

THE EFFECT OF TEMPERATURE AND
STRAIN RATE ON THE DUCTILE-BRITTLE
TRANSITION IN NOTCHED STEEL BARS

by

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ABSTRACT

The liability of a steel to failure by brittle fracture may be judged by several criteria as exhibited in the various standard engineering tests. The relative influence of the many factors governing the ductile-brittle transition may vary greatly from test to test with the result that the various arbitrarily defined transition temperatures will also vary over quite a large range. Despite the complexity of the relationship between tests, some form of correlation is frequently demanded. This thesis presents experimental results in which, starting with the slow bend test on a standard V-notch Charpy specimen, a test has been successively modified, care being taken to alter only one variable at a time. In this way it was possible to study the relative effect of strain rate, size and method of stress application (i.e. notch-tension or notch-bending) on the change in transition temperature between tests.

Load-deflection diagrams were obtained on all types of specimen so that the loads, energies, etc., at any given stage of the test, might be determined. Special instrumentation was developed to obtain diagrams of load-deflection, as well as of load-time, in the impact test. A comparison of these two diagrams allowed the variation of striker velocity in the impact test to be studied, and it was concluded that the

generally accepted assumption of the velocity varying linearly with time was likely to be in error, and that some form of shock loss of velocity occurred at impact.

Finally the experimental results were examined in the light of some of the more recent hypotheses which have been advanced on the brittle fracture phenomenon.

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NOTATION

$\sigma_1, \sigma_2, \sigma_3$ = Principal stresses.

$\sigma_x, \sigma_y, \sigma_z, \tau_{xy}, \tau_{yz}, \tau_{zx}$ = Co-ordinate Component stresses.

σ_s = Shear stress.

σ, σ_n = Nominal applied stress.

σ_m = Maximum stress.

$(\sigma_a)_{\max.}$ = Theoretically determined maximum principal stress.

σ_{nb} = Nominal bending stress at the notch root.

σ_{nt} = Nominal tensile stress at the notch root.

σ_{nf} = Nominal fracture stress at the notch root.

σ_{yd} = Yield stress (uniaxial).

σ_{ys} = Yield stress in shear.

σ_c = Critical stress.

E, μ = Modulus of Elasticity, Modulus of Rigidity.

ν = Poisson's ratio.

k = Stress concentration factor.

u = Displacement.

b = Crack opening dislocation.

$2a$ = Crack length.

r, θ = Polar co-ordinates.

$2T$ = Surface tension.

P = Largely plastic work of crack wall formation.

G_c = Fracture toughness.

G = Strain energy release/unit area.

R = Crack growth resistance.

K = Stress intensity factor.

K_c = Value of K at fracture.

K_{ic} = K_c under plane strain conditions.

W = Plate width.

B = Plate thickness.

t = Shear lip (t_0 = Critical shear lip).

E_{sl} = Energy of fracture in shear lip.

E_{ff} = Energy of flat fracture.

D = Specimen dimension.

r = Radius, Notch root radius.

r_{yd} = Radius of plastic zone.

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1. INTRODUCTION

Whenever the term "brittle fracture" is mentioned today there is a tendency for engineers to think immediately of the much publicised failures of various all welded ships, particularly tankers, during the second World War. In view of the vast amount of research which was stimulated by this urgent problem, the prominence given to them is probably merited. The rather unfortunate consequence of this attitude is to consider brittle fracture as a comparatively modern problem, which is limited to mild steel and which has now been solved. In fact fractures of a brittle nature occur in many materials besides mild steel, some of which are fast becoming of considerable importance as structural materials. Also brittle fracture in steel itself has been present in various guises ever since the introduction of the Open-hearth and Bessemer processes allowed steel to be made sufficiently cheaply for it to be used as a structural material. An explanation of the fact that a normally ductile material can, under certain circumstances, fracture with severely reduced deformation was demanded as early as 1879 (1) and it is a somewhat humbling thought that even now the answers given would be far from complete. Detailed reviews are available of the historical development of the subject (2 to 5) and it is interesting to note the different forms in which the same basic problem has reappeared. It would seem pertinent to mention that although steel bridges were among the earliest

structures to suffer from brittle fracture, yet in 1962 the new King Street bridge in Melbourne encountered similar difficulties of a very serious nature.

A very large amount of experimental work has been done on this topic in recent years, especially with regard to mild steel, and the amount of literature now available on the various aspects of brittle fracture is almost overwhelming and, as Drucker has said, one is now in the position of writing summaries of summaries. Parker (4), Biggs (5) and Tipper (2, 6), however, have greatly clarified the issue by their very full discussions of the problem and the general factors involved should no longer be in doubt, at least in so far as mild steel is concerned. Quite a large part of these discussions dealt with the question of methods of testing for brittle fracture and the interpretation of the results of these tests. It immediately becomes clear that there is no such thing as a single, clear-cut, transition temperature at which the behaviour of a given material changes its mode of fracture from ductile to brittle. A transition temperature for a given material can only be defined for a closely specified test as judged by a given arbitrary criterion. Thus as well as the very large differences in transition temperature which exist between the various tests there can also be a considerable variation, depending on the criterion chosen, and this variation itself is not constant from test to test. The aim of the present work was to select materials, namely mild steels, whose behaviour in the different types of

test had been thoroughly investigated and, by careful control, to study the effects of the individual variables on the transition temperatures as judged by the various criteria. It was also hoped to relate these results to some of the hypotheses which have been presented in an attempt to place the problem on a more quantitative basis.

2. THEORETICAL ASPECTS

The theoretical background to this topic embraces many different fields, and explanations of some particular phenomenon are often advanced from viewpoints which may differ greatly. To obtain a balanced analysis of any new contribution it is often necessary for an investigator to stray into fields which are somewhat removed from his original discipline, and it is inevitable that ignorance may, in some cases, lead to erroneous conclusions. It is felt that the following, rather lengthy, survey is justified as an attempt to present the author's understanding of some of these views in a reasonably comprehensive picture. Any interpretation of the experimental results will be made in the light of this presentation.

2.1. The Ductile Brittle Transition

It is vitally important when considering a topic of this kind to avoid, as far as possible, any confusion in definitions or in terminology. In general terms brittle fracture may only be said to be a fracture which takes place with severely reduced deformation and is not limited to any special type of material. It will later be possible to be more precise and, once a material is specified, to define certain patterns of behaviour which are followed.

(a) The Fracture Mode Transition

It has been found that the type of fractures obtained in many materials is strongly dependent on the applied conditions, such as temperature and rate of stressing especially in the

presence of some discontinuity such as a sizable inclusion or artificial notch. Thus a notched tensile specimen of such a material may fracture in a ductile manner at a given temperature with gross deformation occurring in the region of the fracture, whilst an identical specimen tested at a slightly lower temperature or higher stress rate may fracture in a manner which is obviously brittle, any slight deformation being restricted to the immediate locality of the fracture surfaces. The thickness and/or width of the notched plate has also been found to have a strong influence on the type of fracture (7); the fracture changing from a "shear" fracture on a plane at 45° to the plate surface to a flat fracture normal to the plate surface as the thickness was increased. This form of the ductile-brittle transition does not imply any change in the atomic mode of fracture and is equally possible in metals such as steel (b.c.c.) and aluminium (f.c.c.) or in the notch brittle plastics such as polystyrene, which possesses a complex molecular chain structure.

(b) The Shear-Cleavage Transition

It is possible in some materials to undergo another transition at an atomic level and it is the possible superposition of these transitions that makes more difficult the understanding of the ductile-brittle transition in such materials. Three basic types of fracture have been observed in crystalline materials, two of which are transcrystalline, the third intercrystalline. The latter type is usually a result

either of prolonged loading at high temperatures or of induced weaknesses at the grain boundaries produced by segregation of impurities or corrosion. It will be shown later, however, that typical brittle fractures have taken place in this fashion. Of more general importance are the two trans-crystalline modes, namely shear and cleavage, which produce fractures which differ radically in appearance. It has been found that most crystal lattices possess both slip or glide planes along which one part of the crystal slides along the other (by means of the consecutive movement of dislocations) until complete separation has occurred, governed largely by the applied shear stress, and also cleavage planes at which a direct separation takes place with very little plastic deformation probably obeying a normal stress law. Whether fracture occurs on the slip or cleavage plane is governed by external conditions such as temperature, which may alter the relative values of the two strengths in question, and also by the orientation of the crystal. The various possible slip and cleavage planes in the different types of atomic lattice are given in Ref. 6, Table 1. Cleavage fractures have been obtained in many body-centred cubic metals, including of course alpha iron, in some of the hexagonal close-packed metals, but never in face-centred cubic metals. On a macroscopic level cleavage fractures are readily recognisable by the bright cleavage facets, the surface appearance being described as "crystalline" as opposed to the "fibrous" appearance of a shear fracture. Although a fracture may appear totally

crystalline there may well be some degree of plastic deformation because of the random alignment of crystals in a polycrystalline aggregate requiring bridges of shear fracture to link grains which have already separated by cleavage. Other plastic work is often revealed by surface markings and striations and by possible fracture arrests.

Thus for materials such as mild steel, which are subject to this shear-cleavage transition, it is quite possible that the two forms of transition may take place under similar conditions and are taken as being a single transition. Because of this the fibrous-crystalline transition in mild steel is often termed the "fracture mode transition", but to avoid any ambiguity in this thesis the term "fracture appearance transition" will be used instead.

2.2. Some Concepts of Fracture

One of the earliest fracture criteria is that of Ludwik (8) who proposed that fracture occurred when the maximum principal tensile stress reached a critical value called the strength. This simple idea is illustrated in Fig. 1 (a) which shows how the flow stress (true stress - true strain) curve intersects the fracture stress curve at a point X, in which case fracture will take place after prior plastic flow. Should the yield stress be higher than this ideal fracture stress, however, then fracture will take place at a stress given by the point Y without plastic deformation. It was later suggested by Davidenkov (9) that the ductile and brittle

fracture modes required two separate fracture stresses as indicated in Fig. 1 (b). This idea of two different types of fracture, a ductile fracture obeying a critical shear stress (or possibly a critical shear strain energy) criterion, and a brittle fracture obeying a critical normal stress criterion has been elaborated by workers such as Schnadt (10), and Hendrickson, Wood and Clark (see Section 2.3.). It would seem likely that some dual criterion of this type is in fact operative in the shear-cleavage transition in metals such as mild steel; but it is quite possible that only a single criterion is operative in the fracture mode transition of other materials and that the brittle fracture follows the same mechanism as the ductile, but with highly localised deformation. The dual criterion is used in Ref. 4, Fig. 2.9. to show how either cleavage or shear fracture may occur after either limited or gross plastic flow.

