to those observed in tensile tests at constant true strain rates, cf. Fig. 3. Such oscillations are observed in other materials e.g. copper and nickel, but are hardly visible in aluminium (Fig. 9) (18). The differing form of the stress-strain curves for various materials will be discussed later.

It is worth noting that, in addition to simulating hot working processes such as extrusion or rolling with a planetary mill where large strains are applied in one operation, hot torsion tests can be used to simulate deformation schedules involved in hot rolling or forging. Using a programmed machine, Rossard and Blain (1) have obtained good correlation between structures developed during hot rolling of steel sheet in practice and laboratory tests. A similar technique has recently been successfully applied to hot rolling of aluminium and its alloys (19, 20).

Deformation Mechanisms in Hot Working.

The strength, and also the ductility, of a material during hot working is governed by the balance between work hardening and dynamic softening processes. If work hardening predominates, strength is high but ductility is low while, if softening predominates, strength is low but ductility is high. To understand ductility in hot working processes we must examine the deformation mechanisms during hot working. An indication of the operative softening process should be obtained from analysis of data on rate of strain, stress and temperature coupled with observations of structural changes as a result of deformation.

Correlation of Hot Strength Data. As noted earlier the significant feature of flow curves under hot working conditions is the steady-state region where flow stress is independent of strain. Various attempts have been made to formulate a relation between strength (σ), temperature (T) and strain rate ($\dot{\epsilon}$) for this region. The most successful attempt appears to be that of Sellars and Tegart (21) who propose the relation

$$\dot{\epsilon} = A(\sinh\alpha\sigma)^{n'} \exp(-Q/RT) \qquad (1)$$

where A, α, n' are temperature-independent constants and Q is an activation energy. This relationship is a development of the one proposed for creep by Garofalo (22).

At low stresses ($\alpha\sigma < 0.8$), eq. (1) reduces to a power relation, similar to that used for creep,

$$\dot{\epsilon} = A'\sigma^{n'} \exp(-Q/RT) \qquad (2)$$

and at high stresses ($\alpha\sigma > 1.2$) to an exponential relation

$$\dot{\epsilon} = A'' \exp(\beta\sigma)\exp(-Q/RT) \qquad (3)$$

The constants α and n' are related by $\beta = \alpha n'$ so that α and n' can be simply determined from experimental data for high and low stresses.

Fig. 10(a)-(c) illustrate the application of the relation to data obtained by hot torsion of an 0.25%C steel. Thus Fig. 10(a) is a log-log plot to test the power relation while Fig. 10(b) is a semilog plot to test the exponential relationship. The resultant sinh plot is shown in Fig. 10(c) and illustrates the temperature independence of n'. Hot working data for a number of materials have been satisfactorily correlated by eq. (1). (21, 23-26). The activation energy, Q, can then be obtained from a plot of log $\dot{\epsilon}$, at constant sinh $\alpha\sigma$, against $1/T$. Values of Q from hot working are given in Table 1 together with reported values for creep, self-diffusion, recovery and recrystallization.

Rearrangement of eq. (3) to the form

$$Z = \exp(^{Q}/RT) = A(\sinh \alpha\sigma)^{n'} \qquad (4)$$

permits correlation of the data for different temperatures on a single straight line as shown by Fig. 11. This type of plot provides a reliable method for interpolating data to obtain values of strength at any temperature or strain rate within the ranges studied.

It is clear from Table 1 that in some cases the activation energy for hot deformation remains unchanged over an extremely wide range of strain rates while in others there is a marked difference between the values for creep and hot working. In the former cases, the constant value of activation energy, close to that for self-diffusion, indicates that recovery is operative over the whole range while in the latter cases the values suggest that the softening process is recovery under creep conditions but may be recrystallization during hot working. The possibility of recrystallization during hot working has been discussed in several recent papers and reviews and a dichotomy of opinion exists at the present time. Thus Stüwe (37, 38) and McQueen (6) stress the importance of dynamic recovery as the softening process in all metals and dismiss recrystallization as a possible process, while Rossard (39) and Sellars and Tegart (21) maintain that recrystallization occurs in lower stacking fault energy materials. In the next two sections, the evidence for the occurrence of both recovery and recrystallization during hot working will be briefly discussed in the light of available information.

Recovery as the Softening Process. In the case of aluminium and α-iron, the activation energy for hot deformation is similar for both creep and hot working. The formation of well developed substructures by recovery by cross slip and climb during creep of aluminium and α-iron is well documented. In the case of hot working, if the high temperature structure is retained by rapid cooling after deformation, the structures observed are similar to those after creep as shown in Figs. 12(a) and 13(a). However due to the high strains imposed during hot working the sub-boundaries are more clearly defined than after creep while the original grain boundaries are heavily distorted and are often no longer apparent.

In both hot torsion and extrusion, subgrains remain approximately equiaxed and constant in size throughout the steady state deformation (23, 24, 40). Further, the misorientation measured in extruded aluminium after strains of 10 and 40 are roughly constant in the range 1-4° and are no larger than those observed after low strains in creep (23, 24). This constancy of subgrain size and misorientation leads to a constant dislocation density and hence a constant flow stress.

For the subgrains to remain equiaxed while the original grains undergo marked shape changes the sub-boundaries must be able to relocate their position. Jonas et al (23) suggest that this is accomplished by the continuous disintegration and reformation of the dislocation arrays at the equilibrium spacing. The disintegration proceeds by the applied stress pushing dislocations out of sub-boundaries, the stability of which has been diminished due to disruption by the passage of mobile dislocations. This process is assisted by a limited amount of sub-boundary migration which is made possible by the enhanced climb of dislocations produced by the deformation. The overall process has been termed 'repolygonization'.

Such a process suggests that the size of the subgrains should be dependent on stress. Various workers have reported that subgrain size (t) during high temperature deformation can be related to stress by a relation of the form

$$\sigma = \sigma_o + kt^{-x} \qquad (5)$$

where x is a constant of value 1 to 1.5. The best value of x is at present uncertain since accurate determination of subgrain size is experimentally difficult. However recent statistical analyses show that for aluminium, high temperature creep data are best fitted by $x = 1$ while hot torsion and hot tensile tests at much faster strain rates are best fitted by $x = 1.5$ (40). To understand this discrepancy, it is necessary to realise that subgrain boundaries even at high temperatures can act as barriers to dislocation movement. Thus Cotner (40) has found that when both grains and subgrains are present in an aluminium specimen before it is hot worked the subgrains determine the strength of the specimen during deformation. The strengthening effect of substructure has been investigated by various workers by deforming specimens at low temperature after substructure has been introduced by high temperature deformation. In this case the value of the constant x in eq. (5) is 0.5 and has been interpreted in terms of pile-ups of dislocations against the sub-boundaries.

In the case of creep, the value of $x = 1$ can be understood from the inverse relation between stress and slip band spacing originally developed by Orowan since here the strains are small and a stable substructure develops from the coarse slip bands produced in the early stages of loading. However in the case of repolygonization where the boundaries are being broken down and reformed, the barrier effect of the sub-boundaries must be important and Cotner has suggested that the simultaneous production of sub-boundaries and pile-ups of dislocations against them leads to the observed value of $x = 1.5$.

From an analysis of extrusion data on aluminium, Jonas et al (23) have reported that the activation energy for repolygonization is 41.8 k cal mole, i.e. higher than that for self-diffusion. However a more recent analysis of their data suggests that a better value is 35 k cal mole in line with other workers (24). Such a value implies that the rate determining step in the repolygonization is dislocation climb as in the case of creep.

The important feature that emerges from these results is that in some materials under hot working conditions dynamic recovery processes can operate extremely rapidly to reduce strain hardening and that large strains can be accommodated by the continuous disintegration and reformation of sub-boundaries to maintain an equiaxed structure.

If the hot worked material is held at the deformation temperature for some time prior to cooling, or if cooling is very slow, then the high temperature structure is not retained and significant structural changes can occur. Initially there is a sharpening of the sub-boundaries associated with gradual softening followed by the appearance of recrystallized grains associated with a more marked softening (Fig. 12). At higher temperatures, the structural changes during isothermal annealing proceed more rapidly and in some cases even the most rapid quenching procedure is insufficient to retain the high temperature structure. Thus at temperatures of 500-600°C the rest periods in commercial hot rolling operations on aluminium are sufficient to restore the fully annealed structure before further deformation, while at lower temperatures, say 400°C, relatively little change occurs (19). Such structural changes are important when considering the ductility of material subjected to cycles of deformation followed by rest periods at temperature.

<u>Recrystallization as the Softening Process.</u> In the case of 18/8 stainless steel, copper, nickel and nickel-iron alloys, the activation energies for the softening process are higher for hot working than for creep. We identify the creep process with recovery and the hot working process with recrystallization. As noted previously, this view is not held by other workers in the field and we will now discuss the experimental evidence for our viewpoint.

For copper and nickel, the activation energies for creep determined from steady state conditions are similar to self-diffusion and are associated with the development of a poorly defined substructure indicating dynamic recovery as the softening process. Further, observations on creep of stainless steels of 18/15 and Type 316 varieties show subgrain formation again indicating dynamic recovery as the rate-controlling process (22).

However, in the case of copper and nickel, recrystallization associated with rapid changes in strain rate can occur during creep. Similar behaviour has been observed with gold and lead and it has been suggested that this behaviour is characteristic of low stacking fault energy materials where dynamic recovery is expected to be slow (41). These observations are not disputed but the main point of dispute is whether such recrystallization can occur during hot working. If dynamic recovery is rate controlling, then clearly the activation energy determined for hot working should be similar to that for creep and self-diffusion while if recrystallization is rate controlling then a different activation energy associated either with boundary migration or grain boundary diffusion should be applicable. If one imagines the rapid changes in strain rate occurring in creep at frequent intervals, then the creep rate determined for the process under a given stress will clearly be different from that for the steady state process. This is the situation that we envisage during hot working, i.e. rapid repeated recrystallization.

Observations on specimens quenched rapidly after hot torsion show that recrystallization occurs in 18/8, copper, nickel and nickel alloys (Fig. 14). Interrupted tests show that the initial deformed grains are replaced by new recrystallized grains nucleated at original grain boundaries and that recrystallization proceeds progressively with strain after the maximum torque (16). Clearly there is a danger that structural alterations produced in the small time interval of quenching could obliterate the true high temperature structure. Thus the new grains often appear to be equiaxed and strain free with straight annealing twins suggesting that they form after deformation. Further, if hot worked material is held at the deformation temperature for some time prior to cooling, recrystallization rapidly ensues. At higher temperatures the recrystallization proceeds extremely rapidly and clearly in some cases the most rapid quenching procedure is insufficient to retain the high temperature structure, as in the case of dynamic recovery discussed previously.

However close examination of the structures of nickel and copper after torsional deformation shows that the grain size of the recrystallized grains increases systematically from surface to centre of the specimen. This variation is consistent with the strain rate gradient which exists during torsional deformation and contrasts with the large grains formed at random in a recovered structure in aluminium (Fig. 12(d)) or commercially pure iron held at temperature after deformation (Fig. 13(b)). Also of interest in this respect is the observation that, in zone-refined iron tested in torsion at 500-900°C, the substructure produced in the early stages of testing is replaced by regular equiaxed grains as the strain increases (42). This change in restoration process from subgrain formation to recrystallization with increase in purity is consistent with the observation that impurities retard both the start of recrystallization and the rate at which it proceeds during annealing after cold work (32).

The size of the recrystallized grains also varies systematically with stress as shown in Fig. 15 where the data indicate that, for copper, nickel and zone refined iron, the grain size is inversely related to stress in an analogous manner to that observed for substructure during dynamic recovery. A similar relation has been observed in hot worked nickel-iron alloys (25). Such a relationship would not be expected for random recrystallization over a variety of quenching times at different temperatures and strain rates but is not unexpected if one considers the model proposed for recrystallization during creep (43).

Here the initial stage of the deformation process is the formation of a poorly developed substructure. These sub-boundaries pin sections of the original grain boundaries which then bulge out and migrate because of a strain energy difference across the boundary. Clearly the extent of such migration determines the resulting grain size and under conditions of concurrent deformation it is reasonable to expect the migration to be restricted by the scale of the substructures which as we have seen is dependent on stress.

A further objection to recrystallization as a softening process is that the rate of recrystallization in hot worked structures would be slow since repolygonization would maintain all boundaries at a similar misorientation and hence preferential migration would be prevented. This is certainly true for isothermal annealing of hot worked structures in aluminium and silicon-iron but under conditions of concurrent deformation the situation can be altered. Thus in the case of nickel there is considerable evidence to show that the time for recrystallization can be accelerated by stress. Fig. 16 shows that for creep of nickel, the time to recrystallization decreases as $(\text{stress})^3$ and that times of < 1 sec might reasonably be expected at the operative stresses during hot working. A similar trend is observed during hot torsion testing of nickel-iron alloys over a wide range of temperatures and strain rates (25).

Strong evidence for a difference in softening process between aluminium and commercially pure irons on one hand and copper, nickel and zone refined iron on the other comes from the form of the torque-revolution curves. Aluminium and commercially pure irons exhibit no initial peak in torque as expected from a gradual development of substructure. In contrast, nickel, copper and zone-refined iron exhibit an initial peak followed by a drop to the steady state region as expected from the more abrupt change in structure associated with recrystallization. If the materials are deformed at low rates of strain the differences become more marked in that the curve for aluminium is fairly smooth while the curves for copper and nickel exhibit marked regular oscillations which can be correlated with significant changes in grain size (Fig. 9). Such behaviour is readily understood if the oscillations are considered as analogous to the changes in strain rate associated with repeated recrystallization during creep. Similarly the torque-revolution curves for Fe-25%Cr and Fe-4%Si which show only substructure formation are smooth at low strain rates while stainless steel shows oscillations (17). It is interesting to note that carbon steels in the γ-range also exhibit marked oscillations at low rates of strain and that for these materials the activation energies for creep and hot working are higher than for self-diffusion. Unfortunately the phase transformation on cooling masks the true γ structure but there are indications that the softening process is recrystallization.

The important feature that emerges from this discussion is that in some materials under hot working conditions recrystallization can operate extremely rapidly to reduce strain hardening and that large strains can be accommodated by repeated recrystallization to maintain an equiaxed structure.

While the distinction between repeated recrystallization and repolygonization may appear to be academic since both lead to similar ends, it is important since the interaction of the softening process and the fracture mechanism determines the ductility during hot working.

Fracture Mechanisms in Hot Working.

In the case of creep, the rupture mechanisms have been closely related to the deformation mechanisms and these are now well understood. In the case of hot working, little has been done to link fracture behaviour closely to deformation mechanisms mainly due to the fact that, until recently, relatively few systematic studies had been reported. Ductility in hot working must be intimately linked to the deformation mechanisms since any process which removes strain hardening will allow greater deformation. Thus ductility, as measured by various tests, increases with increased temperature and with increased strain rate. From our previous discussion both of these factors would be expected to increase the rate of dynamic softening and hence reduce strain hardening. As with strength, a more detailed picture of fracture mechanisms should be obtained from analysis of data on ductility as a function of stress, rate of strain and temperature coupled with observations of fracture modes.

Correlation of Hot Ductility Data. The major difficulty with correlation of hot ductility data is that ductility measured in one type of stress system may not be related to ductility measured in another type of stress system or even the same stress system with different specimen sizes. Attempts have been made to relate ductility to temperature and strain rate, but until recently, ductility data have not been available over a sufficiently wide range of strain rates and temperatures.