If, for steel under conditions of complex stress, both yield and shear fracture are taken to follow a critical shear stress criterion, whilst cleavage fracture follows a critical normal tensile stress criterion, then the loci on a three dimensional stress plot would be hexagonal prisms for the former and a cube for the latter. To examine this in two dimensions σ_2 is made zero and some of the alternative arrangements are shown in Fig. 2, these being (a) Yield followed by shear fracture, (b) Yield followed by cleavage fracture, and (c) Cleavage fracture before yield. If, however, the third stress component is increased from zero to some tensile value then Fig. 2 (a) (which

is typical for mild steel at room temperature) is altered drastically to a form such as (d). Thus the importance of a notch, which can produce a stress system approaching triaxiality, may be clearly seen. It must be borne in mind when considering these ideas on fracture, that in practice fracture is always preceded by some degree of plastic flow and the idea of a fracture stress as such is merely a convenient concept. We may now examine the effects of some of the variables in the light of these concepts of fracture.

(a) The Effect of Temperature

It is well-known that structural changes can take place in metals when the temperature is varied and it is fortunate that steel, which proves so complex in many ways, does not undergo any structural change as the temperature is decreased below ambient. The only exception to this can be the transformation of retained austenite into martensite at low temperatures. It is also well-known that decreasing the temperature has the effect of raising the entire flow curve, increasing the discontinuous yield (i.e. the difference between the upper and lower yield stresses), increasing the Luders extension, and increasing the stress required for fracture (11, 12). It is not so easy to obtain the variation of the fracture strengths with temperature because (a) When fracture occurs by the shear mode it usually does so after quite considerable plastic flow and thus usually after work hardening, and also the fracture takes place at some type of necked section

in which the multiaxiality of stress may be hard to estimate, and (b) When fracture occurs by the cleavage mode in simple tension it only does so at very low temperatures and this strength is strongly affected by prior cold work. It may be stated, however, that there is now severe doubt as to the validity of the original idea that the cleavage fracture stress is constant with temperature. A further complication is introduced by the fact that in many metals, including steel at the lower temperatures, twinning is a preferred mechanism to slip and it is not at all certain whether or not twinning may be said to follow a critical shear stress law.

Recent work by Hahn et al. (13) gave a comprehensive picture of the yield and fracture behaviour of steel and also gave definitions of the various stages of transition which may take place. Fig. 3 shows a typical example of their tensile test results and shows how they have divided the results into six temperature regions, which were described as follows:-

In region A the fracture obtained was a typical cup and cone after 50% to 60% reduction in area. A fibrous crack was initiated in the centre of the necked portion and grew radially until, as it approached the free surface, final separation occurred by "shear". (This description of such a 45° fracture is usually employed even though each grain still breaks in tension and it is very doubtful whether in fact one surface slides across the other),

In region B the fracture appearance changed to a fibrous centre surrounded by a rim of cleavage. Fibrous fracture again

started at the centre, but now the fibrous crack, by reducing the effective cross-sectional area and providing an internal notch, was sufficient to produce cleavage. The fibrous crack may thus be regarded as a critical crack size, becoming smaller as the temperature was decreased. A 50% fibrous transition temperature, T_F , was designated as a comparison with that in the standard V-notch Charpy impact tests, T_f .

Region C. The boundary of this region with region B was marked by a discontinuity in both ductility and fracture stress, referred to as the ductility transition, T_d . Immediately below T_d fractures which appeared to be 100% crystalline occurred after 15% to 20% elongation, there being a considerable difference between the yield and fracture stresses. The investigators noted here that the Ludwik-Davidenkov concept that fracture precedes yielding below the transition temperature was not consistent with these findings. They further stated that T_d occurred when the critical fibrous crack size approached the size of cleavage microcracks and implied that microcracks nucleated cleavage cracks below T_d . Although the number of microcracks increased sharply below T_d , they were also observed at temperatures above T_d which casts doubt on Cottrell's hypothesis that the ductility transition occurs when the formation of microcracks (i.e. the initiation of cleavage within the grain) first becomes possible. The alternative suggestion was made that the ductility transition occurs when the requirements for the growth of a microcrack into adjoining grains, and thus the propagation

to fracture, were satisfied.

In region D, cleavage fracture took place at the lower yield stress after some discontinuous yielding. Microcracks were observed only in yielded regions behind the Luders front and it was suggested that gross yielding was a pre-requisite of crack initiation.

Region E was regarded as an extension of region D with cleavage fracture occurring in the first part of the specimen to undergo discontinuous yielding without any observed load drop. No microcracks were observed, presumably because the first crack formed propagated to fracture.

In region F, fracture occurred abruptly at stresses which fell below the level of the extrapolated yield stress curves and possibly below the elastic limit. It was suggested that fracture in this range coincided with the onset of twinning.

(b) The Effect of Strain Rate

As early as 1904 Considère (14) noted that the effect of strain rate on brittle fracture was principally one of raising the temperature at which brittle fracture occurred. It has been shown that the speed of loading only affects the position of the transition curve along the axis and has no influence on the mechanism of fracture, i.e. ductile specimens deformed in slow-bending and in impact gave exactly the same mechanism of fracture, nor was there any change in mechanism between brittle fractures performed at the two rates (15). A unique relationship between temperature and strain rate has been suggested (16) and other proposals include a velocity modified

temperature concept (17) and an equivalence between temperature, strain rate, and state of triaxiality (18). The majority of experimental work has been performed in uniaxial tension and Biggs (5a) has summarised the information available on the effect of strain rate on mechanical properties as follows:-

- (I) Very slow loading eliminates the sharp yield point in steels.
- (II) With increasing strain rate, (i) The yield stress increases rapidly, (ii) The Luders strain increases, (iii) The maximum tensile stress increases, (iv) The value of the yield/maximum stress ratio increases, the rate of increase becoming more marked at higher strain rates.

Some attempts have been made to put the variation of yield stress with temperature and strain rate on a quantitative basis and these are briefly presented in Appendix 'B' together with references to relevant experimental work. With regard to other material properties, Austin and Steidel (19) showed that both the ultimate tensile stress and the actual fracture stress (i.e. fracture load/cross-sectional area at fracture) increased with the strain rate at fracture up to strain rates of 20,000 per sec. They also showed that the percentage elongation and the percentage reduction in area increased at first, but tended to become constant at strain rates above about 10,000 per sec.

At impact rates of loading other effects must be considered.

(i) The fact that stresses within a specimen may be governed by stress wave conditions and inertia effects. This will be discussed at a later stage with reference to the instrumented Charpy impact test (see Section 5.3).

(ii) The highly localised heating which can occur in a nominally isothermal test will lower the yield stress in that region, causing a local instability of the plastic zone.

One way in which these basic concepts of fracture have been employed to predict a ductile-brittle transition will now be shown.

2.3. The Theory of the Initiation of Brittle Fracture in Mild Steel as proposed by Hendrickson, Wood and Clark (22)

Using a hydro-pneumatically operated rapid tensile-testing machine (20) the above writers obtained upper yield stress values for a given annealed steel as a function of applied constant stress rate, $\dot{\sigma}$, at different temperatures (21). This is shown schematically in Fig. 4b. The same machine was then used to test cylindrical tensile specimens of the same steel, containing an hyperbolic notch. Fracture loads were recorded for those specimens which exhibited little ductility and the nominal fracture stress, σ_{nf} (fracture load/minimum cross-sectional area), was calculated. The type of results obtained is shown in Fig. 4a. Also a theoretical elastic-

plastic analysis was developed for the specimen and the maximum principal stress, $(\sigma_a)_{\max}$, was calculated for different values of the nominal stress, σ_n (see Fig. 4c). It was found that $(\sigma_a)_{\max}$ occurred in the axial direction on the elastic-plastic boundary. The analysis was only valid when the plastic zones were confined to the region of the notch root and thus there was a limiting value for the nominal applied stress, σ_n , which was found to be $\sigma_n/\sigma_{yd} = 1.5$. For $\sigma_n/\sigma_{yd} > 1.5$ the zones of plastic deformation spread completely across the body with very little further increase in applied load, and since the relaxation type of solution employed was only valid for small increments in strain, the analysis could no longer be employed. To confirm the analysis the fractured specimens were carefully sectioned, polished and etched, then a microhardness survey was carried out, all hardness readings being taken in regions of ferrite grains. Quite fair agreement was obtained between the theoretical and experimental elastic-plastic boundaries.

The writers then proposed that a "brittle" fracture be defined as one which takes place while the plastically deformed regions are confined to the vicinity of the notch root, i.e. while $\sigma_n/\sigma_{yd} < 1.5$. The maximum principal stress at fracture could then be calculated as follows:-

Let fracture occur at a nominal stress of σ'_{nf} when a

nominal rate of stress rise $\dot{\sigma}_n'$ is applied to the notched specimen at a temperature T_1 , as shown in Fig. 4a. Now the rate of increase of the axial stress at the root of the notch during the period of purely elastic deformation is equal to the product of the nominal rate of stress application, $\dot{\sigma}_n'$, and the axial stress concentration factor, k . Thus the axial stress rate at the notch root is $k\dot{\sigma}_n'$ and we may use this value in Fig. 4b to obtain the corresponding yield stress value σ_{yd}' . Now since we know σ_{yd}' the ratio $\sigma_{nf}'/\sigma_{yd}'$ is known, and if this ratio is less than 1.5 then the ratio $(\sigma_a)'_{\max.}/\sigma_{yd}'$ may be determined from Fig. 4c. Hence the maximum principal stress, $(\sigma_a)'_{\max.}$ may be found.

In their discussion of the results obtained, the writers state that the values of $(\sigma_a)_{\max.}$ obtained are constant within the probable experimental and theoretical errors, namely within the range 200,000 to 226,000 p.s.i. with an average value of 210,000 p.s.i. They further state that this critical stress is independent of temperature and rate of loading. (If, however, the results are plotted against strain rate as in Fig. 5, it would appear that at -129°C there is a small strain rate effect and a tentative extrapolation would indicate that at low stress rates $(\sigma_a)_{\max.}$ tends to become constant at about 199×10^3 p.s.i.)

The writers propose that a necessary and sufficient

condition for the occurrence of brittle fracture in notched steel specimens is that a critical value of the tensile stress is reached before plastic deformation extends to the axis of the specimen. Thus the transition from ductile to brittle behaviour, for the specimens tested, will occur when $(\sigma_a)_{\max.} = 210,000$ p.s.i. at the instant when $\sigma_n/\sigma_{yd} = 1.5$. Now from the data for Fig. 4c the value of $(\sigma_a)_{\max.}/\sigma_{yd}$ corresponding to this value of σ_n/σ_{yd} is $(\sigma_a)_{\max.}/\sigma_{yd} = 2.34$. Therefore substituting for $(\sigma_a)_{\max.} = 210,000$ p.s.i. gives $\sigma_{yd} \doteq 90,000$ p.s.i. This critical value of yield stress may be used in Fig. 4b to determine, for a given temperature, a critical stress rate at which a ductile brittle transition will occur. In fact this method accurately predicted the critical stress rate for temperature T_2 , but it also predicted a transition at a certain stress rate at temperature T_1 , and in fact no such transition was found. Hardness measurements, however, did indicate some degree of ductility transition in the predicted region.

The writers also successfully predicted the Energy Transition Temperature for Izod specimens made of the same material (23). A similar analysis, using a relaxation solution, was carried out on the Izod specimen. The point of incipient plastic instability was found to occur when $\sigma_{nb}/\sigma_{yd} = 0.80$, where σ_{nb} = Nominal bending stress at the root of the notch. The analysis also showed that, at this applied stress ratio, the maximum stress was given by $\sigma_m/\sigma_{yd} = 2.41$.

The ductile brittle transition thus occurred when σ_m was equal to the critical stress for the material (210,000 p.s.i.). Hence the critical value of the yield stress, $(\sigma_{yd})_c = 87,000$ p.s.i., could be calculated. After making allowance for the indentation of the specimen by the striker the stress rate at the root of the notch was calculated to be 9×10^9 p.s.i./sec. By extrapolating the lines in Fig. 4b to this value of $\dot{\sigma}$ a graph could be plotted of σ_{yd} against temperature for $\dot{\sigma} = 9 \times 10^9$ p.s.i./sec., this is shown in Fig. 4d. The temperature corresponding to $(\sigma_{yd})_c = 87,000$ p.s.i. is the transition temperature. This temperature coincided with a discontinuity in the Energy - Temperature curve and corresponded to an energy absorption of 20 ft. lb., an altogether astonishingly good agreement in view of the various assumptions and approximations which were made. Microhardness measurements again seemed to confirm that regions of plastic deformation were confined to the vicinity of the notch at a temperature at the lower end of the transition region, but it will later be seen that there is some doubt as to the validity of these microhardness measurements.

2.4. The Nucleation of Fracture

The simplest idea of the strength of a solid is the cohesive force holding together two layers of atoms. Using this idea approximate theoretical strengths have been calculated on the assumption that failure will occur when the atoms are displaced by one half of the atomic distance, but when these

values were compared with experimentally determined values it was found that the theoretical values were larger by about two orders of magnitude (24). This was found to be true both when a tensile separation and when a slip due to shear was considered.

Griffith (105) suggested that, although stresses of the magnitude predicted by theoretical considerations did not exist throughout the material, it was possible that they could be attained locally at stress concentrations such as microcracks. He then proposed that the strain energy released at fracture per unit crack extension be equal to the energy per unit area required to form two new surfaces giving:-

$$\sigma^2 = \frac{4ET}{\pi a} \quad (2.1)$$

where

σ = Mean applied stress

E = Modulus of elasticity

2a = Crack length

T = Energy/unit area required to create a new surface
= $\frac{1}{2}$ Surface tension

Whilst this theory appeared to be valid for truly brittle materials such as glass, it cannot account for the types of brittle fracture obtained in metals (or even in many brittle plastics) in which a small, but by no means negligible, amount of plastic deformation takes place. Methods adopted to allow for this plastic work will be described in Section 2.5.

In recent years the dislocation theory of slip has been much advanced and hypotheses have been developed to explain such phenomena as the yield points and strain ageing. There is not, however, universal agreement as to the relative importance of factors, such as dislocation locking by solute atoms and the lattice friction stress, when considering the movement of dislocations. Various theories have been advanced to explain and to predict the transition temperature on a basis of dislocation theory, but probably that of Cottrell (26) has aroused most interest. Cottrell does not distinguish between ductile and brittle fracture in the initiation stage but suggests that it is the amount of plastic deformation which accompanies the crack growth to the critical size which is the relevant factor.

2.5. The Propagation of Fracture

Quite a large amount of information may be gathered regarding the propagation of a fracture by observation of the fracture surface. Even the fracture surface of glass is not entirely smooth, escarpments and surface markings usually being visible. The fracture of a sharp notched brittle plastic can be clearly divided into several zones such as the highly reflecting mirror zone at the initiation. Clear surface markings such as River lines and Wallner lines are often visible and an experienced observer can often determine

the velocity of the crack front at a given stage of propagation by the shape of the surface markings, which are functions of the ratio of the crack front velocity to that of transverse waves.

Conditions become more complex in a polycrystalline substance, in which it is found that the direction of propagation in adjacent grains is rarely the same either as each other or as the direction of propagation of the general fracture front. In fact fracture in individual grains may fracture backwards (i.e. in the opposite direction to the main crack front). Tipper (2b) has shown that the fracture front is in fact curved and of irregular outline. Moreover the more ductile the material the more attenuated is the apex and the wider the angle between fracture planes.

One of the most obvious features of brittle fractures in mild steel, whether in wide-plate tests in the laboratory or in service fractures, is the clear chevron pattern (see Ref. 6 Fig. 27). Tipper has compared a propagating fracture in a plate to the repeated fracture of a series of small tensile test pieces, with centres along the centre line of the fracturing plate, and with striations radiating from the centres. Ref. 5 Fig. 128 shows how a series of initiations of this kind can produce a parabolic front of the type observed. Although the chevron pattern is characteristic of the brittle fractures of mild steel in which a large proportion of the grains have fractured by cleavage, it is by no means confined to

this material. Chevron patterns have been observed in some form in aluminium alloys, glass and plastics. Also brittle fractures showing chevron patterns have occurred in austenitic alloy steels and have even been intergranular.

Thus whatever the material it is clear that there is always some degree of plastic flow involved in the propagation of fracture and the simple Griffith criterion of Section 2.4 cannot often be applied. Two methods which have been evolved to account for the plastic work of propagation are explained below:-

(a) Orowan's Theory (for crystalline materials)

Orowan (27) suggested that extremely small cracks are present in every grain and are widened by plastic deformation near the tips as the deformation progresses. It is proposed that the growth of such a crack continues until either the crack reaches a critical size and the Griffith criterion is satisfied, or until it reaches a grain boundary. Because of lattice discontinuities and possibly the segregation of impurities at the grain boundary, it is considered that far more energy is required to propagate a crack across a grain boundary than within the grain. Orowan believed, therefore, that in general the length of a microcrack was of the order of the grain diameter and has no particular relevance to the fracture problem. Instead of a critical crack length he suggested that the condition for brittle fracture was the condition for transcrystalline crack acceleration.

$$\sigma = \left(\frac{EP}{a} \right)^{\frac{1}{2}} \quad (2.2)$$

where P is the largely plastic work of crack wall formation which is far larger than the surface energy associated with the pure Griffith crack.

(b) The Griffith-Irwin Theory and an Outline of Fracture Mechanics

The large amount of published work in the field of fracture mechanics (chiefly by Irwin and his collaborators) has been somewhat complicated by a certain amount of confusion over notation. Surveys of published work have been made (28, 29), however, and these help considerably in understanding the basic ideas of the theory.

Irwin replaced the surface energy term in the Griffith theory by a plastic work term, the Fracture Toughness, G_c , which is the total work irreversibly absorbed in creating a unit area of fracture. Denoting the strain energy release/unit area by G, the criterion for crack propagation was suggested as

$$G \geq G_c \quad (2.3)$$

Another approach to the Griffith-Irwin theory is based on a knowledge of the elastic stress distribution around a crack. For small plastic zones confined to distances from the crack surface small in comparison with the crack length, the theory of elasticity is considered to be applicable.

The stresses very near the root of the notch have been shown by Irwin (30) using Westergaard's stress function

solutions(31) to be:

$$\sigma_x = \frac{K \cos. \theta/2}{\sqrt{2\pi r}} (1 - \sin. \theta/2. \sin. 3\theta/2) - \sigma_{ox} \quad (2.4)$$

$$\sigma_y = \frac{K \cos. \theta/2}{\sqrt{2\pi r}} (1 + \sin. \theta/2. \sin. 3\theta/2)$$

$$\sigma_z = \frac{2\nu K \cos. \theta/2}{\sqrt{2\pi r}} - \nu \sigma_{ox} \quad (\text{plane strain})$$

or $\sigma_z = 0$ (plane stress)

$$\tau_{xy} = \frac{K \cos. \theta/2}{\sqrt{2\pi r}} \sin. \theta/2 \cos. 3\theta/2$$

$$\tau_{yz} = \tau_{zx} = 0$$

where

r and θ are polar co-ordinates starting from the tip of the crack, K and σ_{ox} are stress field parameters.

σ_{ox} is relatively small in relation to σ_x close to the crack and for this and other reasons it is usually neglected. Thus K is the only remaining stress field parameter which relates the magnitudes of the stresses to the applied loads and specimen geometry. K is called the stress intensity factor (not to be confused with the stress concentration factor) and it has been evaluated for various cases of interest. For an edge notched plate, width W , notch depth on each side = a , subjected to a uniform stress σ remote from the notched section.

$$K \doteq \sigma (W \tan \pi a/W) \quad (2.5)$$

Irwin later modified this formula to satisfy certain boundary conditions and suggested

$$K = \sigma (W \tan. \pi a/W + 0.1 W \sin. 2\pi a/W)^{\frac{1}{2}} \quad (2.6)$$

He later added a further correction to allow for the plastic zone ahead of the crack giving:-

$$K = \sigma (W \tan. u + 0.1 \sin. u)^{\frac{1}{2}} \quad (2.7)$$

where $u = \frac{\pi}{W} (a + K^2/2\pi\sigma_{yd}^2)$

This correction was obtained by substitution of the yield stress σ_{yd} for σ_y in equation (2.4) along the x axis to find r_{yd} the radius of the plastic zone.

$$r_{yd} = \frac{K^2}{2\pi\sigma_{yd}^2} \quad (2.8)$$

Irwin has implied, without in fact making a direct statement, that a criterion of fracture is $K = \text{constant} = K_c$ at fracture. This may be interpreted from equation (2.4) as implying a maximum nominal elastic stress criterion of fracture, but other physical interpretations are possible. It should be emphasised that K is a stress field parameter independent of material whereas K_c is a material constant. In fact K_c was thought to be a "variable constant", since for a given material it is independent of test-piece width and crack depth, but is dependent on plate thickness as well as more obvious factors such as temperature and material condition. The thickness effect is probably related to the degree of plane stress or plane strain in the stress field at the crack tip.