Thus an attempt to link ductility to atomic mobility was made by Robbins et al (44) who proposed a relation of the form

$$P = SD^{\frac{1}{2}}f(x) \qquad (6)$$

to describe the high temperature ductility of iron in hot torsion at a constant strain rate. Here P is ductility (revolutions to failure), S is a constant roughly equal to the number of active slip systems, D is self-diffusivity and f(x) a function of factors such as purity, amount of grain boundary sliding, etc. The exponent of $\frac{1}{2}$ was justified in terms of the atomic processes involved in self-diffusion during deformation. Robbins et al presented limited data to support their proposed relation but later work by Reynolds and Tegart (45) on irons of similar purity indicated that the initial agreement was fortuitous. Another attempt to relate ductility to rate of strain has been made by White and Rossard (46) who showed that for a constant temperature there was a power relationship between number of revolutions to failure and strain rate for an iron-25% nickel alloy tested in hot torsion.

In view of the success of the sinh relationship in correlating strength data for both creep and hot working, it is not unreasonable to examine the possible extension of a similar approach to rupture. Thus for short rupture times in creep, the time to rupture (t_r) is inversely related to the steady state creep rate (22, 47) which in turn can be related to the applied stress through the sinh relationship. Thus it is possible to write:-

$$t_r = A'''(\sinh \alpha\sigma)^{-n'} \exp(Q/RT) \tag{7}$$

At low stress levels, this can be approximated by:-

$$t_r = A^{iv}\sigma^{-n'} \exp(Q/RT) \tag{8}$$

while at high stress levels, this reduces to:-

$$t_r = A^{v}e^{-\beta\sigma} \exp(Q/RT) \tag{9}$$

The validity of such an approach has been tested using data on a series of nickel-iron alloys obtained by using hot torsion over a wide range of strain rates and temperatures. Results for a nickel-20% iron alloy are shown in Fig. 17. Thus Fig. 17(a) is a log-log plot to test the power relation while Fig. 17(b) is a semilog plot to test the exponential relationship. The resultant sinh plot is shown in Fig. 17(c) and illustrates the temperature independence of n'. The value of n' obtained from strength data is compared in Table 2 with that obtained from rupture data.

Since the lines are parallel on the sinh plot it is possible to obtain a value of Q from a plot of log t_r, at constant sinh $\alpha\sigma$, against $1/T$. Values calculated in this manner are compared in Table 2 with those calculated from strength data. There is a remarkable degree of agreement in the values computed from the two correlations.

Müller (26) has also applied the proposed rupture correlation to his data on fracture of a series of nickel-chromium stainless steels. The results for the wholly austenitic alloy together with those on two phase alloys consisting mainly of austenite fall in a series of straight lines on a sinh plot but the data for the wholly ferritic alloy give a poorer correlation, presumably due to a different fracture mode as discussed later. In the case of the austenitic alloy, the n' and Q values calculated from rupture data are in good agreement with those calculated from strength data.

As in the case of strength data, the use of a temperature compensated parameter

$$Z = t_r \exp(-Q/RT) = A'''(\sinh \alpha\sigma)^{-n} \tag{10}$$

permits correlation of data for different temperatures on a single straight line as shown in Fig. 18. For pure nickel, creep rupture data can be correlated with hot torsion rupture data in a similar fashion to strength. This confirms the close relationship between high temperature creep and hot working suggested by the strength correlation (21). From a practical viewpoint such a rupture correlation could be useful in predicting rupture behaviour for different stresses, and hence, from the earlier correlation, for different temperatures and strain rates.

As in the case of creep, eq. (7) - (9) do not provide specific knowledge concerning the factors which govern the nucleation and growth of cavities during hot working. They do indicate however that nucleation and growth of cavities is closely related to deformation processes during hot working.

In the case of nickel, the values of activation energy obtained from strength and rupture data are lower than that for lattice diffusion but are similar to those for grain boundary migration (Table 1). The latter values are very dependent on purity and Detert and Dressler (48) have recently reported a value of 30 k cal mole for grain boundary migration in zone-refined nickel. However, while the value of 54 k cal mole is lower than that for lattice diffusion, it is in the range of values reported for grain boundary diffusion (49). Thus the activation energy for grain boundary diffusion is independent of the misfit angle from 20° to 70° with a value of 26 ± 1.5 k cal mole while beyond these limits the activation energy increases smoothly to the activation energy for lattice self-diffusion.

These results on nickel during hot working can be compared with those of Hull and Rimmer (50) on creep rupture of copper where the value of the activation energy for rupture (25 ± k cal mole) was also appreciably lower than that for self-diffusion but close to that expected for grain boundary diffusion (~ 20 k cal mole). A value of 29 k cal mole was also obtained by Boettner and Robertson (51) for growth of voids in copper. On this basis, grain boundary diffusion of vacancies has been identified as the process controlling intergranular void growth in creep rupture.

Further support for this viewpoint comes from the work of Ratcliffe and Greenwood (52) who showed that cavity development could be eliminated by the application of a hydrostatic pressure equal to the creep stress. Other contributions to the growth process, e.g. by plastic deformation processes, were ruled out by the further observation that in severely cavitated specimens no void growth occurred when a hydrostatic pressure equal to the creep stress was applied although the material continued to creep at a rate unaffected by the pressure.

The importance of a hydrostatic component of stress in minimising cracking is familiar to those engaged in hot working of metals. However relatively few systematic studies are available. Thus Dieter et al (15) have shown that during hot torsion of Inconel 600 at 650-870°C the application of a longitudinal compressive stress equal to 50% of the yield stress increased the strain to fracture by a factor of 8-10. In contrast the application of a corresponding tensile stress only reduced the fracture strain by roughly one-third.

While such effects can be understood in terms of a rupture mechanism based on grain boundary diffusion of vacancies as in cavitation creep, there is considerable evidence to show that triple point cracking occurs extensively during hot working, as discussed in the next section. Further, the success of the proposed correlation for both single phase and two phase alloys where interphase cracking is known to occur (26) suggests that other processes may control crack growth. It is thus significant that recent work of Waddington and Williams (53) shows that, for an aluminium-20% zinc alloy which fractures in tensile creep by triple-point cracking, the application of a hydrostatic pressure equal to the creep stress increases both fracture life and total elongation at fracture. Since triple point cracks nucleate as a consequence of stress concentrations produced by shearing along grain boundaries, nucleation should be unaffected by the application of hydrostatic pressure equal to the creep stress. Thus the results of Waddington and Williams clearly indicate that ductility is altered because the crack growth rate is reduced under hydrostatic pressure. A rupture mechanism based on propagation of triple point cracks controlled by grain boundary migration thus offers an alternative explanation for the effects of hydrostatic stresses in hot working.

Observation of Fracture Modes. In the case of creep-rupture, numerous observations indicate the occurrence of apparently two types of fracture. The first are wedge, or w-type, cracks which are associated with triple points. Rupture occurs by the joining of fissures generated by these cracks. The second are round or elliptical, or r-type, voids which are associated with grain boundaries transverse to the applied tensile stress. Rupture occurs by the coalescence of a number of such voids. In the case of creep of Nimonic 90, McLean (54) has related the occurrence of these fracture modes to stress and has shown that w-type cracks are not observed below a tensile stress of 5 tons/in^2 at temperatures between 750 and 950°C.

However, the problem of nucleation of cracks has recently been discussed in some detail by Smith and Barnby (55), and they point out some corrections and amendments to the previously accepted models for triple point cracking due to grain boundary sliding as based on the work of Stroh (56). The nucleation stress, τ, in the modified form of Stroh's analysis is

$$\tau = \left[\frac{3\pi\gamma_B G}{8(1-\nu)L}\right]^{\frac{1}{2}} \tag{10}$$

where γ_B is surface energy of grain boundary,

G is shear modulus,

ν is Poisson's ratio,

L is sliding distance (assumed equal to grain diameter),

and predicts nucleation to be appreciably easier than the original form of the equation used, for example, by McLean (54). For $\nu = 0.3$, this becomes:

$$\tau = \left[\frac{1.7\gamma_B G}{L}\right]^{\frac{1}{2}}, \tag{11}$$

Smith and Barnby (56) have considered a detailed model of nucleation in terms of the interaction between two orthogonal dislocation pile-ups and develop a fracture nucleation criterion of

$$\tau \sim \frac{\gamma}{100 b_E} \qquad (12)$$

where b_E is a magnitude of Burgers vector of edge components of the dislocations. This is a lower stress than that deduced from the Stroh relation with a reasonable assumption of the number of dislocations involved in this pile-up mechanism, namely

$$\tau \sim \frac{\gamma}{13 b_E}. \qquad (13)$$

They conclude that the procedure of correlating experimental results on the appearance of triple point cracks with the original Stroh relation is invalid since nucleation would be expected to occur at much lower stresses. It is suggested that the observed stresses correspond to those required to make cracks grow to an observable length since very small cracks can form at stresses much lower than those predicted by Stroh's original formula.

The case of w-type cracking corresponds to a situation where high stresses generated by dislocation pile-ups are supported by an infinite amount of material. In contrast, r-type voids correspond to a situation where the stresses are supported by finite quantities of material, e.g. grain boundary ledges or precipitate particles. Smith and Barnby (57) have considered crack nucleation in such cases and show that, if local stresses generated by grain boundary sliding are not relaxed prior to the onset of cracking, then particles of width 2c at an interparticle spacing of 2d in a boundary can nucleate cracks when:

$$\tau = \frac{\pi}{2}\left(\frac{c}{d}\right)^{\frac{1}{2}}\left[\frac{4\gamma G}{(1-\nu)d}\right]^{\frac{1}{2}} \qquad (c \ll d) \qquad (14)$$

For a given situation, this expression predicts that nucleation will occur as a much smaller stress than that from Stroh's analysis using 2d as the sliding distance in place of L. However in such cases it may not be possible for subsequent grain boundary sliding to enlarge the nucleated void up to the critical size required for vacancy growth.

These recent treatments show clearly that the stresses required for crack nucleation are much smaller than previously believed. Further they suggest that the difference between w and r-type voids is more apparent than real since cracks can form readily at both triple points and along grain boundaries due to stresses generated by grain boundary sliding. Their rate of growth will then determine their subsequent size and shape.

These suggestions are confirmed by recent work. Thus Williams (58) studying the propagation of triple point cracks in an aluminium-20% zinc alloy found that, although crack growth was sensitive to stress, all the fracture times could be correlated with stress according to eq. (9). Moreover, at low stresses in addition to triple point cracks at grain boundary junctions, cavities or secondary cracks formed at serrations in the grain boundaries. During growth the larger triple point cracks linked up with these secondary cracks. Crack growth was slow and in most cases was proportional to the amount of grain boundary sliding entering the crack. As a consequence the growth rate of wedge cracks was dependent on the angle between the growing crack and the stress axis, increasing by an order of magnitude as the angle changed from 90° to 50-60°.

Taplin and Wingrove (59) in a study of intergranular failure of iron by electron microscopy using replica techniques found that over a wide range of conditions triple point and grain boundary cracks were developed simultaneously. The subsequent growth of these depended on both temperature and strain rate with mechanical tearing predominating at low temperatures and high strain rates and with diffusional processes predominating at high temperatures and low strain rates.

Returning to the case of fracture of nickel under hot working conditions we can make an estimate of the stresses for w-type cracking using the value of 700 ergs/cm^2 for the grain boundary energy of nickel at 900°C as suggested by Inman and Tipler (60). Thus from eq. (12) $\tau = 2.8 \times 10^8$ dyne/cm^2 $\sim$ 4000 psi while from eq. (11) with $L = 10^{-2}$ cm and $G = 4 \times 10^{11}$ dynes/cm^2 (a value appropriate to 900-1000°C), $\tau = 2.2 \times 10^8$ dynes/cm^2 $\sim$ 3100 psi, a somewhat surprising result in view of Smith and Barnby's conclusions. No quantitative information on the variation of grain boundary energy with composition is available but if it is assumed that γ_B for the 20% alloy is roughly half of that for pure nickel (a not unreasonable assumption in view of reported reductions of grain boundary energy due to impurities), then from eq. (12) $\tau \sim 2000$ psi while from eq. (11), $\tau \sim 3100$ psi. (In the case of the alloys, the grain size is finer and a value of 5×10^{-3} cm is more appropriate). Although Smith and Barnby's criterion gives a smaller stress in this case, the difference is not as marked as they suggest. However such stresses are readily reached in tests at high temperatures even at low strain rates and we expect both triple point and grain boundary cracks in hot worked materials.

While various workers have reported some details of fracture modes during studies of hot working, relatively few systematic studies have been reported. However the fracture behaviour of several pure irons has been extensively studied by Reynolds and Tegart (45) and Keane et al (42) over the range 700-1250°C at strain rates of 0.8 and 8 sec^{-1} using hot torsion. In the α region where repolygonization is believed to be the softening process, although small voids associated with original grain boundaries appear in the specimens, final fracture occurs by propagation of cracks into the specimen from surface irregularities. In the γ region where recrystallization is believed to be the softening process, extensive intergranular cracking occurred at temperatures low in the γ range and failure occurred by linking up of these cracks. At higher temperatures in the γ range, intergranular cracking was reduced and extensive small voids were observed. Failure in these cases occurred by linking of these voids by ductile fracture to give large internal cavities. In the temperature range of the $\alpha - \gamma$ transformation, where two phase structures were observed, failure was associated with internal voids developed by interphase cracking.

Recently a systematic study of fracture modes in an iron-25% nickel alloy, where recrystallization is believed to be the softening process, was carried out over a range of strain rates and temperatures by White and Rossard (46). In specimens that failed at low ductilities, they showed the presence of cracks at both triple points and at grain boundary irregularities in the initial grain boundaries (Fig. 19). These then linked up to give intergranular fracture (Fig. 20(a)). Although the triple point cracks were often larger than the grain boundary cracks, the differences appear to be associated with the relative amounts of grain boundary sliding in the two cases rather than with a difference in mechanism in agreement with earlier discussion. In specimens that failed at high ductilities, the initial cracks were isolated from the original grain boundaries which disappeared due to recrystallization (Fig. 20(b)). New cracks formed in the boundaries of the new grains. These observations are strikingly similar to those of Williams (57) discussed earlier. During further work on an iron-25% chromium alloy, where repolygonization is believed to be the softening process, White (61) observed that fracture resulted from the formation of very large holes in the interior of the specimen with a distinct absence of marked intergranular cracking.

These observations suggest that while cracks can be readily nucleated during high temperature deformation, their subsequent propagation and the final mode of failure is dependent on the interaction between deformation and fracture mechanisms.

The Interaction between Deformation and Fracture Mechanisms in Hot Working

The fracture mechanisms operating during hot working are similar to those operating during creep yet very much larger strains can be imposed prior to fracture in hot working than in creep where failure occurs after a few per cent elongation. While the presence of compressive stresses will retard the growth of cracks, there must clearly be an interaction between deformation and fracture mechanisms which further slows down the growth of cracks during hot working permitting large strains prior to failure.

Such an interaction has been briefly discussed by Davies and Wilshire (62) when considering creep results on various grades of nickel and a nickel alloy. They suggest that whenever recovery processes in the region of the grain boundaries cannot take place, low ductilities and a high crack incidence will result. When either complete recrystallization in the necking region or grain boundary migration can occur, ductile fractures are obtained, with little or no intercrystalline cracking. Similarly Harris (63) suggests that recrystallization in grain boundary regions leads to enhanced tensile ductility of magnesium during high temperature deformation.

Rhines and Wray (64) observed a minimum in tensile elongation at an intermediate temperature region in a number of face-centred cubic metals and alloys and suggested that this resulted from marked grain boundary sliding on the original grain boundaries below the temperature range for rapid recrystallization which could wipe out the original boundaries and thus reduce grain boundary sliding. A similar minimum around 760°C was observed during hot torsion studies on Inconel 600 by Dieter et al (15) while recrystallization in this material was not observed till 50-60°C higher.