Using a hypothetical work done in closing a crack, Irwin was able to relate the energy and the stresses and developed the following relations between K and G (and K_c and G_c):-

$$\left. \begin{array}{l} \text{(a) Plane Strain } K^2 = \frac{EG}{1-\nu^2} \\ \text{(b) Plane Stress } K^2 = EG \end{array} \right\} \text{_____} \quad (2.9)$$

It has been suggested that the K_c value associated with fracture under plane strain conditions, designated by K_{Ic} , is perhaps a true material constant. In fact experimentally derived values seem to fall between the two extremes of plane strain and plane stress.

Unfortunately recent work (32 , 33) has indicated that K_c is not only dependent on the specimen thickness, B , but also on the specimen width, W . Within certain limits an increase in the W/B ratio results in a larger shear lip fraction (see the following section, "The Fracture Mode Transition") at instability and hence a larger K_c value. The variation of K_c with plate width can be avoided by stipulating $\sigma \leq 0.8\sigma_{yd}$ as a test requirement, but this immediately excludes many investigations of the greatest interest and importance. Tests on aluminium have shown that if the W/B ratio is varied so that the nominal stress at fracture is changed through the range $0.8\sigma_{yd} < \sigma_{nf} < \sigma_{yd}$ then K_c is increased by 15% to 20%.

It is possible to examine this idea graphically if the G_c concept is employed. If we now take R as the crack growth

resistance and G_c as the critical value of G at instability, then Fig. 6 shows how it has been suggested that G_c is given by the point of tangency. At first sight this diagram may appear unstable in that, at low values of crack length $G > R$. But if we consider a specimen possessing some notch or crack of length a , then at low stresses the curve of G will be as '1'. As the stress is increased the G curve will swing up through '2' to '3', at which stage it is suggested a stable extension of the crack takes place. The value of R may increase to some value R_3 . Further increase in stress will swing the G curve still higher and increase R in some fashion until the two curves become tangential, giving a condition of instability. Any variation of G_c would seem to be reflected in the shapes of the two curves. If it is possible to cause the R curve to be shifted bodily along the crack-length axis, then the position of the point of tangency will be very much affected by the shape of the R curve, a rounded curve allowing considerable variation in G_c whilst a more squared off curve would keep G_c almost constant.

The Fracture Mode Transition

Three modes of cracking can be envisaged, namely, (1) opening, (2) sliding and (3) screw sliding. In notched tensile fractures of sheet materials "oblique shear" (45°) and "transverse tensile" (90°) fractures have been observed and

these correspond with mode (3) and mode (1) respectively. This has appeared to be a plate thickness effect, thick plates tending to fracture by mode (1), thin plates by mode (3) and plates of intermediate thickness by a mixture of the two modes, having a central zone of 90° fracture with oblique shear edges.

It has been suggested that the type of fracture mode transition under discussion is governed by the size of plastic zone associated with the crack tip, in relation to the plate thickness. If the plastic zone is approximately equal to the plate thickness, then plastic shearing in the thickness direction along planes at 45° to the plate surface is possible through the whole plate thickness and subsequent 45° fracture is possible. With a smaller plastic zone 45° planes starting at the plastic zone boundary on opposite plate surfaces would not join up in the middle of the plate thickness. The type of mechanism suggested is shown in Boyd's diagram, Fig. 7.

The Effect of Plate Thickness

The size of the plastic zone at the onset of rapid fracture is approximately twice its radius and thus, from equation (2.8) is

$$\frac{K_c^2}{\pi \sigma_{yd}^2} \quad \text{-----} \quad (2.10)$$

Irwin (46) made another estimate considering the principal stresses rather than the cartesian components, the maximum principal stress occurring at $\theta = 60^\circ$. With σ_x small he equated this maximum principal stress to σ_{yd} giving a plastic zone size

$$\frac{1.64K_c^2}{\pi\sigma_{yd}^2} \quad \text{-----} \quad (2.11)$$

He also suggested that expressing the plate thickness B as

$$B = \lambda \frac{K_c^2}{\sigma_{yd}^2} \quad \text{-----} \quad (2.12)$$

then a value of λ decreasing from 1.5 to 0.5 corresponded to the transition from 90° to 45° fractures. Irwin's experimental plot of K_c against $1/B$ is shown in Fig. 9. The low values of K_c for thick plate correspond to K_{ic} for plane strain 90° fractures and a rapid increase in K_c is shown as the plate thickness decreases to give plane stress 45° fractures. The dotted line on the diagram is a plot of

$$B = \frac{1}{\pi} \left(\frac{K_c}{\sigma_{yd}} \right)^2 \quad \text{-----} \quad (2.13)$$

which it is suggested divides the plane stress and plane strain regions and may be relevant to the 45° to 90° fracture transition. Irwin also suggested an empirical relationship between K_c , K_{ic} and λ of the form

$$K_c = K_{ic}(1 + 1.4/\lambda^2) \quad \text{-----} \quad (2.14)$$

A rational explanation for this observed characteristic has been advanced by Bluhm (34) who made the following assumptions. (1) The total fracture toughness (integrated over the fracture surface) derives from a dual fracture mode mechanism i.e. either a combination of a flat break (which may be fibrous or crystalline) and a shear lip. (2) The shear lip occupies the full sheet thickness (for a given temperature) until a critical thickness t_0 is reached; beyond this critical value the total lip width no longer increases (see Fig. 8). (3) Once fully developed the shear lip remains constant in width along the length of the fracture path, apart from end effects at onset and arrest (such as are found in the Charpy specimen). (4) The mechanism of flat fracture is essentially a surface phenomenon (energy of fracture $\propto (t - t_0)$ the width of the flat area), whereas the mechanism of lip formation is volumetric in nature (energy of fracture $\propto (\frac{1}{2}t)^2$ until the lip is fully developed and is then a constant).

The last assumption may be expressed algebraically as:-

$$E_{sl} = \frac{1}{2} k_{sl} t^2 L \quad t \leq t_0 \quad \text{-----} \quad (2.15)$$

and

$$E_{sl} = \frac{1}{2} k_{sl} t_0^2 L \quad t \geq t_0$$

$$E_{ff} = k_{ff}(t - t_0)L$$

where:-

The subscripts 'sl' and 'ff' refer to shear lip and flat fracture respectively. The E's are energy values and the k's

are material constants associated with the mode of fracture.

L is a characteristic length determining the total energy.

The total energy, $E_T = E_{sl} + E_{ff}$ _____ (2.16)

Now the total energy per unit nominal fracture area is identical with G_c and therefore,

$$G_c = E_T/Lt \text{ _____ (2.17)}$$

Substituting equations (2.15) and (2.17) into (2.16) gives:-

$$G_c = \frac{1}{2} k_{sl} t_o (t/t_o) \quad \text{for } t/t_o \leq 1 \text{ _____ (2.18)}$$

and $G_c = \frac{1}{2} k_{sl} t_o (t_o/t) + k_{ff}(1 - t_o/t)$ for $t/t_o \geq 1$

An arbitrary assumption of $k_{ff} = 1/20 k_{sl} t_o$ gave a good approximation to the observed results.

Effect of Temperature

The exact meaning of the temperature transition in this field is at present far from clear. The transition may be primarily governed by the changing yield stress altering the plastic zone or shear lip, or, on the other hand, some material property such as K_{ic} may itself be changed.

Experimental Verification of the Theory

Several methods of measuring K_c have been made (35) and most applications have been in the field of high strength steels for pressure vessels (36) or rocket motor casings. The fracture of some sharp-notched brittle plastics, namely polystyrene and polycarbonate, has also been shown to follow a law of the type

suggested by equation (2.6) and values of K_c have been determined experimentally for these two materials (37). These fractures all took place at stresses well below general yield and thus no variation of K_c with the W/B ratio had to be considered. Attempts have been made to apply the theory to the brittle (and usually cleavage) fracture of mild steel and the following section shows one of the most hopeful lines of argument in this direction.

(c) The Crack Opening Dislocation (C.O.D.)

It has been pointed out by Wells (38) that most assessments of critical flaw sizes for structural parts employ some type of stress intensity K criterion, whilst the relevant material property is usually assessed on the basis of some test of the V-notch Charpy impact type, in which fracture nearly always occurs after general yield. He proposed the following link between the two.

"The existence of a local plastic zone size at the tip of a crack, depending on the crack extension G (or stress intensity factor K) and the material yield strength σ_{yd} , implies a uniquely related crack opening dislocation 'b', depending only on the elastic modulus in addition. The corresponding crack opening dislocation in the Charpy specimen is given by the product of the angle of bend and the distance below the notch root of the plastic hinge axis. For limiting fracture conditions in the two cases it is proposed that there is a

proportional relationship between the two types of crack opening dislocation which is invariant with temperature. It is proposed that the proportionality factor differs from unity mainly because of the different degrees of triaxiality and straining rate between the testing conditions compared."

Wells used the simple approximations to the plastic zones as shown in Fig. 10, and assuming that the peripheral elastic stresses and strains at the yield boundary were at their simply determined limits he obtained:-

<u>Plane Stress</u>	<u>Plane Strain</u>
$\frac{b}{2\pi r} = \frac{\sigma_{yd}}{E}$	$\frac{b}{2(2\pi r)} = \frac{\tau}{c} = \frac{\sigma_{yd}}{2c}$
$\therefore b = \frac{2\pi r \sigma_{yd}}{E}$	$\therefore b = \frac{2\pi r \sigma_{yd}}{c}$

(2.19)

Wells also made another estimate of the C.O.D. using Westergaard's expressions (Equation (2.4)) for the crack opening displacement within the crack at any distance r from its tip.

$$\eta = \frac{2K}{E} \sqrt{\frac{2r}{\pi}} \quad (2.20)$$

Now on the yield boundary $r = r_{yd}$ is given by equation (2.8) (for plane stress) as

$$r_{yd} = \frac{K^2}{2\pi \sigma_{yd}^2}$$

substituting for $r = r_{yd}$ in equation (2.20) gives:-

$$b = 2 \eta = \frac{4K^2}{\pi E \sigma_{yd}} \left(= \frac{4G}{\pi \sigma_{yd}} \right) \text{ (2.21)}$$

If the above equation for r_{yd} is substituted into the simply determined value for plane stress, equation (2.19), then it is seen that equation (2.21) only differs from the simply determined value by a factor of $\frac{4}{\pi}$.