Detailed studies of interaction under hot working conditions have been made recently by White and Rossard (46) using hot torsion of an iron-25% nickel alloy in which recrystallization is the softening process, and by White (61) using hot torsion of an iron-25% chromium alloy in which repolygonization is the softening process. While some of the observations are specifically related to the stress system operating during hot torsion (see 'Methods for Studying Hot Workability'), the ideas are applicable to hot working generally. We consider two cases of importance in practice namely continuous and interrupted deformation.

Continuous Deformation. The suggested model is that the initial stress developed for an applied strain rate is sufficient to form cracks at the original grain boundaries both at triple points and at irregularities developed in the boundary. These cracks can then grow under the combined action of vacancy diffusion along grain boundaries and applied tensile stresses. (In hot torsion these are generated by restraint of the specimen but in many working processes secondary tensile stresses are operative). Conditions of low ductility correspond to the propagation and coalescence of these cracks to give intergranular rupture along essentially the original grain boundaries.

In the case of materials where recrystallization is the softening process, conditions of high ductility correspond to the migration or recrystallization of the original grain boundaries, thus isolating the initial cracks and preventing further immediate growth. Further growth occurs by 'capturing' a moving grain boundary for a sufficient time for vacancy diffusion and the applied tensile stress to lengthen the crack a little before the boundary again breaks away. New cracks may form in the boundaries of the recrystallized grains and these would then proceed to grow in a similar manner to the initial cracks. The controlling process in crack propagation is thus grain boundary migration.

In the case of materials where repolygonization is the softening process, although the original boundaries tend to lose their identities, the migrating sub-boundaries would not be expected to sweep through the material in the same way as grain boundaries. Further, the measured average low angles of misorientation of the sub-boundaries would not favour vacancy diffusion along boundaries over vacancy diffusion through the lattice and hence growth by this process would be expected to be slow. However, in localised zones adjacent to the original grain boundaries, high misorientations can be achieved and further cracking can occur at these new triple points and boundary serrations. These can link up under the applied stress giving a serrated, cracked boundary region as observed by Williams (58) during creep. The controlling process in crack propagation is thus repolygonization which acts to maintain all sub-boundaries at the same small average misorientation.

During cold working, when the applied stress is high, these materials are not significantly more ductile than those which recrystallize, but during hot working the situation is markedly altered and materials exhibiting repolygonization are inherently more ductile. This viewpoint is supported by some hot torsion results on various materials of similar initial grain size at the same value of $^T/T_m = 0.7$ (13). It was found that copper, nickel and α-iron (0.2%C) failed at shear strains of 34, 23 and 34 respectively while an iron-$3^1/_4$% silicon alloy, an iron-25% chromium alloy and aluminium failed at shear strains of 182, 57 and 87 respectively. The work of White (61) on an iron-25% chromium alloy leads him to conclude that, in the absence of inclusions, the ductility of this material is 'quasi-infinite'.

The nature of the softening process is thus seen to be the major factor in determining the ductility of pure metals and alloys during continuous deformation.

Interrupted Deformation. In many hot working processes, the workpiece is not subjected to a continuously applied deformation but instead undergoes a series of deformations separated by various times of holding at temperatures. During these holding periods, structural changes can occur and we would expect these to be reflected in changes in ductility. Relatively little attention has been paid to this aspect of hot working. Thus White and Rossard (46) have reported results for an iron-25% nickel alloy under programmed hot torsion while Cotner (40) has studied a series of aluminium-magnesium alloys using a similar technique.

In the case of the iron-nickel alloy, White and Rossard carried out a series of tests at 1100°C at a strain rate of 2 sec^{-1}, using shear strain increments of 3.85 separated by holding times of 5 seconds, 1 min. or 10 min. For continuous deformation the specimen failed after a shear strain of 44, while in interrupted tests with the above holding times, the specimens failed after shear strains of 38, 34 and 54 respectively. In all these tests, the shear strain increment of 3.85 had already been shown to form grain boundary cracks and a further test was made in which the deformation was applied in shear strain increments of 0.77 separated by holding times of 1 min. In this case the specimen failed after a shear strain of 119.

While these total strains are clearly greater than any imposed in practice, the results indicate the marked effect that deformation schedules can have on ductility. Thus ductility is improved either by applying a large strain coupled with a large time interval or a small strain coupled with a small time interval. In the former case the conditions are such that the original grain structure is replaced by a completely new structure and growth of the initial large cracks is inhibited. In the latter case the initial cracks are very small and although only partial recrystallization occurs the growth of the cracks is still inhibited. However a situation can arise where the initial cracks are large and the amount of recrystallization is insufficient to remove the initial structure. Further deformation cycles of a similar type can then lead to more rapid failure than in the continuous case.

In the case of super-pure aluminium and aluminium-magnesium alloys, Cotner (40), carried out a series of tests over the temperature range from 150°C to 600°C at a strain rate of 0.86 sec^{-1}. The main features of the results are shown by the data in Table 3 on the 2% magnesium alloy.

Thus below 350-400°C, programmed deformation leads to greater ductility, the greatest ductility being obtained with the longest rest period. However, above 380°-400°C, programmed deformation leads to lower ductility, the lowest ductility being obtained with the longest rest period. This behaviour can be understood in terms of the competition between recovery and recrystallization in this material. As pointed out earlier for pure aluminium isothermal annealing studies of hot worked material show that there is initially static recovery characterised by a sharpening of the subgrains with gradual softening followed by recrystallization associated with a more marked softening. For the aluminium-2% magnesium alloy, Cotner found that after a shear strain of 1.27, the times to roughly 50% recrystallization decreased rapidly with increased temperature as 390 secs at 400°C, 42 secs at 450°C and 7 secs at 500°C.

Thus, at the lower temperatures, appreciable static recovery during the rest periods allows migration of the sub-boundaries in the region of the original grain boundaries so that any initial cracks are isolated and their propagation is slowed down. This process would be enhanced by increased time allowing greater migration. About the critical range, the time between rest periods is such that marked migration and rearrangement occurs allowing some coalescence to give an increased subgrain size and greater angles of misorientation. With only a slight increase in temperature this structure changes to one consisting of subgrains having such high misorientation that cracks can develop in a similar fashion to those developed in the original grain boundaries. With further increase in temperature, recrystallization occurs to give a completely new grain structure which on further deformation develops further cracks. These will eventually link up to give failure but as the results show, ductility under interrupted deformation is relatively insensitive to temperature above 400°C. The increased rest period presumably leads to grain coarsening and hence a greater propensity to cracking.

Thus again the nature of the softening process is seen to be the major factor in determining ductility of pure metals and alloys during interrupted deformation. Further systematic studies are clearly necessary to improve our understanding of the fracture behaviour of materials in practical hot working processes.

Effect of Inclusions, Impurities and Structural Inhomogeneities on Ductility

The discussion of the previous section has been confined to pure metals and some homogeneous single phase alloys. While such materials can give basic information on deformation and fracture mechanisms, they constitute a small fraction of the tonnage of hot worked products and the majority of materials contain inclusions, impurities and structural inhomogeneities. While the effects of some of these can be understood in terms of the basic ideas presented earlier, in large sections of complex alloys, particularly in the cast state, inhomogeneity of structure leads to localised situations which are not representative of the bulk material. Such situations which give rise to many of the problems associated with hot working operations e.g. hot shortness associated with local melting of a segregated phase or preferential cracking due to interdendritic segregation or massive carbide phases, usually cannot be predicted from laboratory tests.

Much of the published information relates to steels and an excellent review of factors affecting hot workability has been given by Nicholson (3). In this final section we consider briefly the influence of solid solution alloying, second phases, cast structure and inclusions on hot ductility.

Solid Solution Alloying. The general effect of solid solution alloying is to increase strength and decrease ductility during hot working. However the effect is more pronounced in materials where recrystallization is the softening process. This is not unexpected in view of the more marked effect of alloying on recrystallization than on recovery during annealing after deformation at low temperatures.

Thus copper-nickel alloys under hot torsion show a marked ductility trough above about 20 per cent alloying element with either copper added to nickel or nickel added to copper (21). Clearly the retardation of recrystallization means that substantial crack propagation can occur under the higher stress required to deform the alloys before recrystallization operates to slow down crack propagation. In the case of the concentrated alloys, failure in fact occurs along the original grain boundaries. Similarly, White (61) has found with an iron-25% nickel alloy that the addition of 0.005% boron increases ductility while the addition of 0.15% copper or 0.10% tin decreases ductility. These trends are similar to the reported effects of alloying on recrystallization of iron after deformation at low temperatures (32).

Hot torsion tests of aluminium-magnesium alloys (40) show that, although the ductility is decreased by the addition of magnesium, the most marked decrease only occurs with alloys near the solubility limit. This effect must arise from the influence of alloying on the strength of the sub-boundaries since, for the same subgrain size, the aluminium-5% magnesium alloy has a flow stress roughly 7 times that of pure aluminium. Such strengthening could result in a resistance to migration and to repolygonization thus allowing greater initial crack development. Further, in contrast to his results with the iron-25% nickel alloys, White (61) found that the addition of boron, copper and tin had little if any effect on ductility of an iron-25% chromium alloy. Unfortunately the effects of alloying in altering the rate of recovery in iron are not yet clear (32).

Duplex Alloys. The presence of a duplex structure in an alloy normally leads to reduced ductility. A good example is the low ductility of low carbon steels when tested in the ($\alpha+\gamma$) range (44, 45). The two phase structure is much less ductile than either of the component phases due to the inhomogeneity of deformation in the two phases. If coherency between the phases is good then fracture can occur in either phase depending on the proportions of the phases and their relative deformations. If coherency between the phases is poor, due possibly to preferential segregation of impurities, then interphase cracking occurs. In the case of iron, the so-called 'red shortness' around the transformation range has been explained by the presence of an extended ($\alpha+\gamma$) region produced by the addition of impurity elements rather than by segregation leading to a low melting film at the grain boundaries of ferrite (45).

Detailed studies of hot working of two phase alloys consisting of austenite + δ-ferrite in a series of wrought chromium-nickel steels have been made by Castro and Poussardin (65) using tension and Müller (26) using torsion. As shown in Fig. 21 there is a ductility trough at about 30% ferrite over a range of temperatures. Fracture is initiated by interphase cracking which is particularly severe at about 30% ferrite where most of the deformation is concentrated in the austenite. As noted earlier, the fracture data on these alloys can be correlated using eq. (7) indicating that rate of crack propagation rather than the mode of crack initiation is the important variable in determining ductility.

The influence of composition on ductility can often be rationalised in terms of the proportion of second phase present in a given material. Thus, for the titanium-12.5% tin-2.5% aluminium alloys used by Reynolds (10) which had a composition close to the β-phase boundary and a minimum upsetting temperature of 1100°C, the addition of small amounts of alloying elements had marked effects on the hot workability. Thus additions of carbon, lead, silicon and zirconium raised the minimum upsetting temperature by 100°C whereas the remaining additions either had no effect in spite of increased solute content or reduced the minimum upsetting by up to 100°C. However their individual effects could be rationalised in terms of the proportion of β-phase present in the structure as shown in Fig. 4.

The degree of dispersion of the second phase is clearly an important factor and there is some evidence that ductility of brittle hexagonal materials can be improved by having a dispersion of fine coherent particles which restrain grain growth and restrict crack propagation (66).

Cast Structures. Most of the experimental results on hot workability have been obtained on wrought materials, but in practice, the major problems in hot working arise during the breakdown of the as-cast structure. There are formidable experimental difficulties associated with the difficulty of relating laboratory casts to practical conditions where the different scale leads to different segregation and structural features and of reproductibility of results, particularly in situations of limited ductility, associated with sampling problems.

Recently Decroix et al (8) have examined the hot workability of a series of nickel-chromium stainless steels cast in 25 kg ingots and a typical result for an 18% chromium-11% nickel steel is shown in Fig. 2. The hot ductility is lower for as-cast material and the maximum ductility is reached approximately 50°C lower than in the forged material. In this case, a distinction was made between columnar and equiaxed structures. Because of the scatter in data, columnar and equiaxed structures have been grouped together up to the maximum. However above the maximum the equiaxed samples exhibited lower ductilities than the columnar ones. Fracture in these materials is initiated in areas of interdendritic segregation either within ferrite or at austenite-ferrite interfaces. For an average content of 4% ferrite in the as-cast steel of Fig. 2, the equiaxed structure contained 5% ferrite while the columnar structure contained 3% ferrite; this difference can possibly account for the lower ductility of the equiaxed structure in the higher range of temperatures. Further, in contrast to the forged material where recrystallization was associated with high ductility, no recrystallization was observed in the as-cast material.

It is clear from this brief discussion that a combination of the factors discussed in the earlier sections is operative in the inhomogeneous regions of cast materials.

Inclusions. The deletorious effect of certain inclusions on hot ductility is well known. However it should be noted that the deletorious effect is very dependent on the stress system imposed during hot working e.g. it may be possible to successfully extrude or even roll a material containing inclusions but such a material will not stand up to rotary piercing operations. Further, the initiation of fracture at inclusions is dependent on the relative deformation characteristics of the inclusion and the matrix. Thus the effects of sulphur, oxygen and manganese on ductility of iron alloys have been rationalised in terms of the change of shape and distribution of inclusions with change of composition (67).

The initiation of fracture at sulphide and silicate inclusions in mild steel under conditions simulating two roll rotary piercing is shown in Fig. 22, taken from a recent work of Cottingham (68). Clearly silicates are more dangerous than sulphides; however, because of their generally larger size and poor deformation characteristics, oxides are much more dangerous than sulphides. Cottingham emphasized the importance of the hydrostatic tensile stress component in a material in determining the initiation, rate of growth of cavities and coalescence by internal necking. Thus, using a notched tensile test, he found a linear correlation between the volume fraction of oxides + sulphide inclusions and notch ductility. In contrast a normal tensile test was relatively insensitive to the presence of inclusions and no correlation was obtained.

The very dangerous effects of oxide inclusions on ductility is illustrated dramatically by Fig. 23 which contrasts the ductility of a commercially pure iron with that of the same iron purified by zone refining (42). The oxygen content of the iron was reduced from 1000 ppm to ~ 7 ppm during the treatment and examination of the fracture mode indicated that in the commercially pure iron marked internal cracking was associated with oxide inclusions while in the zone refined iron failure occurred by propagation of cracks from surface irregularities. The marked reduction of ductility by oxide inclusions is comparable with that reported by White (61) for an iron-25% chromium alloy.

Acknowledgements

Much of the data and ideas used in this paper are the result of work carried out over a period of several years in the Department of Metallurgy, University of Sheffield, with my former colleague, Dr. C.M. Sellars and a large number of enthusiastic research students. My grateful thanks are also due to my friends at l'Institut de Recherches de la Siderurgie Francaise, St. Germain-en-Laye, for their generous co-operation over the past years.

I trust that their respective contributions are clearly evident from the text and references.

I have also drawn heavily upon contributions to the recent conference on 'Deformation Under Hot Working Conditions' held at the University of Sheffield in July 1966.

The preparation of this paper would not have been possible without the co-operation of my staff at Cranfield. I am particularly indebted to Mr. M.J. Luton for his assistance with the plotting of data and for useful discussions on the text, to Miss O. Lewis for her preparation of the manuscript and to Miss C. Hague for her preparation of the figures.