Wells concludes that the C.O.D. is strictly related to the crack extension force and stress intensity factor under locally plastic conditions but that the proportionality may depend on the degree of plane stress or plane strain. He suggested that, within these limitations, the C.O.D. might be regarded as an index of fracture toughness where fracture ensues only under conditions of full plasticity such that the local plastic zone size ceases to have any practical meaning. If slip line field solutions for the V-notch Charpy specimen are employed, then b may be calculated for an angle of bend θ° on the assumption that the two halves of the specimen rotate around the rigid core. Wells used a pure bending solution and obtained $b = 1.82 \theta \times 10^{-3}$ in. If a three point loading solution (40) is used instead then

$$b = 2.49 \theta \times 10^{-3} \text{ in. (2.22)}$$

This also is only an approximation, however, because the slip line field solution is considerably complicated when the notch root radius is considered.

This concept of a critical displacement criterion for

fracture has also been suggested in another context. It has been shown that under certain circumstances, including many cases of service failures, fractures do occur at stresses well below the level of general yield. For normal notch-tensile specimens, however, it has been found that fracture does not occur until the applied stress produces a general yield across the entire load bearing section embracing the root of the notch (41). To produce fractures before general yield in mild steel it has been necessary to "damage" the material at the notch root in some fashion such as pre-compression (42, 43) or nitriding. A concept involving an "Exhaustion of Ductility" has been introduced to explain this phenomenon. The following theory has been developed which uses the critical displacement concept to give one explanation of the above in relation to specimen size.

(d) A Macroscopic Theory of the Initiation of Fracture, by Cottrell (44)*

At temperatures above the nil-ductility temperature for a notched specimen the material at the notch root becomes ductile and the plastic strain in this region becomes much greater than the elastic strain. Such plastic strains can only be produced at the root of the notch if they are accommodated

*Note: Although a theory of fracture initiation, this theory has been included in this section as a comparison with the Crack Opening Dislocation concept.

by far-reaching plastic zones which extend to large distances from the notch. The location of such zones in notch-bend specimens has been shown by Crussard et al. (45) and by Knott, by etching with Fry's reagent. Rigid-plastic analyses on such specimens have been performed by Green and Hundy (39) and by Alexander and Komoly (40) and these predict similar zones. The type of deformation pattern obtained is shown in Fig. 12. The plastic zones consist of thin curved arcs which act as plastic hinges around the rigid central portion. It is possible to work the metal heavily at the notch root to the stage at which it undergoes ductile fracture and causes a crack to run into the unworked material behind the root, even though the total work remains small. If this running crack propagates with comparatively little further plastic work, then an essentially brittle fracture is obtained. If, on the other hand, the crack advance is accompanied by further far-reaching plastic zones, then the total plastic work will be high and the fracture ductile.

In a more recent paper (44), Cottrell has placed these factors on a somewhat more quantitative basis. He uses an analysis by Swinden et al. for a two-dimensional slit in an infinite body subjected to an overall shear stress σ_s , which produces yielded zones spreading symmetrically outwards in its own plane, as shown in Fig. 11.

Taking L and a as the distances from the centre line to

the ends of the plastic zone and of the slit respectively, u , the plastic displacement at the end of the slit, and measuring the applied shear stress in units of the shear stress for general yield, σ_{ys} ($=\mu/1000$), the results shown in Table I were obtained.

Table I - Spread of Yield Zones from a Slit Subjected to Shear Stress

σ_s	0.1	0.5	0.75	0.9	0.95	0.99	0.999	0.9999
$10^4 u/a$	0.16	4.4	12.1	23.6	32.4	53	82	112
L/a	1.01	1.41	2.6	6.4	12.8	64	640	6400

It can be seen that for small stresses the yield zones are localised but as σ_s approaches σ_{ys} , the yield zones spread to immense distances from the slit, in the limit merging into general yield. For small stresses the equations on which Table I is based reduce to the following, simpler, form:-

$$\sigma_s = \sqrt{\frac{2\mu \sigma_{ys} u}{\pi a}} \quad (2.23)$$

Now $\sigma_{ys} u$ is the plastic work done in shearing unit area of the yield zone to a displacement u . If we now identify u with the required displacement for ductile fracture then $\sigma_y u$ is P the effective surface energy of fracture and we may re-write

equation (2.23) as:-

$$\sigma_s \doteq \sqrt{\frac{\mu P}{a}} \quad (2.24)$$

This is essentially the Griffith-Orowan criterion for fracture with σ_s as the shear fracture stress (c.f. equation (2.2)).

It is necessary, if these results are to be applied to fracture initiation at a notch, to decide upon a critical value of u for fracture and Cottrell suggested a value of $u = .003$ in. Substituting this value of u into Table I gives the values of a and L shown in Table 2.

Table 2 - Estimated Stresses and Notch Sizes to Initiate Fracture from a Sharp Notch (assuming $u = .003$ in.)

σ_s	0.1	0.5	0.75	0.9	0.95	0.99	0.999	0.9999
a	180	7	2.5	1.3	0.9	0.6	0.37	0.27
L	182	10	6.5	8.3	11.5	38.5	240	1730

It is seen that in large structures, which are able to accommodate the plastic zones, fracture can occur at stresses below the general yield stress. Also with a large notch, say 2.5 in., if the structure is able to accommodate the 6.5 in. plastic zone, then fracture can occur at $\sigma_s = 0.75 \sigma_{ys}$. This would help to explain the observations of Wundt (46) that a 9 in. square bar containing a notch 1.8 in. deep and of notch root radius 0.001 in. fractured at approximately one half of the

general yield stress, whereas smaller specimens were able to withstand stresses equal to or greater than general yield before fracture. Cottrell also concludes from Table 2 that, for mild steel, although fractures below the general yield stress can occur, they cannot occur at normal service stresses unless the notches are so large as to be obviously detectable.

It was noted that the above mechanism is only true for fractures which are initiated in a ductile manner, if the tip of the notch is embrittled in some way far smaller values of u become possible and brittle fractures may be directly initiated.

Now the order of plastic strain at the notch root envisaged by Cottrell was very high; in fact, in obtaining the value of $u = .003$ in. he tentatively suggested a ductile fracture strain of 50%. This is a far higher value than was thought likely by earlier writers and it has been generally accepted that the surrounding elastic region limited the plastic strains to about 0.4% before the onset of general yield. Rendall (47) pointed out that it can be deduced from the results of Green and Mundy, also of Lequear and Lubahn (57) that for notches where $a/r = 40$ (r = notch root radius), the maximum plastic strain at general yield is not greater than about 3%. If the material under discussion is ductile in a normal tensile test at the same temperature, then it does not seem likely that ductile initiation of fracture will take place at 3% strain. For $a/r < 40$, however, there are few examples of

fracture before general yield at temperatures at which the material is ductile in a plain tensile test. In addition to Hendrickson, Wood and Clark, Rendall also mentions Lean et al. (48), but in both cases quite high strain rates were employed. A further point raised by Rendall was that Cottrell's calculations did not include a term for the thickness of the plate.

The above theory introduces one tentative explanation of a size effect; other factors involved will now be examined.

2.6. Size Effect

The existence of a size effect was noted at an early stage in the study of brittle fracture. Stanton and Eatson (62) were probably the first investigators to take care to obtain true geometrical scaling. They tested five sizes of Charpy impact specimen, geometrical similarity in the test piece and loading system being maintained throughout. Their results showed that the energy absorbed per unit area at the bottom of the notch exhibited a marked increase as the specimen size was reduced especially in the case of the 5 mm. square specimens. Docherty (63) confirmed that this effect was still present under conditions of slow bending (in this case Izod type specimens were used) and also noted that the energy absorbed was increased by the speed of loading, provided that the specimen remained ductile. It is interesting to note here that Docherty mentioned the fact that the cracking of the

largest specimens was extremely violent, the broken halves of the specimen jumping several feet from the testing machine. These papers and others have been reviewed by Tipper (6, 49) and by Wells (50).

It is desirable, before the effect of a notch is discussed, to consider the consequences of increasing dimensions in an unnotched specimen and some of the points mentioned in the above reviews are as follows:-

(1) Metallurgical Effects

Mackenzie (51) took a large number of samples of ship and structural quality steels and from the results of V-notch Charpy impact tests at a given temperature he found that the energy absorbed and the fracture appearance were closely dependent on the thickness to which the plate had been finally rolled. Ancillary investigations indicated that this effect of increased embrittlement with increased thickness was associated with the larger grain size associated with the thicker plates. The increased grain size is itself a consequence of the higher temperature at which rolling finishes and a slower rate of cooling from this temperature.

(2) Increase in the Number of Defects

Weibull (52) postulated that the distribution of defects in a material is such that a relatively small number of large defects and a large number of small defects exists. Thus for a given severity of distribution the probability that a particularly severe defect will occur in a specimen increases

with specimen size. Irwin (53) showed that the fracture size effect of brittle materials can be represented by a statistical distribution of flaws as proposed by Weibull. Also, since yield and fracture often begin on the surface of a specimen, the larger specimens having a greater surface area and thus a greater probability of surface flaws, will tend to fracture at lower stresses.

(3) Available Elastic Strain-Energy

The strain energy available for sudden yielding or for fracture depends not only on the stiffness characteristics of the testing machine, but also on the dimensions of the specimen. There is no doubt that the available strain energy plays an important part not only in the size effect in the ductile-brittle transition, but also in the stability of fracture in general. This will be seen later when unstable ductile fractures are discussed.

Although, so far, unnotched specimens have been considered, any consideration of defects in the material implies the introduction of a notch and the effect of notch type and geometry must now be considered.