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Table 1 Activation Energies for Processes Associated with Hot Deformation (k cal/mole)

Material	Self-diffusion	Creep	Hot working	Recovery	Recrystallization
Al	33 (22)	33-36 (22)	37 (24)	23-64 (31)	34-59 (34)
Ni	70 (25)	65-67 (22)	54 (25)		70-90 (35), 32 (36)
Ni-5% Fe			71 (25)		
Ni-10% Fe			81 (25)		
Ni-20% Fe	53 Ni in Ni-Fe (27a) 52.5 Fe in Ni-Fe (27b)	48 (29)	88 (25)		
Cu	44-56 (22)	47-56 (22)	72 (21)		30 (34)
Fe-0.03%C(α)	57-67	68 (22)	66 (24)	22 → 67 (32)	
Fe-2.8% Si(α)	48±6 Si in Fe-Si (30)		80 (24)	70 (33)	70 (33)
Fe-25%Cr(α)	50 (22)		85 (21)		
Fe-0.05%C (γ)	65-74 (22)	61 (22)	67 (21)		
Fe-0.25%C(γ)	59 (22)	74 (22)	73 (21)		
Fe-18Cr-8Ni(γ)	67 Fe (28)	75 (22)	99 (21)		

Table 2 Comparison of Parameters n′ and Q Derived from Strength and Rupture Data on Nickel-Iron Alloys

Alloy	Strength correlation		Fracture correlation	
	n'	Q(k cal/mole)	n'	Q(k cal/mole)
Nickel	5.4	54	5.1	53
Nickel-5% iron	5.4	71	5.4	54
Nickel-10% iron	5.4	81	5.6	73
Nickel-20% iron	5.2	88	6.3	88

Table 3 Results of Programme Tests on Al-2% Mg Alloy

Deformation schedule Shear strain increments + rest time	Shear Strain at Failure				
	150°C	200°C	250°C	300°C	350°C
Continuous	6.3	5.6	6.0	6.3	17
1.27 + 30 sec.	14	13	16.5	16.5	29
1.27 + 2 min.	46	48	50	49	46
	400°C	450°C	500°C	550°C	600°C
Continuous	74	96	115	134	144
1.27 + 30 sec.	41	43	44	44	41
1.27 + 2 min.	37	32	31	30	28

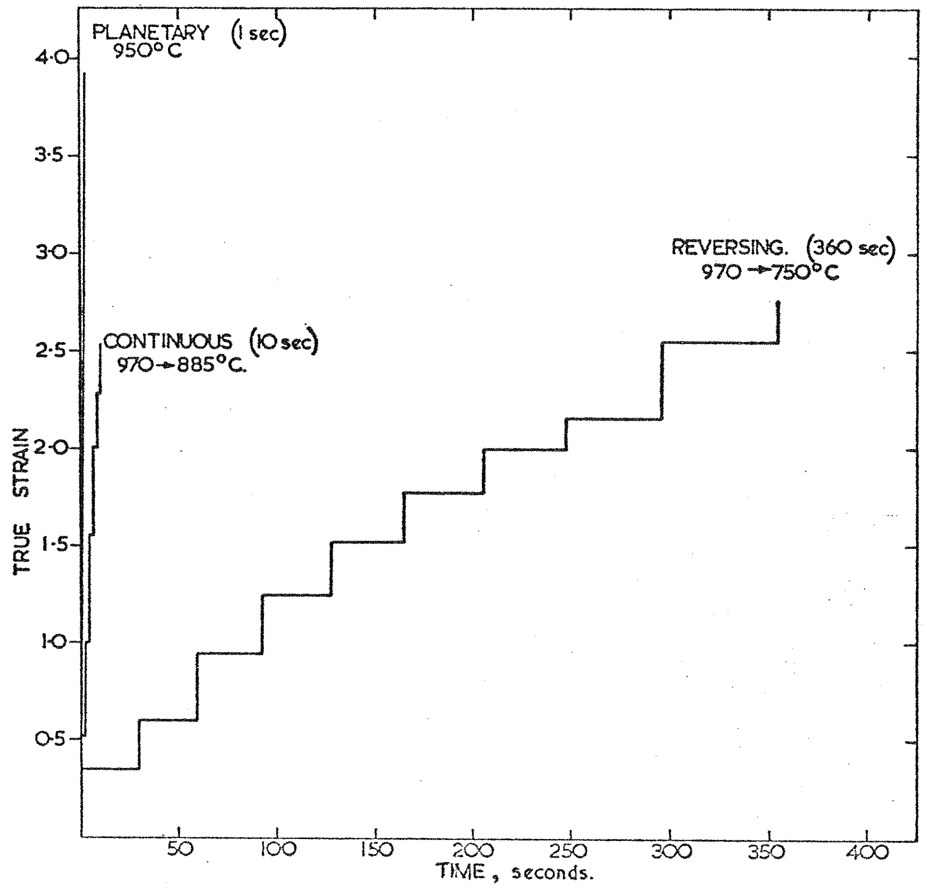


FIG. 1 DEFORMATION-TIME SCHEDULES INVOLVED IN VARIOUS TECHNIQUES FOR HOT ROLLING OF SHEET STEEL (1)

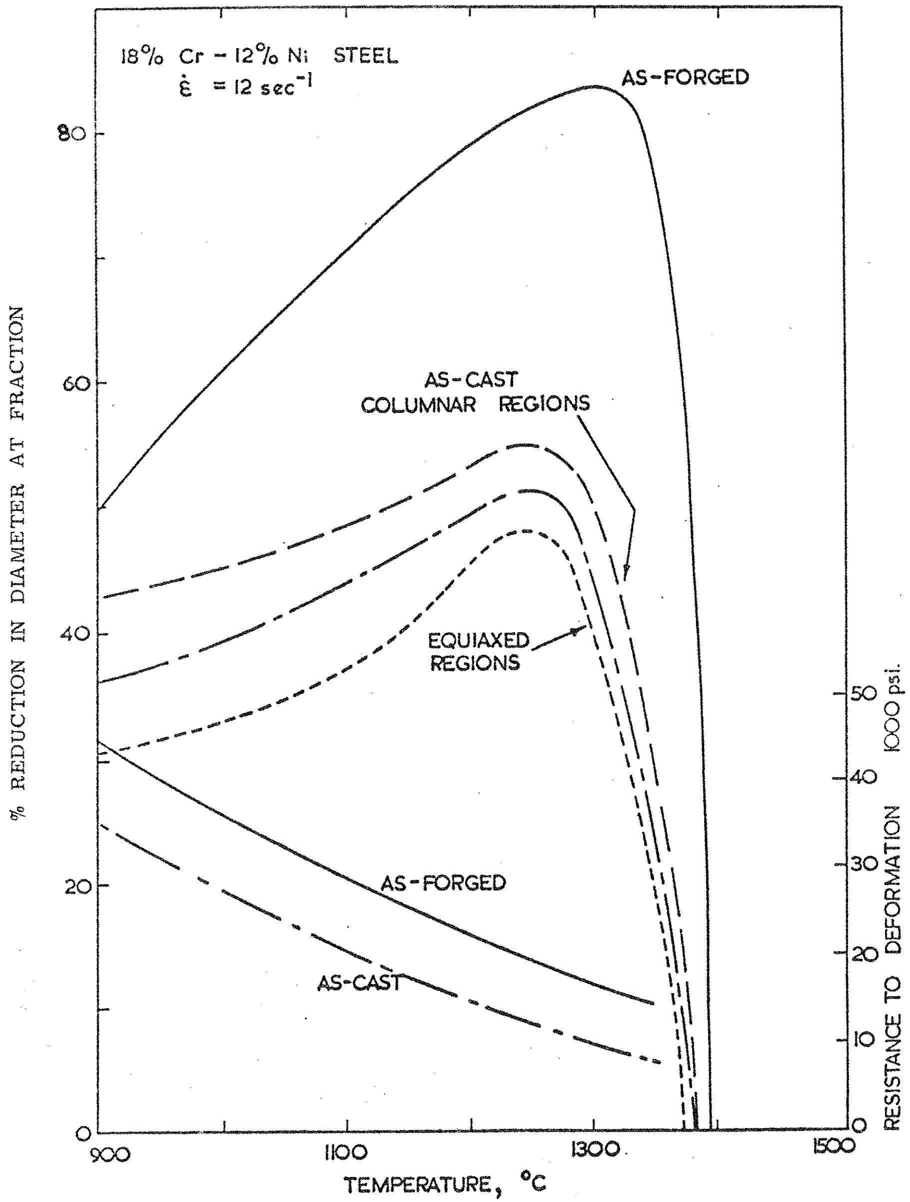


FIG. 2 DUCTILITY AND STRENGTH OF CAST AND FORGED 18% CHROMIUM-11% NICKEL STEEL AT VARIOUS TEMPERATURES AT A STRAIN RATE OF 12 SEC^{-1} (8)

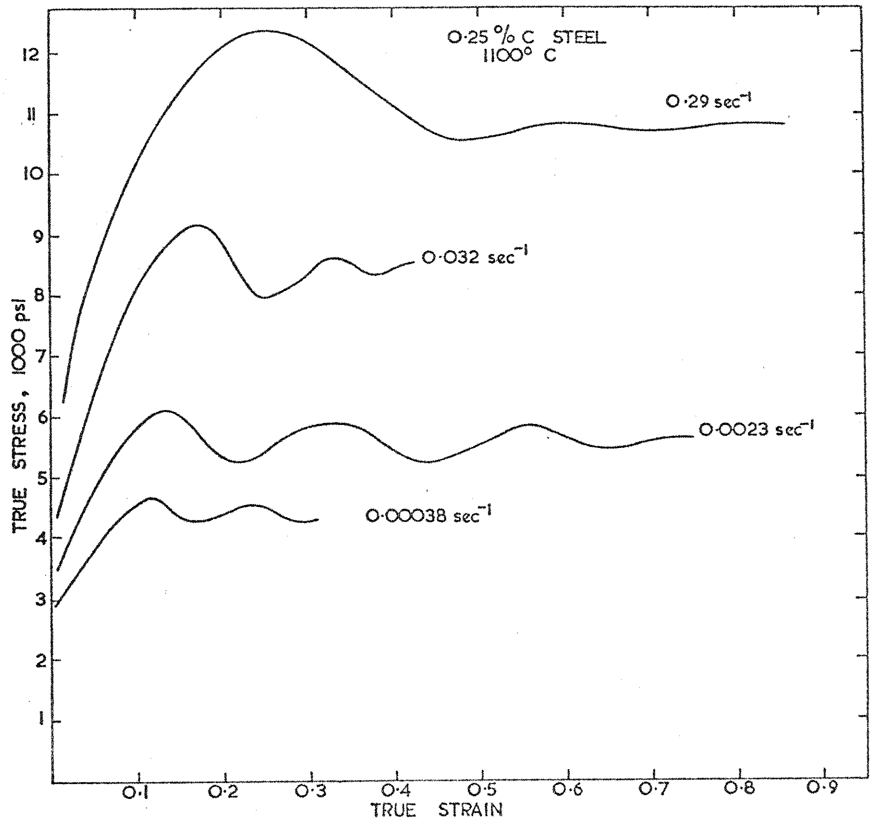


FIG. 3 STRESS-STRAIN CURVES FOR AN 0.25% C STEEL DURING TENSILE TESTS AT CONSTANT TRUE STRAIN RATES AT 1100°C (9)

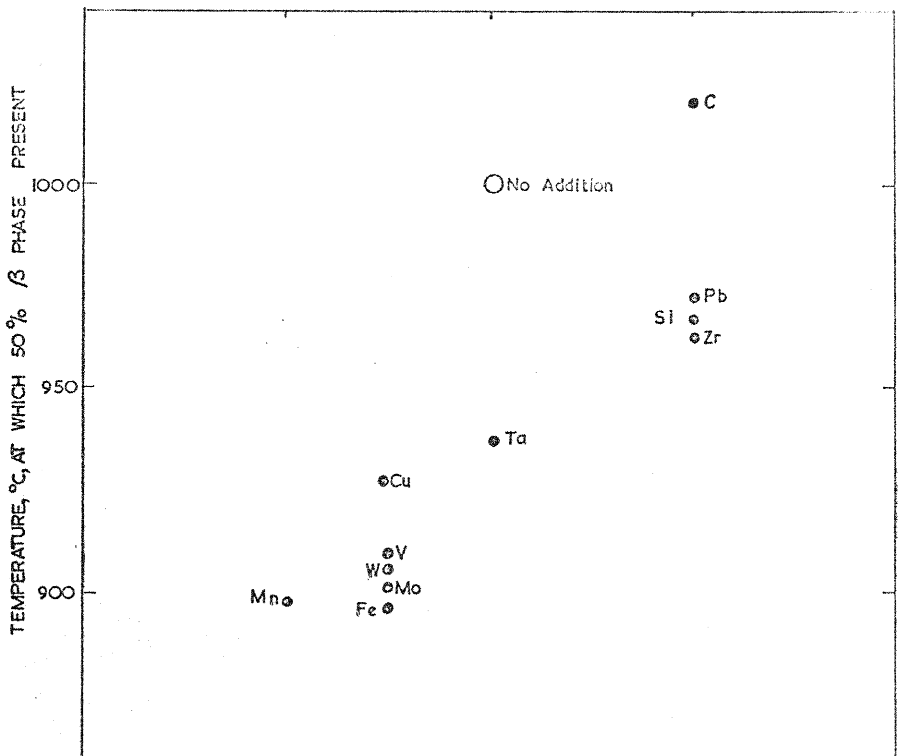


FIG. 4 RELATION BETWEEN β PHASE AND MINIMUM UPSETTING TEMPERATURE OF TITANIUM-12-5% TIN - 2.5% ALUMINIUM ALLOY WITH MINOR ADDITIONS (10)

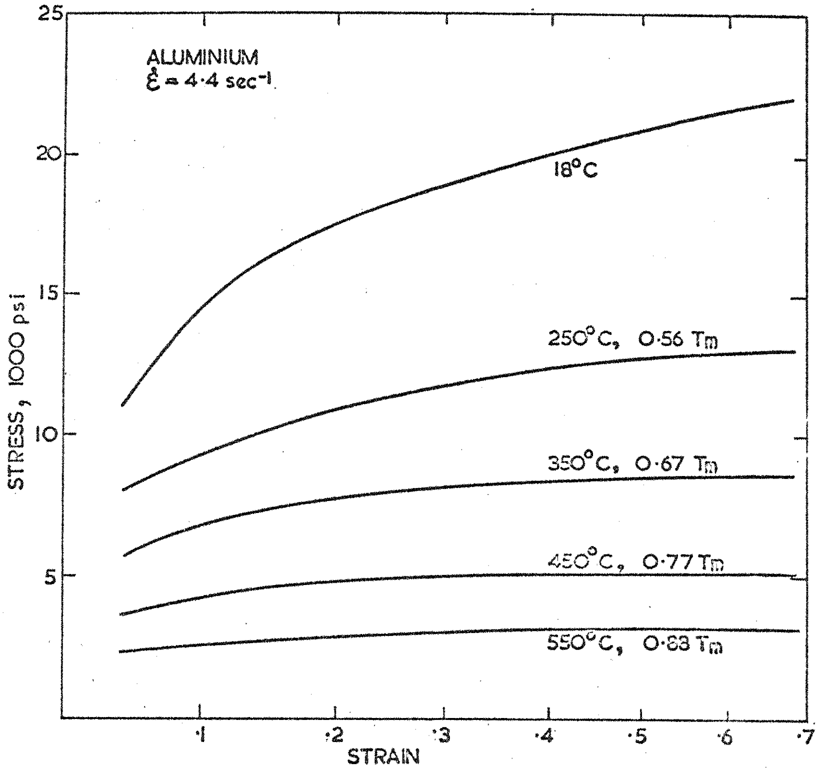


FIG. 5 STRESS-STRAIN CURVES FOR ALUMINIUM DURING COMPRESSION TESTS AT VARIOUS TEMPERATURES AT A CONSTANT TRUE STRAIN RATE OF 4.4 SEC^{-1} (11)

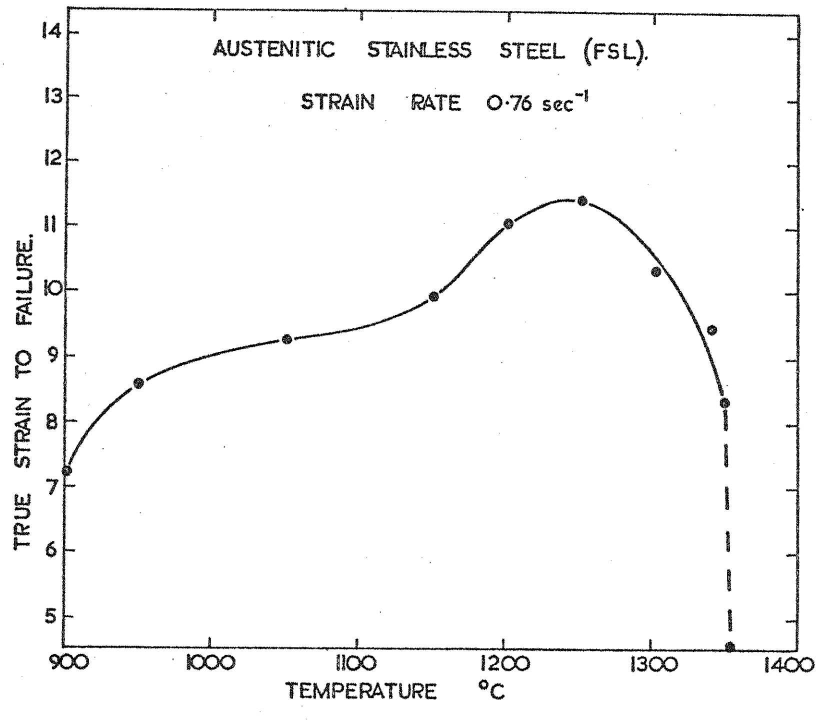
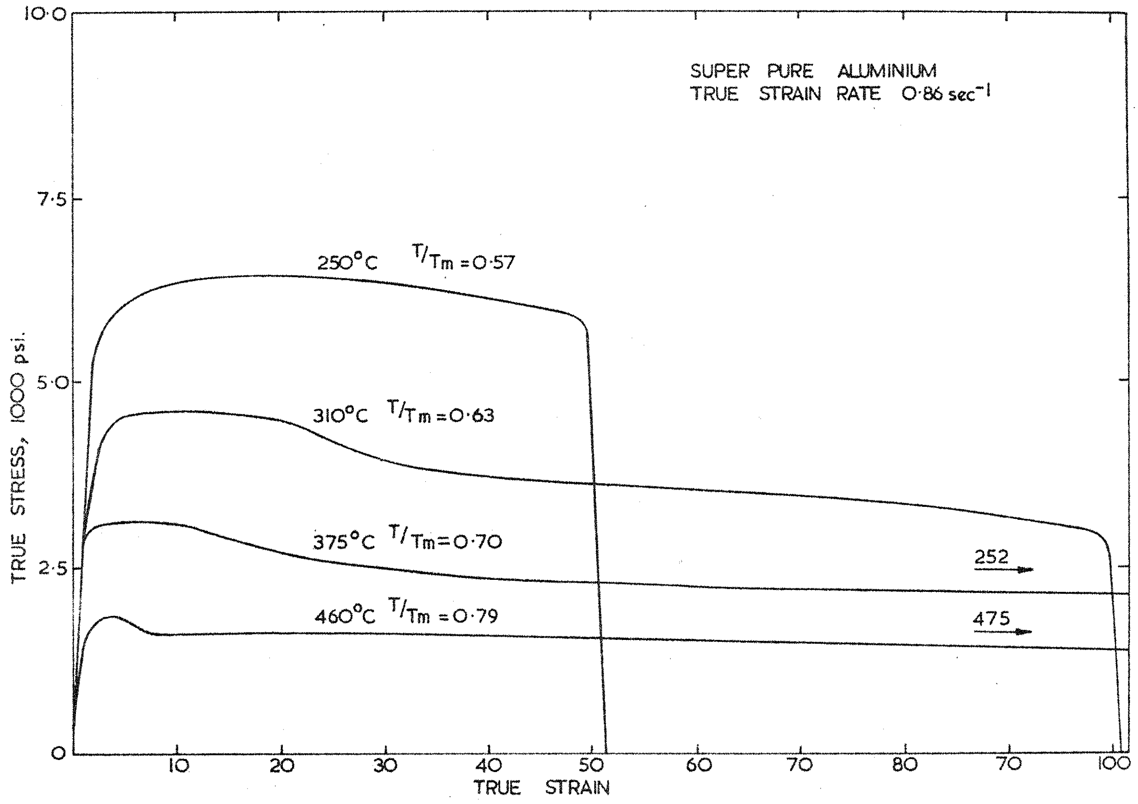


FIG. 6 DUCTILITY-TEMPERATURE CURVES OBTAINED DURING HOT TORSION OF AN AUSTENITIC STAINLESS STEEL. THE PEAK AT 1250°C CORRESPONDS TO THE OPTIMUM TEMPERATURE FOR ROTARY PIERCING (12)



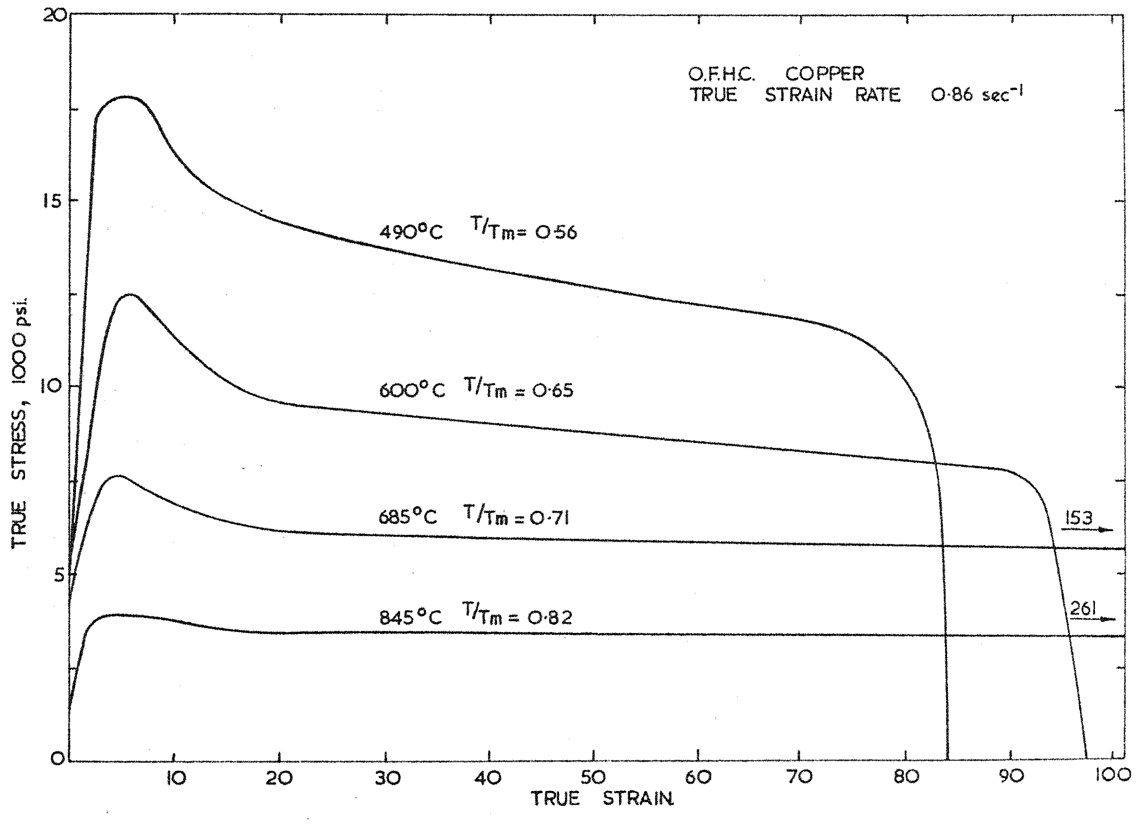


FIG. 7 STRESS-STRAIN CURVES DERIVED FROM HOT TORSION TESTS ON ALUMINIUM AND COPPER AT VARIOUS TEMPERATURES AT A CONSTANT STRAIN RATE (16)

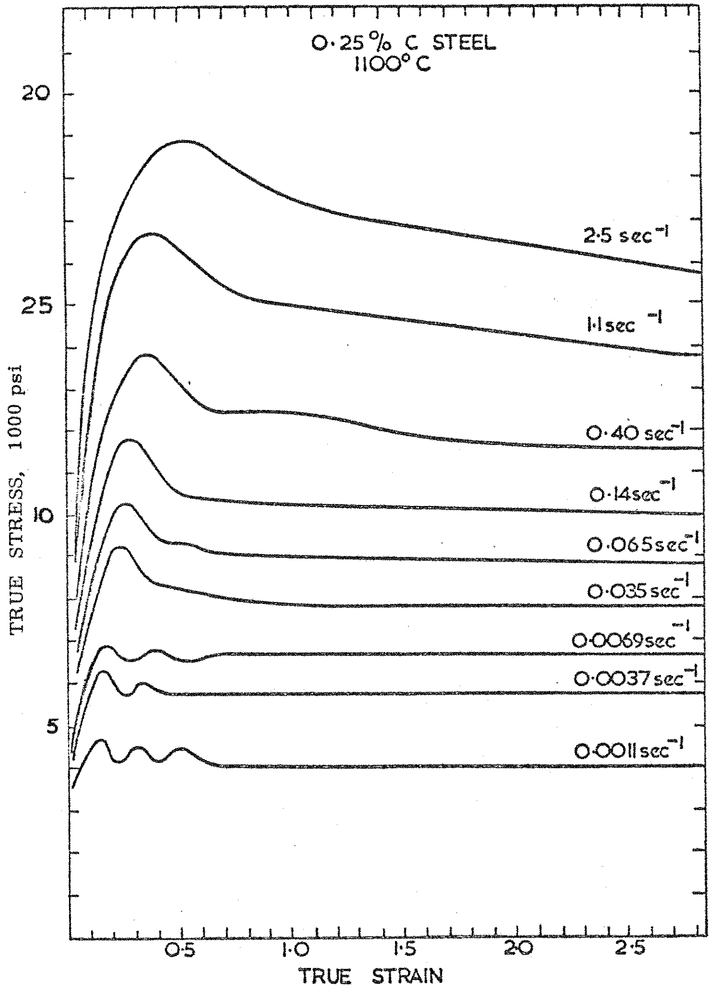


FIG. 8 STRESS-STRAIN CURVES DERIVED FROM HOT TORSION TESTS ON AN 0.25% C AT VARIOUS RATES AT 1100°C (17)

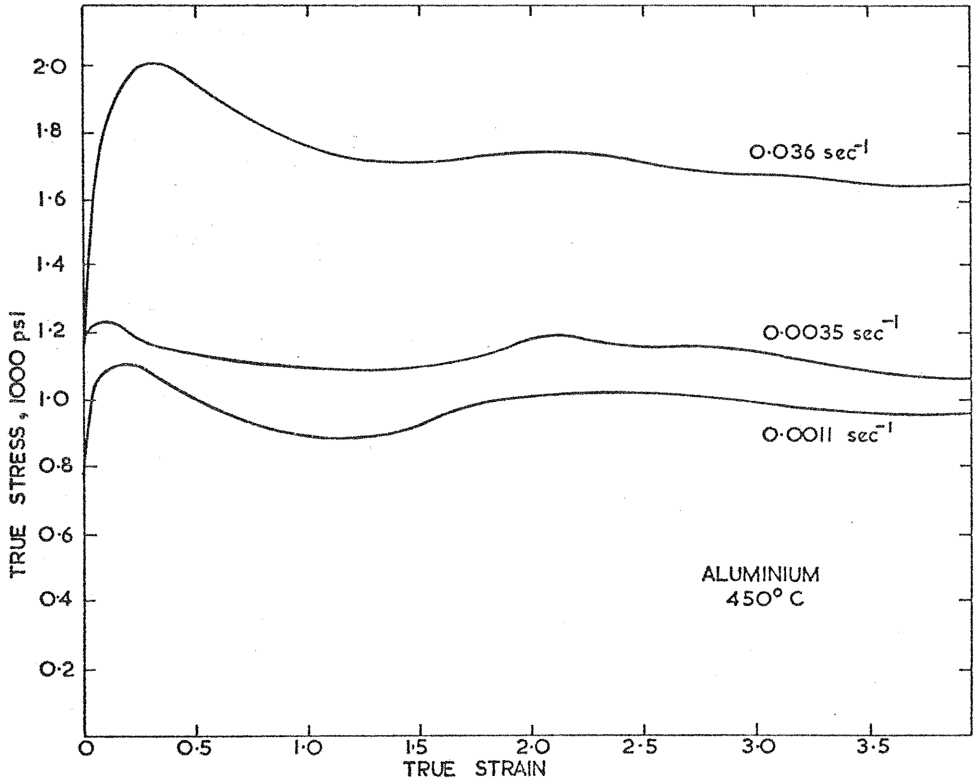


FIG. 9 STRESS-STRAIN CURVES DERIVED FROM HOT TORSION TESTS ON COPPER AND ALUMINIUM AT LOW STRAIN RATES (18)

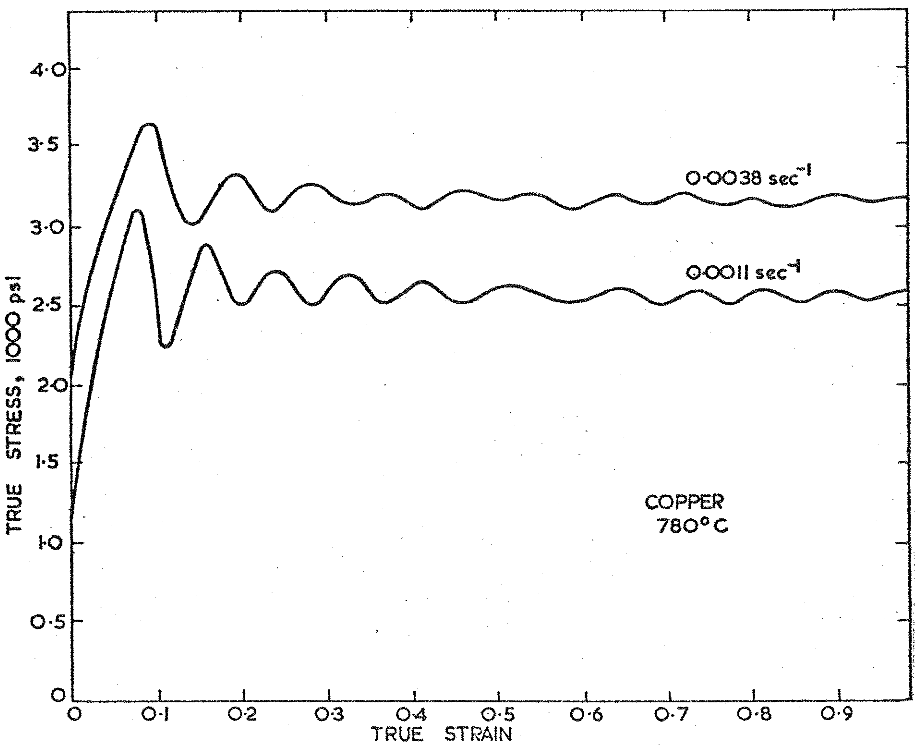
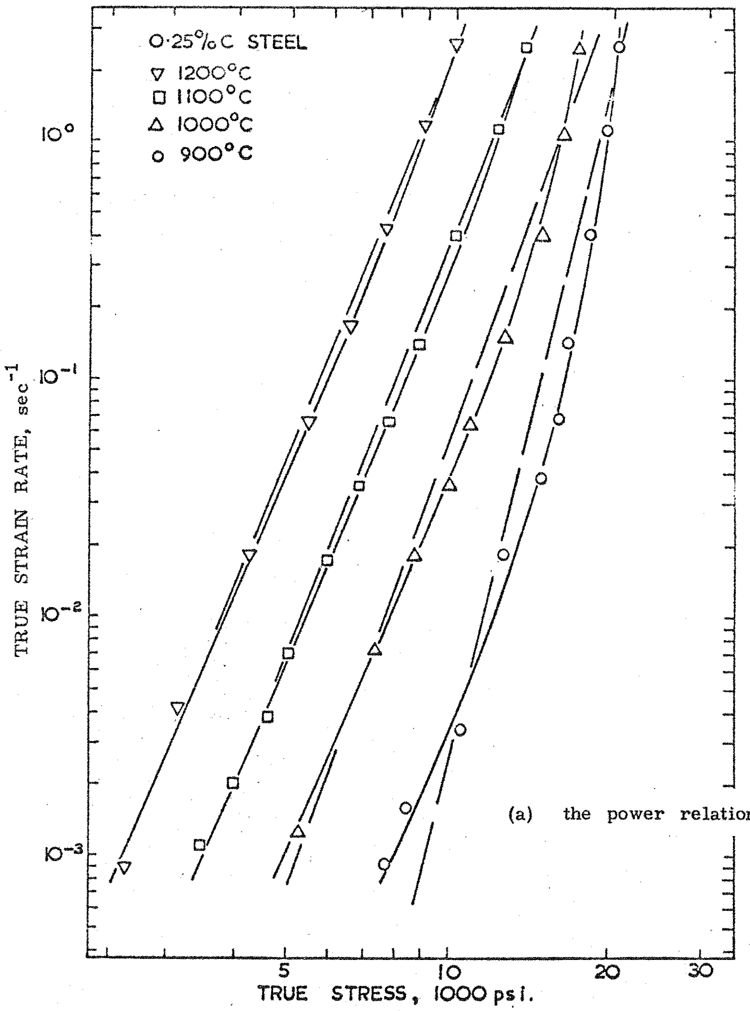