The energy absorbed in standard impact tests decreases with notch sharpness until the root radius is about 0.010 in. (i.e. the V-notch Charpy impact test), beyond this point there is some uncertainty whether increasing the acuity has any effect. It has been suggested that the V-notch Charpy

specimen approaches very closely the most severe conditions that may be imposed in a standard impact bar (54). More recent investigations (55), however, have shown that a fatigue crack or arrested cleavage crack can lower the brittle fracture strength by about 35% compared with a 0.005 in. notch. This large difference was only observed on the large size of specimen and it was suggested that the difference might disappear altogether if relatively small specimens were tested. It must of course be remembered that the acuity of a given type of crack will not vary with the specimen size and thus geometrical similarity in notched specimens is no longer maintained once a crack has formed, the crack being effectively sharper in the larger specimens.

Lubahn, Sachs and their collaborators have published a series of papers on the effect of temperature and size on the notch strength and ductility of steels (e.g. 56 - 59) and these have been briefly summarised by Rendall (47). Lubahn found that at room temperature gradual tearing was followed by sudden or intermittent cleavage, depending on the size and geometry of the specimen. He also pointed out that both depth and width are necessary for catastrophic failure and that neither large width nor large depth alone is sufficient to cause this size effect, if the other dimensions are small. Lubahn took care to distinguish between a fracture which was initiated in a brittle manner and one which undergoes

considerable plastic flow, even at points remote from the notch bottom, before brittle crack propagation. The fracture energy considered in size effect tests of the Docherty type is of course the total energy and is dependent on both phases. Lubahn considered that there was no size effect as far as initiation of fracture is concerned (59). This was based on the fact that the ductility at fracture, as judged by direct measurements on a small gauge length at the notch bottom or by bend angles, was of the order of 45% for both $3\frac{1}{2}$ in. square and 0.4 in square specimens.

Later work by Lubahn and Yukawa (60), however, using a cycling type of slow bend test on geometrically similar Charpy V-notch specimens of nickel-molybdenum-vanadium steel found that the notch ductility at fracture decreased from about 30% to 3% when the specimen dimension changed from 0.2 in. to 9 in. This size effect was found to be more severe in the smaller size range.

Wells tested a series of geometrically similar notch-bend specimens and was able to confirm Docherty's results, i.e. suppression of brittle fracture in the small sizes with a gradual increase with size of the percentage brittle area in the larger specimens. He also used measurements of the temperature wave set-up by the fracture to calculate the surface energy, and obtained a good comparison with his calculated value. The surface energy values were found to decrease with specimen size.

We may examine size effect from the point of view of

fracture mechanics if we take an equation such as (2.1). Then at fracture,

$$K_c = \sigma (W \tan. \pi a/W + 0.1 W \sin. 2 \pi a/W)^{\frac{1}{2}}$$

This is not strictly a true size effect since unit thickness is maintained throughout, but it is seen that if the notch depth, i.e. the ratio a/W is kept constant then the value of σ at fracture will depend on W . For small values of W , σ will have to be large to give the desired value of K_c . It has already been noted that the direct applicability of an equation of the above type to the cleavage fracture of mild steel is in some doubt. Ludley and Brucker (61) used their standard pre-compression treatment (42) to obtain brittle fracture initiation at a nominal stress below the original yield stress in notched tensile test pieces. All the specimens tested were $\frac{1}{2}$ in. thick and 10 in. long, but the width was varied from $3\frac{1}{3}$ in. to 20 in. The results exhibited quite a large amount of scatter, but no size effect was exhibited by those plates which had a width of $6\frac{2}{3}$ in. or more. The $3\frac{1}{3}$ in. plates, however, did not fracture at all at nominal stresses up to yield and it was suggested that this could imply a cut-off size effect. They concluded that a Griffith-Irwin type of theory was not applicable. The exact significance of the pre-compression treatment, however, is by no means certain, and although such prior damage can undoubtedly make the initiation of a brittle crack possible at a much lower stress than would be possible in an undamaged plate, it cannot be categorically

stated that this treatment entirely eliminates the "initiation barrier" in mild steel.

If we consider the $G - R$ concept (section 2.5(b)) then it will be seen that whereas G is a strain energy term depending on D^3 (where D is a specimen dimension), R is a surface energy term depending on D^2 . Thus the instability point G_c will be reached at relatively lower applied stresses in the larger sizes. If, however, we consider Bluhm's suggestion then it is seen that E_{sl} is also a volumetric term and it is only E_{ff} which is $\propto D^2$. Thus since $k_{sl} > k_{ff}$ there would be little size effect on this basis until the temperature conditions were such that an almost entirely flat fracture was obtained.

In most of the theories of brittle fracture the yield stress is not considered to be a quantity which is dependent on the size of the specimen but only on such factors as stress system, strain rate, and temperature. Richards, however, has determined the yield stress for mild steel specimens tested in tension (64) and in bending (65) and from these results he has suggested empirical relationships between the yield stress and the specimen volume. In both cases there was a considerable drop in yield stress as the size was increased (e.g. tensile yield stress decreased from 60,300 p.s.i. to 53,550 p.s.i. when the specimen diameter was increased from $\frac{3}{8}$ in. to $1\frac{1}{4}$ in.).

When considering notched specimens, the yield stress is affected by another factor. X-ray measurements have indicated that the upper yield point of low carbon steels may rise abnormally high, up to two or three times the value of the lower yield point, if the high stress is concentrated in a very small volume such as the elastic stress concentration at the root of a notch (66).

SECTION 3 - PRELIMINARY TESTS

Samples of four Admiralty pedigreed steels were available for testing (A, B, U and V, see Ref. 67), and it was necessary to select two of these which were likely to differ in their behaviour with respect to strain rate in order that suitable stocks for a testing programme might be obtained. The most simple way of introducing a velocity effect was to employ the variable pendulum release angle facility of the available impact machine (see Section 5.1.). A few specimens of each steel were tested at different temperatures to give an outline of the transition region and a temperature was selected for each steel at a point just at the lower end of the transition range. A series of specimens were tested at this temperature, the initial angle of the pendulum being decreased for successive specimens until fracture failed to occur. Under these circumstances the pendulum rebounded and could be caught with the manually operated brake before a second blow took place. The usual measurement of energy absorbed was made for each specimen together with the expansion at the struck face and the contraction at the root of the notch. The results enabled the steels to be divided into two groups. With one group, the energy absorbed was approximately constant with decreasing initial angle until the kinetic energy available was insufficient to cause fracture. This is the type of behaviour exhibited by steel 'F'

in Fig. 22. With the other group, however, specimens would fracture with low energy absorption at high impact velocities, yet absorb up to three times as much energy without fracturing at low impact velocities, the behaviour being of the type shown in Fig. 23. It appeared that there might also be a difference in behaviour between the two steels as revealed by the lateral expansion measurements, but there was a fair amount of scatter present and the results were far from conclusive.

Unfortunately, with the available testing machine, it was not possible to vary the pendulum mass so that the kinetic energy available at impact might be held constant and only the impact velocity varied. Even if the observed difference between the two groups was only partially a velocity effect it was deemed a sufficient basis, in the absence of other evidence, to select two steels, namely 'A' and 'V'. The actual steels obtained were in fact the ordinary mild steel 'P', which was similar to 'A', and Plate 2997, which was steel 'V' but from a different melt than the original sample.

Similar tests were then performed on the new steels 'P' and 'V' and the results obtained are those in Figs. 22 and 23. These results do not in fact show the difference between the two steels as clearly as did the earlier tests and are presented in preference to them mainly as a logical addition

to the other results which are all on the same two steels. In particular, the original sample of steel 'A' did not show the "valley" in the curve of % lateral expansion against velocity which is seen at low velocities in the curve for steel 'P'. The former was, however, tested at a lower temperature ($-30^{\circ}\text{C}.$).

SECTION 4 - SLOW NOTCH-BEND TESTS

4.1. Apparatus

The slow notch-bend tests on all three sizes of specimen were carried out in a standard 10 ton Denison testing machine, which had the advantage of possessing a number of comparatively low load ranges (0.5, 1, 2, 5 and 10 tons) together with a fairly rigid construction and a large loading area. The anvils and insulated bath were located on the lower, weighing, cross-head whilst the striker was fixed to the upper, movable, cross-head. The Denison testing machines are sufficiently well-known not to require a full description and it need only be said that the hydraulic system consisted of a variable delivery pump with coarse and fine by-pass valves connected in the input line to the loading rams. The desired rate of cross-head movement was obtained by adjusting the rate of pumping. Once the appropriate rate of pumping had been set, the coarse by-pass valve was the only control required. When partially open this allowed the cross-head to remain stationary whatever the rate of pumping and, having a simple lever control, it could be closed quickly and conveniently to start the test. The fine by-pass valve, which is employed for most routine testing, was only required for calibration tests. For the bend tests it was found that a more uniform rate of cross-head movement was obtained if the by-pass valves were both kept closed during

the test. Another useful machine facility was a cross-head movement limit switch which was adjusted to switch off the motor just after the specimen had been pushed through the anvils, so that there was no danger of damage occurring by the specimen being compressed against the anvil block.

The basic slow bend jig for standard specimens is shown in Fig. 13. The striker and the anvil inserts were made of "Microlin" quenched and tempered to approximately 500 D.P.H. The anvil inserts were bolted into an anvil block of mild steel, which was itself bolted to the base of an insulated bath. To prevent excessive heat loss by conduction to the lower cross-head, the bath was placed on a block of "Sindanyo", which was located directly beneath the anvil block. A clamping device bearing on the vertical tie rods of the testing machine allowed accurate adjustment of the anvil position with respect to the striker. The overall loading system is shown in Plate 1.

The loading system in the slow-bend tests was required to resemble, as closely as possible, that in the Charpy impact test. Unfortunately there are no universally accepted standard dimensions for Charpy impact machines and dimensions such as the nose radius of the striker may vary considerably. The Charpy impact machine which was available will be discussed in greater detail later, but its dimensions were found to be in accordance with

B.S.131, 1933. Thus the standard size jig was made to the same specification*. To keep geometrical similarity when testing double and half size specimens, the striker nose radius, anvil insert radius, and the distance between the anvil inserts were doubled and halved respectively. The anvil inserts for the half size specimens were designed to fit into the standard anvil block, but the inserts for double size specimens required a more massive anvil block. Three separate strikers with the appropriate nose radii were required.