FIG. 9 (cont.) STRESS-STRAIN CURVES DERIVED FROM HOT TORSION TESTS ON COPPER AND ALUMINIUM AT LOW STRAIN RATES (18)



(a) the power relationship

FIG. 10 (a) STRENGTH DATA DERIVED FROM HOT TORSION TESTS OF ROSSARD AND BLAIN (17)

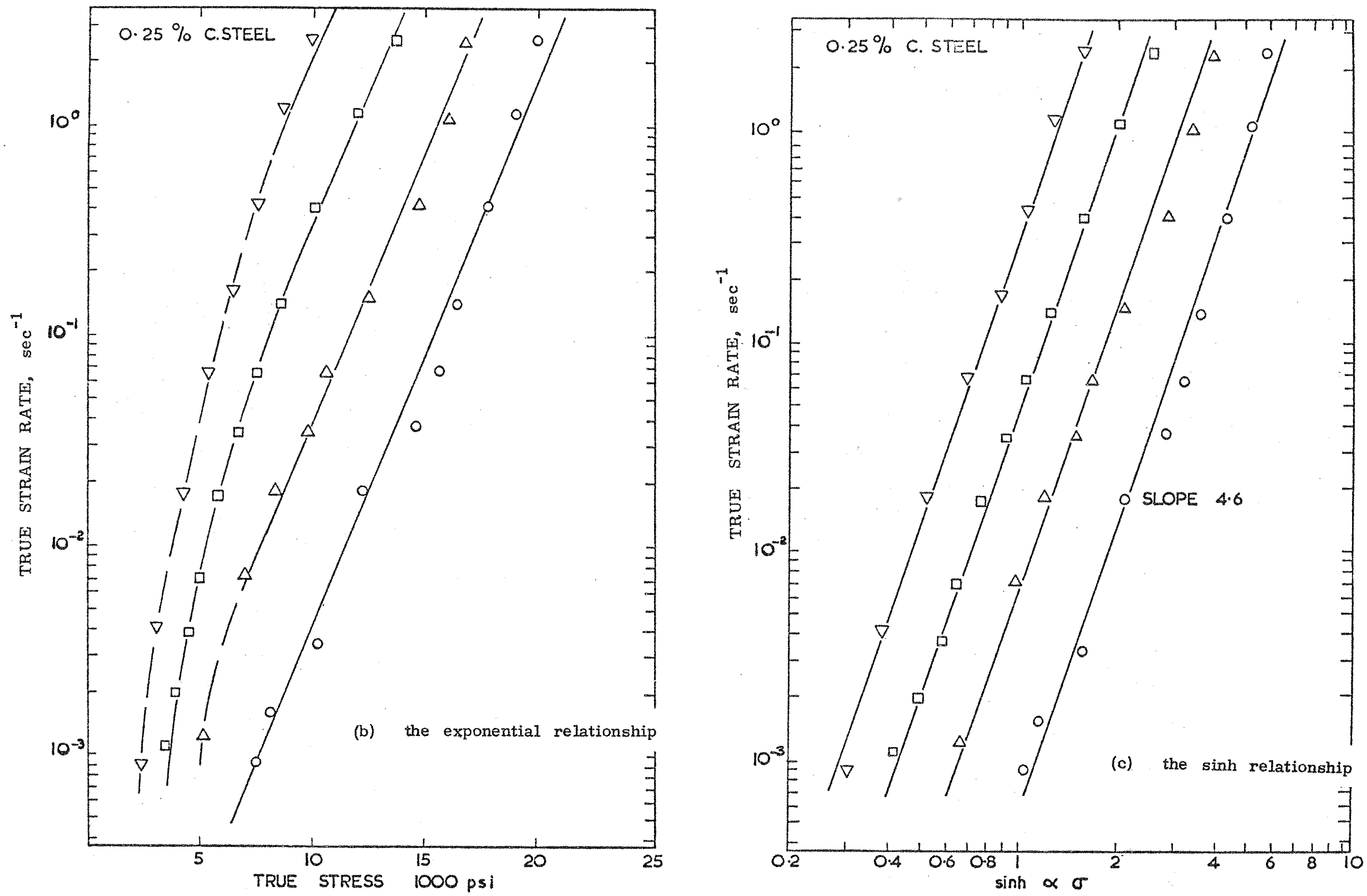


FIG. 10 (b) & (c) STRENGTH DATA DERIVED FROM HOT TORSION TESTS OF ROSSARD AND BLAIN (17)

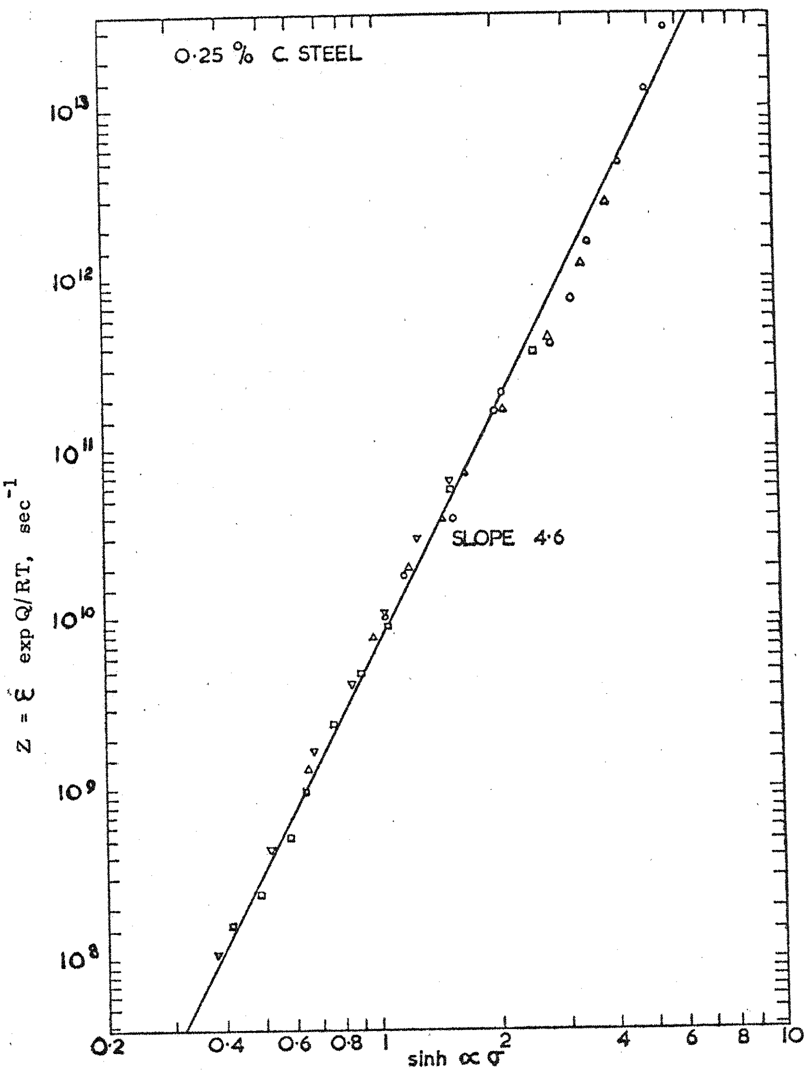
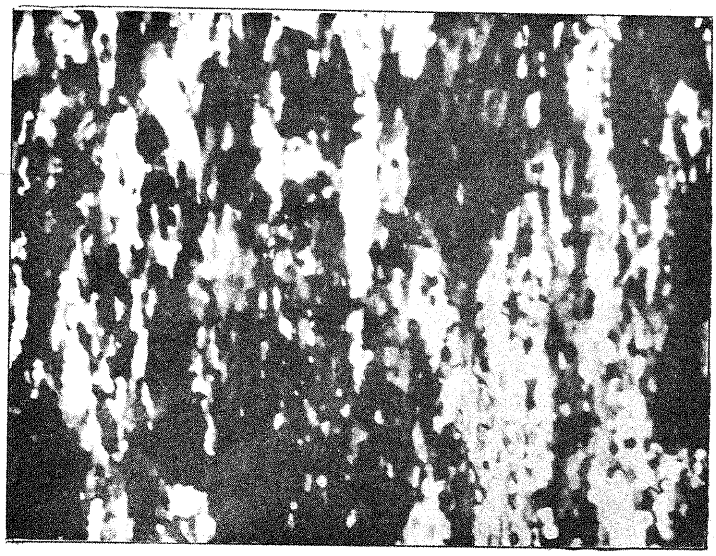
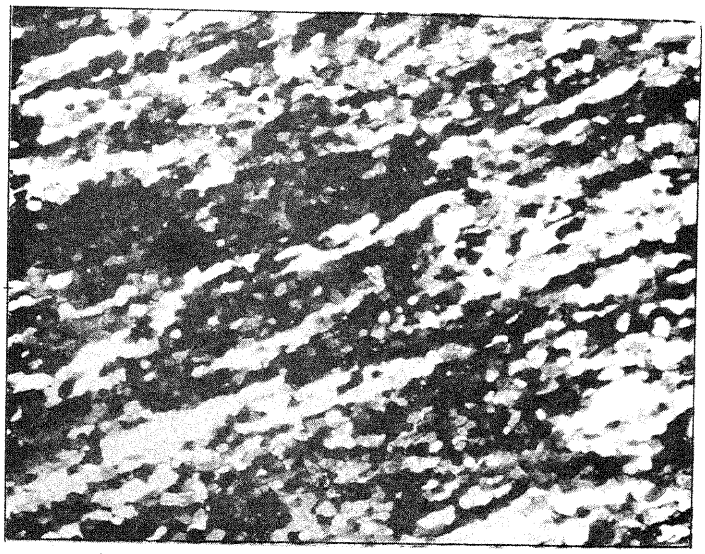


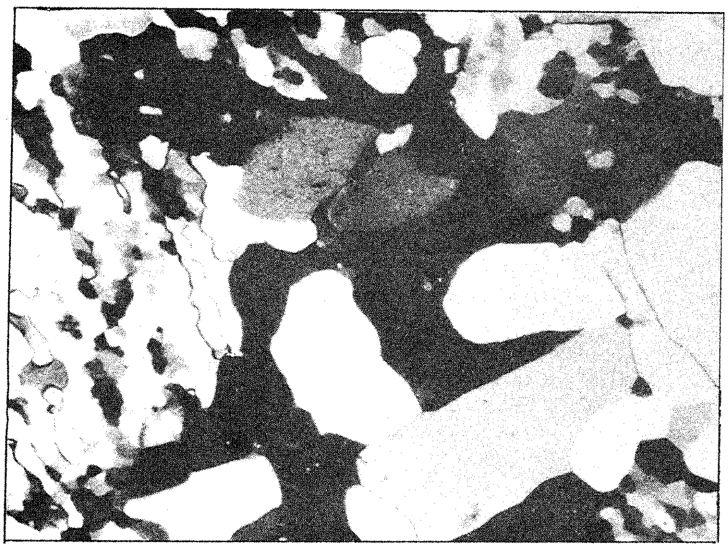
FIG. 11 CORRELATION OF STRENGTH DATA OF FIG. 10 USING A TEMPERATURE COMPENSATED STRAIN RATE PARAMETER



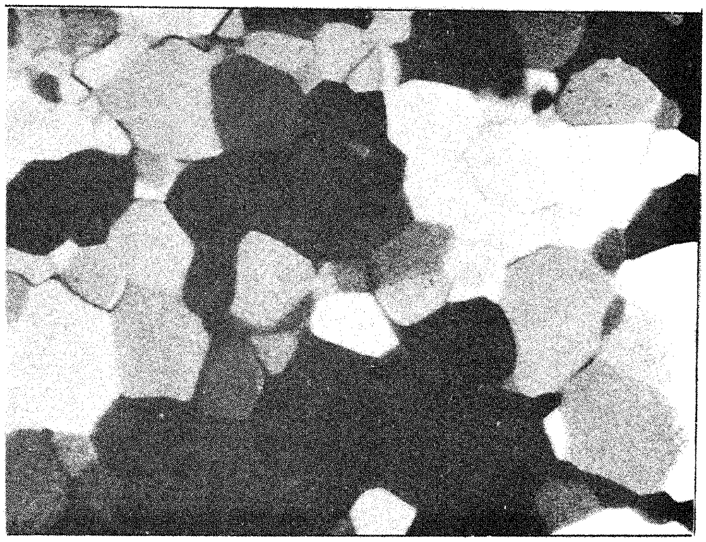
(a) immediately quenched



(b) 5 min. anneal

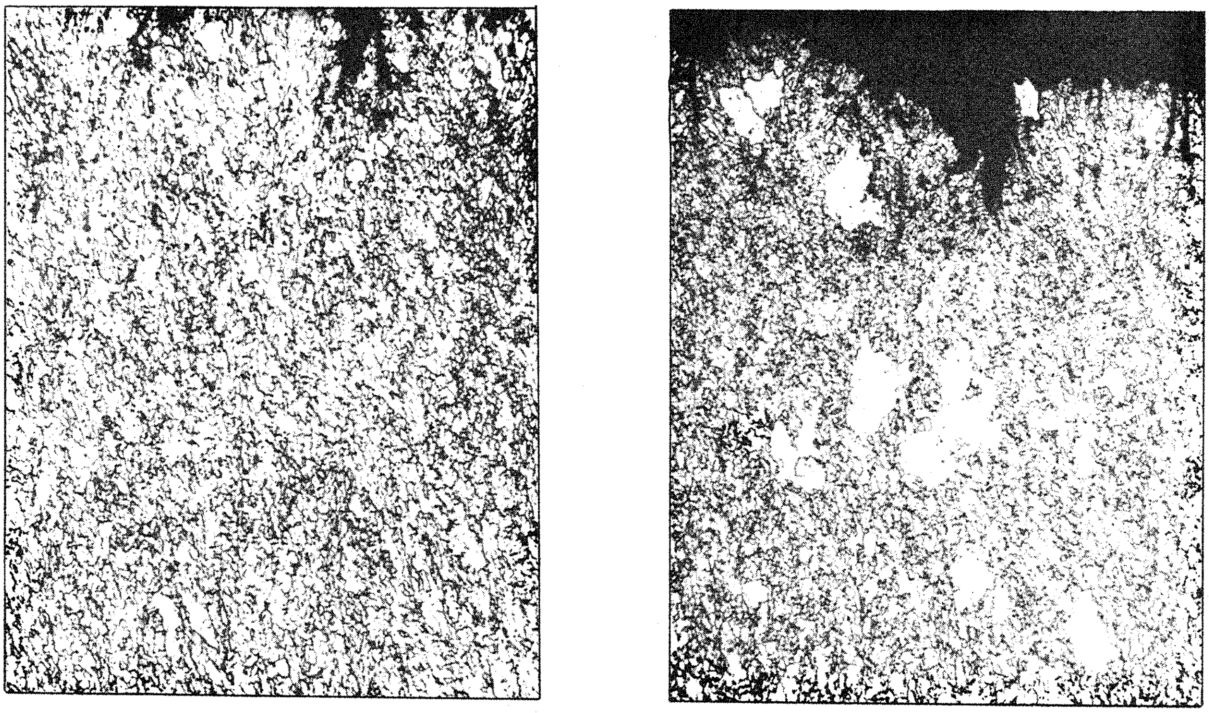


(c) 10 min. anneal



(d) 30 min. anneal

FIG. 12 STRUCTURES OBSERVED IN PURE ALUMINIUM AFTER A STRAIN RATE OF ~ 2 SEC^{-1} AT $400^{\circ}C$ (x 200)



(a) immediately quenched

(b) 45 min. anneal.

FIG. 13 STRUCTURES OBSERVED IN COMMERCIALLY PURE IRON AFTER A STRAIN OF 4 AT A STRAIN RATE OF 0.5 SEC^{-1} AT 800°C (x 63)

(a) strain of 1.4 (x 250)

(b) strain of 2.3 (x 44)

(c) strain of 10 (x 250)

FIG. 14 STRUCTURES OBSERVED IN O.F.H.C. COPPER AFTER VARYING AMOUNTS OF STRAIN AT A STRAIN RATE OF 0.5 SEC^{-1} AT 870°C

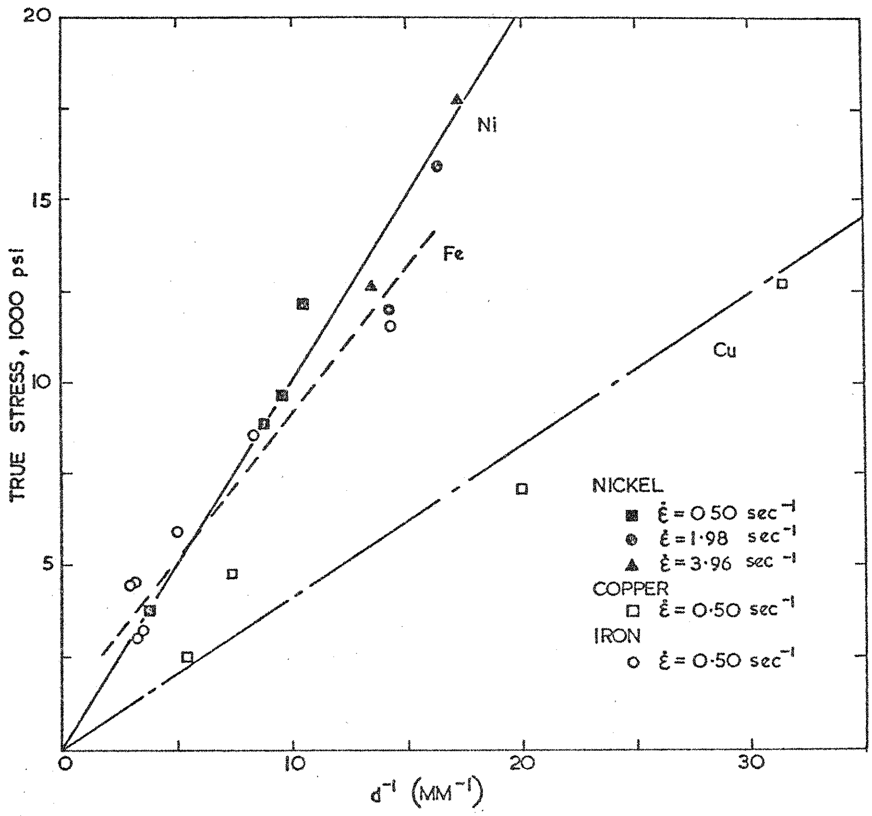


FIG. 15 RELATIONSHIP BETWEEN HIGH TEMPERATURE STRENGTH AND GRAIN SIZE FOR COMMERCIALLY PURE COPPER, NICKEL AND IRON

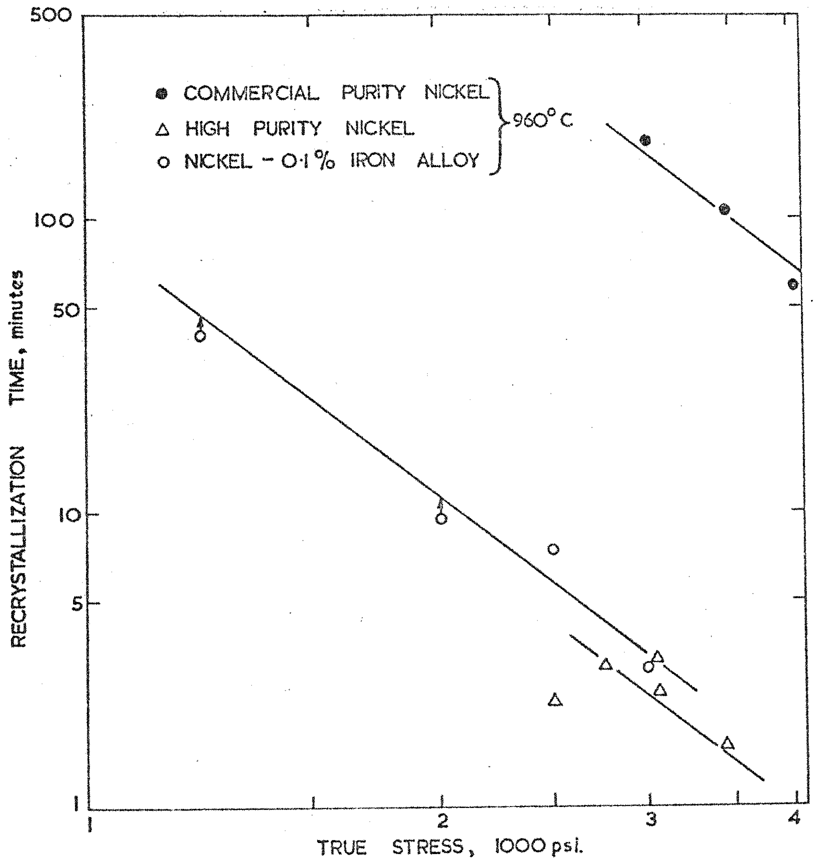


FIG. 16 INFLUENCE OF STRESS ON RECRYSTALLIZATION TIME DURING CREEP OF TWO GRADES OF NICKEL AND A DILUTE NICKEL-IRON ALLOY

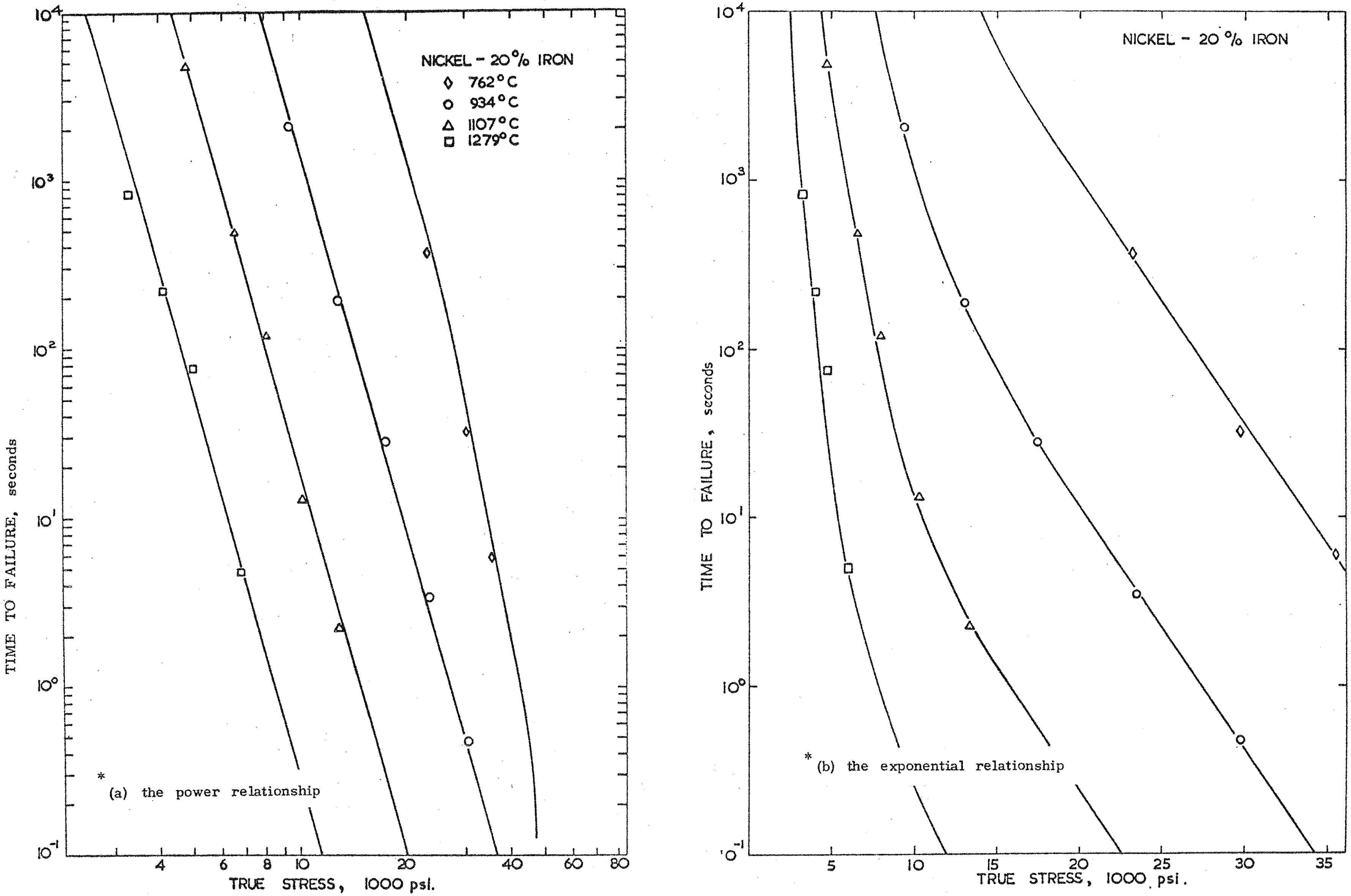
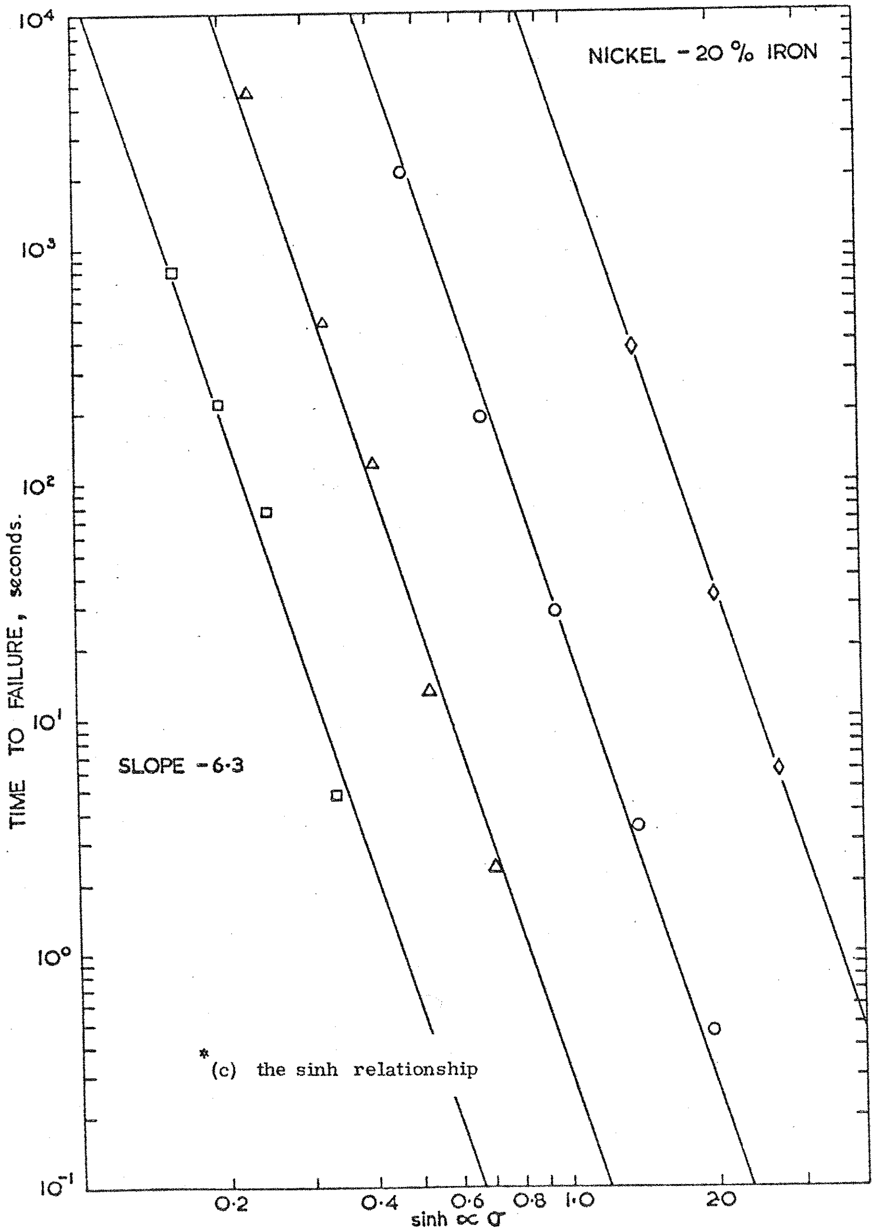


FIG. 17 (a) and (b) RUPTURE DATA DERIVED FROM HOT TORSION TESTS OF LUTON (25) ON NICKEL-20% IRON ALLOY PLOTTED TO DEMONSTRATE*