To ensure accurate location of the specimen two locating stops were provided, (i) a back-stop which located the specimen at the mid-width of the striker, and (ii) an end-stop which ensured that the specimen was struck immediately opposite the notch. The latter stop was in the form of a pin which stood slightly proud from one of the anvil inserts. Care was taken not to use too large an end-stop in case of interference with the free bending of the specimen. With the aid of these stops and a locating tool, specimens could be quickly and accurately located even when the anvil itself was obscured by the liquid in the bath or by vapour, which was generally formed at the lower temperatures.

*Note: The British Standard Charpy impact testing machine dimensions have recently been altered (B.S.131: Part 2, 1959).

(a) Instrumentation

(i) The Autographic Recorder

For the original slow notch-bend tests on standard size specimens, the rate of loading was considered sufficiently low to employ a mechanical recorder, and the autographic recorder fitted to the testing machine was modified accordingly. The standard recorder used a system of belts and pulleys to convert the extension on a given specimen gauge length into a certain rotation of a drum, which was fitted with small dowels to engage in holes in a paper chart. The paper feed was from a reel at the back of the recorder. The load was indicated by a pen clipped to the end of a light rod, which was attached by a pin joint to the final lever of the testing machine weighing system in such a way that the total pen travel across the graph paper was caused by full scale deflection on the machine dial. The pen movement was constrained by guides which gave a load axis parallel to the axis of the drum. For the slow notch-bend tests, however, the movement of the striker and hence the cross-head was required and to provide this the cross-head position indicator, which was fitted to the machine, was utilised. The upper cross-head carried with it a rack which engaged with a pinion fixed to a short shaft, the rotation of which was converted into horizontal movement by means of a screwed block sliding in guides and was finally

indicated on a dial gauge. To drive the recorder, an eight inch diameter pulley was fitted on to this short shaft and connected to the recorder drum shaft by a flat belt drive. The original extensometer fitting was not removed, however, because it incorporated a tensioning device for the drive to the recorder drum and this was still required and can be seen in Plate 1.

The drive from the new pulley had the effect of rotating the drum in the opposite direction to its previous motion and because of this it was necessary to provide a device to rotate the storage reel and take up the slack in the paper. This was done very simply by wrapping a cord around the spindle of the chart reel and hanging a small weight on the end.

No calibration of the load scale was needed because the pen travel was directly proportional to the testing machine dial reading. The testing machine itself had recently been checked against a dead weight load tester and no further calibration was thought necessary except to ensure that the chart and machine dial readings did in fact coincide. The deflection scale was easily calibrated directly against a dial gauge mounted on the anvils with the plunger bearing on the top cross-head. The deflection of the cross-heads and, more important, the various parts of the bending jig, were taken into account by loading on a rigid specimen and taking a

load-deflection record, thus in effect giving an inclined load axis.

It was obvious from the records that the mechanical recorder was in error when a sudden drop in load occurred due to a brittle crack and it was suspected that at the fastest rate of slow-bending (12,000 lb./sec.) the recorded load might in fact be lagging.

A load cell employing resistance strain gauges was constructed and calibrated, and, as will be described later, true load-time curves were obtained. A constant speed motor with reduction gearing was arranged to drive the recorder drum and thus records of the mechanically indicated load against time could be obtained. As a check on the paper speed, a time marker was fixed to the recorder producing time marks at $1/5$ th. second intervals. Thus a true load time curve could be superimposed on the indicated load time record. The results of slow-bend tests on two Charpy specimens at different rates of loading are described in Appendix 'D'. It was apparent from these tests that the diagram at the low rate was fairly accurate and it was possible to apply a correction factor to those records already taken. At the higher loading rate, however, a quite considerable error was introduced and the recorder in its existing form was not suitable. It was decided to investigate the vibration characteristics of the final lever and weighing system of

the Denison Testing Machine.

By the methods described in Appendix 'D' the response of the recorder was greatly improved and gave no apparent error when testing ductile specimens at low rates of loading, but at higher rates the error was too great to be acceptable and it was decided to design a new recording system.

(ii) Electronic Recording System

Load Cells

Three separate load cells were made to suit the load ranges required by the three sizes of specimen. For the double size specimens a cylindrical load cell with three longitudinal and three transverse 1,500 ohm resistance strain gauges was employed, the longitudinal and transverse gauges forming two adjacent arms of a bridge. To hold the striker firmly in place it was necessary to have a pre-load on the load cell and this was arranged by loading through Belleville washers to give a pre-load of 4 ton (see Fig. 14). It was necessary to have this pre-load applied through a spring system because otherwise the direct load on the cell, by compressing the load cell, would relax the pre-load and cause non-linearity. With the spring system in place the small amount of compressive deformation that the cell underwent did not affect the pre-load appreciably.

This simple type of load cell could not be used for the lower loads because of the danger of buckling in very thin

cylinders and ring load cells were designed instead. The dimensions of the load cell for half size specimens together with its clamps are shown in Fig. 15. The cylindrical boss attached to the top clamps could be bolted underneath the moving cross-head of the Denison Testing Machine in place of the normal compression block. Two 1,500 ohm resistance strain gauges were fixed side by side at each of the positions G1, G3, G5 and G7 (i.e. Gauges G1 and G2 at position G1, Gauges G3 and G4 at position G3, etc.). The gauges were connected as shown in Fig. 16 to form a four arm bridge. The load cell for the standard size specimens was of exactly the same type, but had the following dimensions for the ring:- 5.025 in. O.D., 4.149 in. I.D. and 1.43 in. wide. All gauges were coated with "Araldite" both as a waterproofing agent and for protection against accidental damage.

Collection Pick-ups

For the standard size specimens a long duralumin cantilever of 0.500 in. x 0.189 in. cross-section was clamped horizontally on a block fixed to one of the rear tie rods of the testing machine. A spherically capped screw of the type used on dial gauge plungers was screwed into the cantilever at a point 17 in. from the built-in end. The movement of the striker was transmitted to this screw by means of a rigid arm bolted to the striker. Resistance strain gauges were

fixed to the cantilever at a point approximately 1 in. from the built-in end. Initially two 1,500 ohm gauges were used, one on each side of the cantilever, but two more were added later to increase the output. The gauges were used as two adjacent arms of a four arm bridge and the bridge output was found to vary linearly with deflection over the range required (approximately 1 in.). The load produced by the cantilever was small in comparison with the total load taken by the striker and the error caused was less than $\frac{1}{2}\%$.

This was not true, however, when a suitable cantilever was designed to measure the deflection of the half size specimens; in this case a considerable error (7%) would have been introduced had the same system been employed. It was therefore found necessary to develop an alternative system and a carbon strip potentiometer device was constructed. A half inch wide strip of pure carbon on a backing of paxolin was mounted on a vertical steel plate carried by the striker (see Plate 2). An insulated terminal was fitted at the top of the strip and holes were tapped at various pairs of points on either side of the strip so that small terminal clips could be screwed in to connect the strip to the steel backing plate. It was thus possible to apply a potential difference between the top of the strip and any desired point along its length. In the case of the half size specimens, it was found convenient to apply a potential of 6 volts across a $1\frac{1}{2}$ in. length of strip. A rigid pillar was

bolted to the anvil block and carried at the top a short ebonite arm which could be pivoted in a horizontal plane and clamped as desired. A spring-loaded contact was fitted at the end of the ebonite arm so as to wipe along the carbon strip, and to ensure a good contact with the strip a small hemispherical platinum contact was used. The potential difference between the backing plate and the moving contact gave a deflection signal which varied linearly with striker travel. This was a true measure of the movement of the striker relative to the anvils and was unaffected by the deflection of the ring load-cell. It was convenient to apply a backing off voltage to this circuit so that the resultant output from the deflection pick-up was initially negative and reduced to approximately zero at the point when the striker first made contact with the specimen. This is shown in Fig. 17.

This system, which was capable of resolution down to at least 0.002 in., was also used for a few of the later tests on standard size specimens and a very similar system (with the steel backing plate mounted directly on the moving cross-head) was used for all the tests on double size specimens.

Calibration

The load cells were calibrated directly against the Denison Testing Machine. The Southern Instruments bridge

unit employed contained a built-in calibration device which allowed given resistances to be shunted across one arm of the bridge, thus simulating a resistance change in the gauges (see Fig. 16). This calibration was carried out immediately after the direct calibration in the testing machine, allowing the indirect calibration to be related to the true load calibration (e.g. 0.5 megohm calibration resistance = 1 ton). Thus at any stage during the test programme a calibration signal could conveniently be inserted as a check on the battery voltage or the gain value of the amplifiers. A similar bridge unit was used with the cantilever type of deflection pick-up, but for the carbon strip type which did not need a bridge unit, only the supply voltage was checked during the test.

The direct deflection calibrations were carried out using a dial gauge reading to 0.0001 in. carried on a magnetic base. The deflection of the anvil inserts etc. was allowed for by loading on a triple height un-notched specimen of hardened steel and recording the apparent load deflection diagram, the shear and bending deflections of this specimen being calculated and a correction made.

Amplifying and Recording Gear

Matched units made by Southern Instruments Ltd. were used and the connections are shown in the block diagram, Fig. 18. The output from the bridge units (or from the carbon strip

potentiometer when employed) was fed into two d.c. pre-amplifiers, which incorporated built-in drift correcting amplifiers and hum correctors. The latter were able to eliminate any 50 c/s pick-up in the signal. The gain of these amplifiers was only adjustable in three steps and a suitable bridge voltage was chosen to allow as large a part of the total amplification required as possible, to take place in the drift corrected amplifiers rather than in the oscilloscope amplifiers or pen-amplifiers which were not drift corrected. A limit was placed upon this by a maximum output of 20 volts peak to peak from the pre-amplifiers. The oscilloscope used was of the double-gun cathode-ray tube type with independent beam and plate controls, and the load and deflection signals from the pre-amplifiers were fed into the Y and X plates respectively of both guns. The gain on one of the X-amplifiers was made greater than that on the other, and with suitable adjustments of the various gain values two load-deflection traces were obtained, one trace showing the initial part of the diagram to an enlarged scale. The camera used was capable of continuous feed, drum, or single-shot operation, but for all the tests only the single-shot facility was required. The shutter was opened manually just before the striker made contact and was closed when the load dropped to zero. An illuminated graticule assisted in correlating the diagrams with their appropriate calibrations. The load and deflection

signals from the pre-amplifiers were also fed into pen-amplifiers, which had coarse and fine gain controls thus allowing a continuously variable gain, and which gave a sufficiently large output voltage to drive pen-recorders. The twin channel pen-recorder was capable of a frequency response of 100 c/s and paper speeds of 1, 5, 25 and 100 mm./sec. were obtainable by means of a selector switch, operating a mechanical gearbox.