*(c) the sinh relationship

FIG. 17(c) RUPTURE DATA DERIVED FROM HOT TORSION TESTS OF LUTON (25) ON NICKEL-20% IRON ALLOY PLOTTED TO DEMONSTRATE*

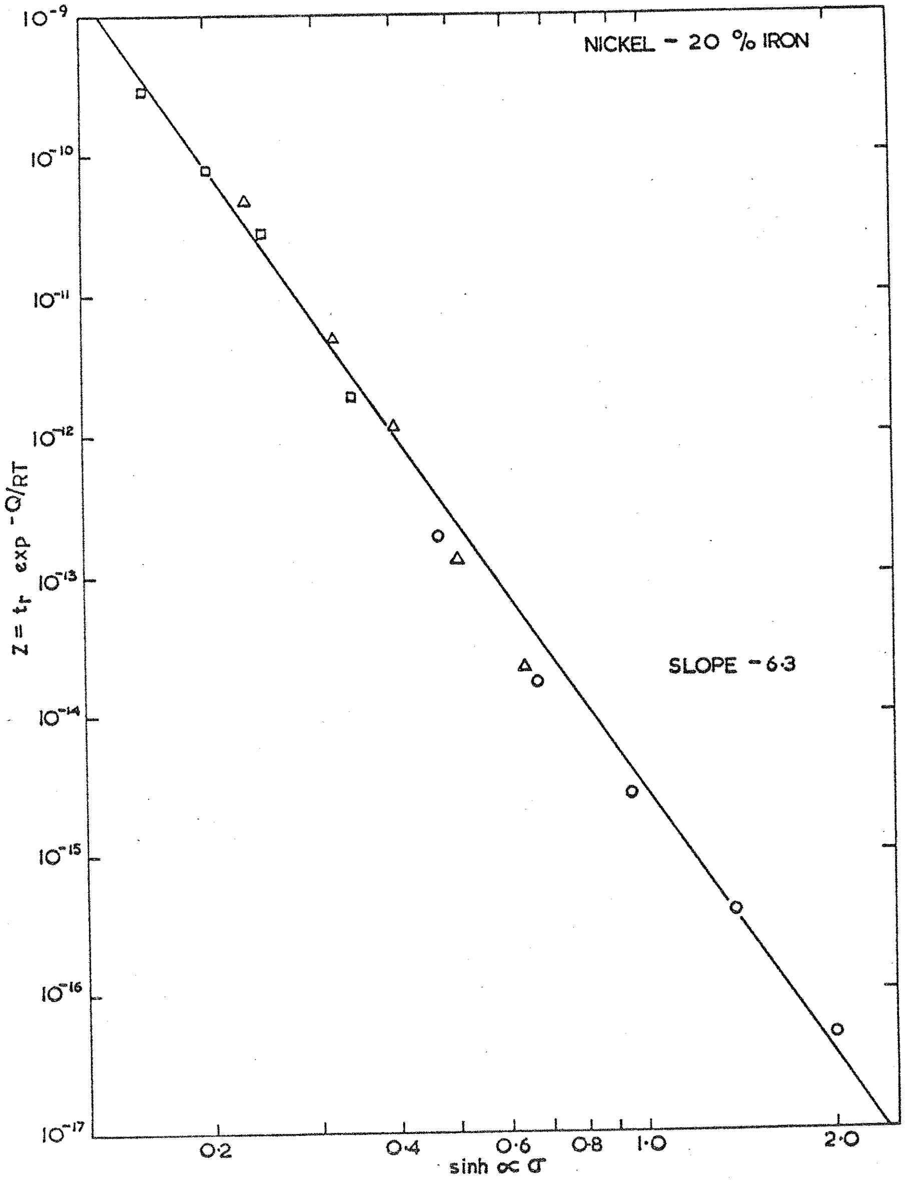
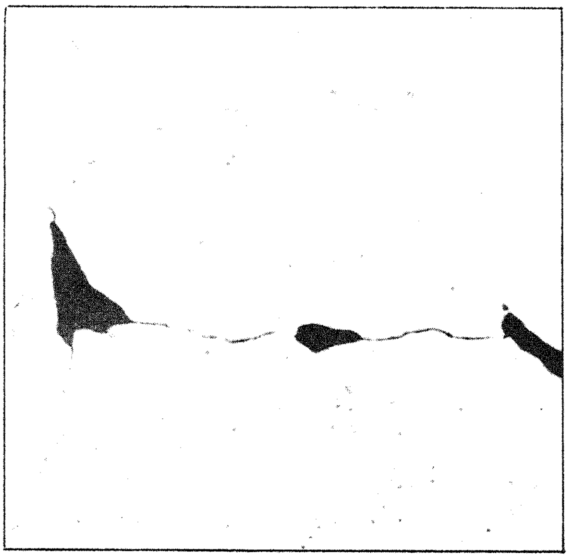
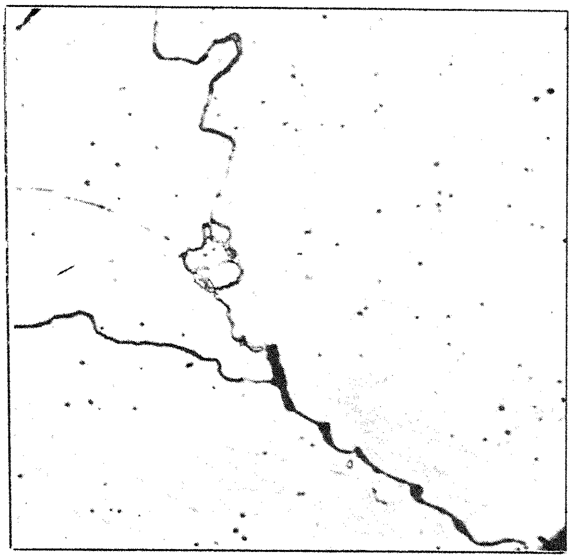


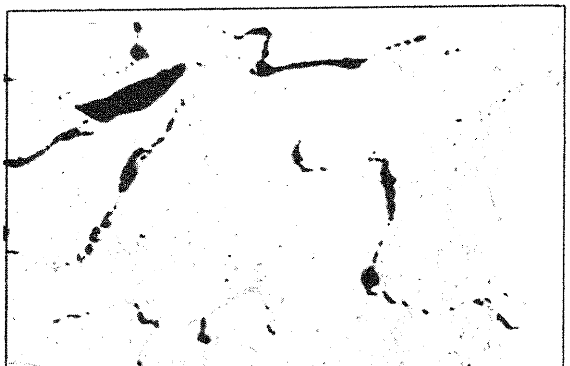
FIG. 18 CORRELATION OF RUPTURE DATA OF FIG. 17 USING A TEMPERATURE COMPENSATED RUPTURE TIME PARAMETER





(a) after a strain of 1.1 at 0.01 sec^{-1} (b) after a strain of 1.4 at 0.001 sec^{-1}

FIG. 19 TRIPLE POINT AND GRAIN BOUNDARY CRACKING IN AN IRON-25% NICKEL ALLOY AT 1100°C (46) (x 760)



(a) intergranular fracture along original grain boundaries

(b) voids formed by isolation of intial cracks by recrystallization

FIG. 20 FRACTURE MODES IN AN IRON-25% NICKEL ALLOY AT 1100°C (46)

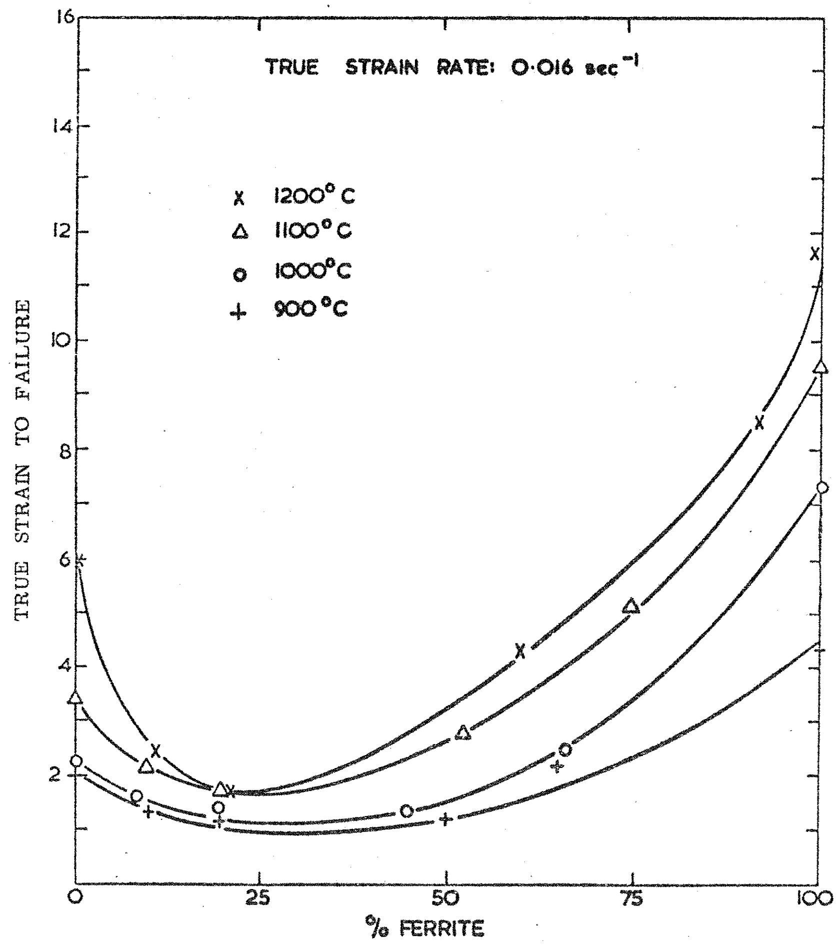


FIG. 21 DEPENDENCE OF DUCTILITY OF WROUGHT IRON-CHROMIUM-NICKEL ALLOYS IN FERRITE CONTENT AT A STRAIN RATE OF 0.02 SEC^{-1} (26)

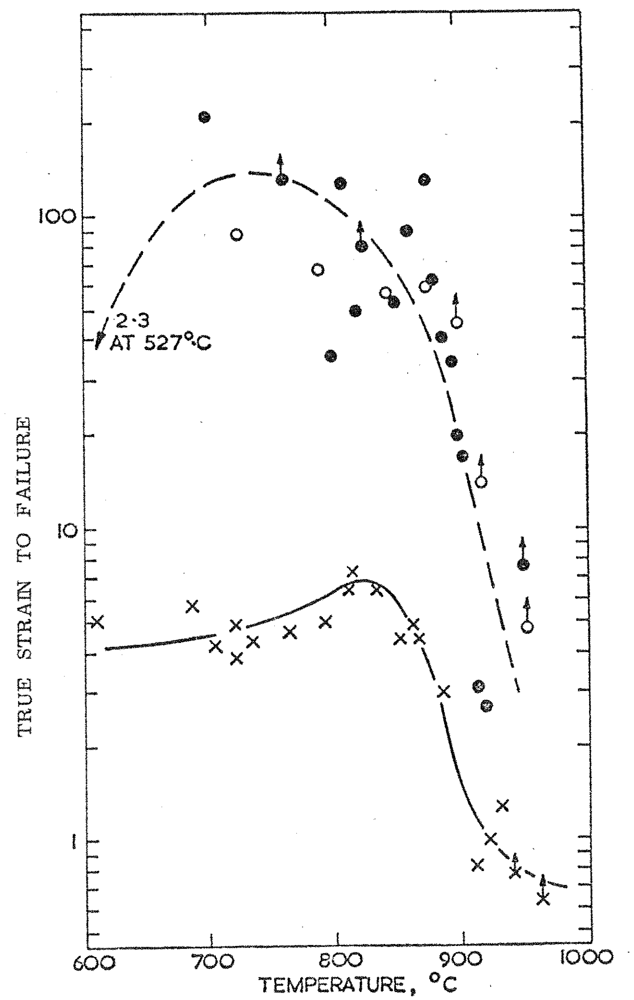
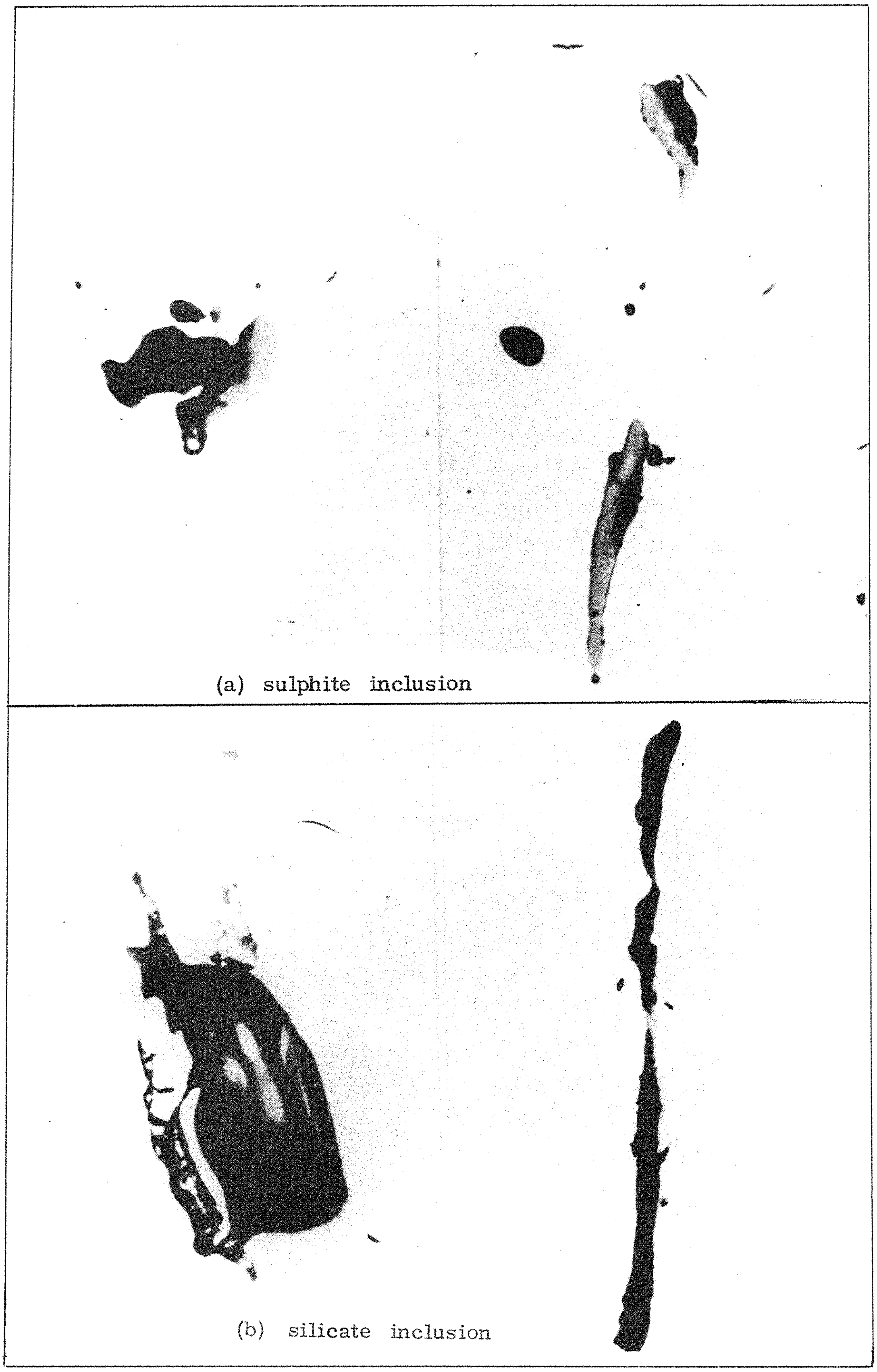


FIG. 23 EFFECT OF ZONE REFINING ON DUCTILITY-TEMPERATURE CURVES FOR IRON (x COMMERCIALLY PURE; ● AFTER 15 ZONE PASSES; o AFTER 25 ZONE PASSES) (↑ INDICATES SPECIMEN FAILED OUTSIDE THE GAUGE LENGTH)



(a) sulphite inclusion

(b) silicate inclusion

FIG. 22 INITIATION OF FRACTURE AT INCLUSIONS IN MILD STEEL AT 1200°C IN A SQUARE NOTCHED TENSILE TEST (69)