(b) Temperature Control and Measurement

When temperatures above room temperature were required the bath was usually filled with water. It was found necessary, however, to use oil at temperatures near 100°C . because steam from the bath tended to condense on the load cells. Two small electric immersion heaters were built into the bath, but although they were capable of maintaining a given temperature, they took far too long to heat the large volume of liquid in the bath. It was therefore found convenient to heat the liquid to the required temperature before pouring it into the bath. To obtain temperatures below room temperature down to about -75°C . the bath was filled with petroleum-ether and small pieces of solid carbon-dioxide were added to reduce the temperature as desired. The addition of liquid nitrogen at this stage allowed temperatures of -100°C . to be attained, but at lower temperatures a small bath was found to be desirable. A simple

insulated bath was then fitted inside the empty main bath and this smaller bath was filled with high octane petrol which had previously been cooled with solid carbon-dioxide. The addition of liquid nitrogen to this small bath enabled temperatures of nearly -130°C . to be attained before the petrol froze. For the isolated tests at -196°C . side pieces were fixed on to the anvil block and sealed with "Hermatite" to form a small tank. The continuous addition of liquid nitrogen enabled this temperature to be maintained during a test.

To measure the specimen temperature a copper-constantan thermocouple was fixed to the specimen with "Cellotape" and the thermocouple voltage measured with a thermocouple potentiometer. A calibration chart was available for the thermocouple employed, but this was checked and a small correction had to be made to the curve at temperatures below -100°C . The only temperatures easily checked in the low temperature region were those of solid carbon-dioxide and of liquid nitrogen, the former agreed exactly with the calibration curve whilst the latter was approximately 0.1 mv. ($\pm 6^{\circ}\text{C}$.) in error. The lowest temperature for which the calibration curve was required, however, was -130°C . and at this temperature the correction was only about $1\frac{1}{2}^{\circ}\text{C}$., thus any error was likely to be small. In general temperatures could be measured to an accuracy of $\pm \frac{1}{2}^{\circ}\text{C}$.

4.2. Testing Procedure

The initial dimensions of all the specimens to be tested were carefully measured using a ball-ended micrometer, the critical dimensions being the width of the specimen at the root of the notch and at the point of application of load. Factors such as the notch depth had previously been checked on each batch of specimens after machining. The alignment of the bending jig was then checked by putting a light coating of grease on top of a specimen, locating it in the jig and carefully lowering the striker until it just made contact. The position of the line made in the grease was then checked with a travelling microscope and any required adjustment was made. This method also allowed any uneven distribution of load to be detected.

The appropriate initial calibrations as specified in Section 4.1. were then performed, after which the fine by-pass valve was closed and the desired rate of pumping set on the machine. The appropriate pumping rates had been determined by preliminary investigations using rough specimens. The particular loading rates chosen were of no special significance and were merely governed by the maximum and minimum pumping rates which were available.

The bath could now be brought to the desired temperature and a specimen, with a thermocouple attached, placed in position on the anvil and allowed to soak at that temperature

for a few minutes whenever possible. A few other specimens were placed in the bath at the same time in order to save time in bringing them to their testing temperature when the time came. The coarse by-pass control was operated to bring the striker to within a few inches of the specimen and when conditions were steady the valve was closed to start the test. When the electronic recording gear was used, the pen recorder was started and the oscilloscope camera shutter opened manually just before the striker made contact. The determination of the right moment for this operation was facilitated by the cross-head movement indicator possessed by the testing machine. The pen recorder was stopped and the shutter closed as soon as the load had dropped to zero. The specimen was then removed from the bath and carefully dried, examined and measured.

In this way diagrams of load-deflection - and, whenever electronic recording was used, of load-time - were obtained over a range of temperatures for the three different sizes of specimen of both steels. Additional tests were performed at the fastest available loading rate on standard size specimens of both steels. These will be called "fast notch-bend tests". The total temperature range employed was -196°C . to $+66^{\circ}\text{C}$. The records obtained from the oscilloscope trace were all enlarged by a factor of exactly five in a Hilger Measuring Projector and a tracing made of the enlargement.

Measurements were made on these tracings and on the autographic records to obtain the loads and deflections at various stages of the test. Measurement of areas with a planimeter allowed the corresponding indicated energies absorbed to be calculated.

4.3. Results and Discussion

The measured rates of loading, together with the suggested rates of elastic stressing at the notch root in the different notch-bend tests, are given in Appendix 'D'. The experimentally determined values of loads, deflections and energies are listed in Appendix 'A' together with the measured values of % crystallinity, % lateral expansion etc. Tables 3 and 4 contain the results obtained on standard size slow notch-bend specimens of steels 'P' and 'V' respectively, whilst Tables 5 and 6 contain the corresponding fast notch-bend results. The results obtained on half and double size specimens are given in Tables 9 and 10, and 11 and 12 respectively.

The variation of the type of autographic record obtained over the transition range for standard size slow notch-bend tests is shown in Fig. 24, for steel 'P'. The horizontal trace, which can be seen on some records at the lower temperatures, is the superimposed time-marker. The testing temperature (in °C.) is shown on each reproduction of the diagram and the corresponding observed % crystallinity of the

fracture surface is noted beneath it. The diagrams obtained on half size specimens of steel 'P' are shown in Fig. 25 (deflection increasing from right to left) superimposed upon the corresponding transition curve. Two traces can be seen in each diagram; these are both of load-deflection but one is to a magnified scale. The scale used for the standard trace is the same in all the diagrams, but the scale used for the magnified trace is enlarged by a factor of 1.65 in the diagrams at -82.4°C. , -108.3°C. and -196°C. Similar diagrams, again superimposed on the transition curve, are shown in Fig. 26 for double size specimens of steel 'V'. The scale of both traces is enlarged by a factor of about 5 in the diagrams at -96°C. and -123°C.

It will be noticed that, in the temperature range where fractures occur after the maximum load, the value of this maximum load increased with decreasing temperature, the increase in maximum load being about 15% to 20% for 100°C. drop in temperature. This was, however, in general less than the corresponding increase in the yield load which was of the order of 50% (see Section 8.2).

The usual % crystallinity and energy transition curves are shown in Figs. 41 - 48 and these will be discussed together with the results of the other tests, with regard to the size and strain rate effects, in Section 8.

It may be seen from the Tables that the % lateral

contraction has only been measured for the slow notch-bend tests on steel 'P'. The contraction measurements were made with a pointed micrometer and the degree of accuracy obtainable was not high, especially when the two broken halves had to be fitted together to obtain a reading. It has been shown by Hartbower (68), however, that the maximum lateral contraction (which is not always in the vicinity of the notch root) is directly proportional to the lateral expansion. He verified this on many specimens involving a variety of metallurgical variables. The results obtained on steel 'P' exhibit considerable scatter but do show the same trend (see Fig. 27). In view of this and of the practical difficulties involved in an accurate measurement of contraction, it was decided to measure only the expansion in the other tests.

SECTION 5 - INSTRUMENTED IMPACT NOTCH-BEND TESTS

5.1. Apparatus

The machine available for instrumentation was an old French machine with anvils and striker nose radius corresponding to B.S.131: 1933. The friction losses in this machine were found to be rather high, but the machine had the advantage of possessing a variable pendulum release position so that the velocity and kinetic energy at impact could be varied. The pendulum release catch was mounted on a sector of a toothed wheel driven by a worm, and this also allowed the pendulum to be wound down slowly by hand during alignment and calibration.

The actual striker was a hardened steel insert 5 in. overall diameter, which was a force fit in the mild steel body of the pendulum (see Plate 3) and to remove this insert or to adjust its alignment it was necessary to enlarge slightly the opening in the pendulum body by forcing apart the two "jaws" of the pendulum. This was done by machining a small flat on each of the pendulum jaws and inserting a small jack. Once the insert had been correctly set, the jack was removed and the insert appeared completely rigidly fixed, but as an added precaution two clamping bolts were provided.

Load Measurement

To measure the loads in the striker, four platinum-iridium resistance strain gauges, having a gauge factor of 6, were

fixed to the striker. The two active gauges were fixed longitudinally on the tapered portion and the two dummy gauges transversely on the full width portion, as shown in Fig. 19. All the gauges were covered with a protective coating of Araldite. To calibrate the system the pendulum was removed from the impact machine and a static calibration was performed in the Denison testing machine. Calibration resistors were available in the bridge unit, as before, to allow spot calibrations to be made in situ. In applying the calibration load a standard Charpy specimen was placed between the compression block of the Denison testing machine and the striker nose, so as to reproduce as closely as possible the load distribution encountered in the actual test. In fact the active gauges were so positioned that, according to Cotterell (69), the load distribution should have had little effect. A few tests with uneven load distributions verified this. It will be shown later that the dummy gauges were scarcely affected by the applied load and served only for temperature compensation. Because the gauges were of finite length ($\frac{1}{2}$ in.) it was not easy to detect individual stress waves, although an attempt will be made to explain some of the recorded oscillations on this basis. The arrangement, however, proved quite satisfactory for the main purpose of the test namely, the determination of the load-time and load-deflection diagrams.

Deflection Measurement

To measure the deflection, two carbon strip potentiometer devices were used and one of these can be seen on the pendulum in Plate 3. These were similar to the one used for the slow bend deflection measurements, a 4 volt accumulator being used to apply a potential difference across a 2 in. length of the strips. Steel pillars were bolted on each of the hardened steel anvils, as shown in Plate 4, and the spring-loaded contacts were carried by adjustable ebonite blocks mounted on top of the pillars. These contacts were initially positioned by stops on the pull-rods, but as the pendulum swung down and hit the contacts the spring force on them was transferred to the pendulum. Further pendulum movement rubbed the strips along the contacts. Strips of "Sellotape" covered that part of the contact path on the pendulum which did not include the carbon strips. This was not for insulation purposes, but merely to prevent wear on the platinum tips of the contacts. The potential difference between the front point of voltage application and the contact gave the deflection signal. It was convenient to employ an adjustable backing-off voltage as shown in Fig. 20. This applied a negative voltage to the amplifier before the active part of the strip reached the sliding contact and allowed the amplifier input to be made approximately zero at the point of impact, thus making the best use of the drift-corrected amplifier

without saturating it.

By slowly winding the pendulum down a direct deflection calibration was made against a $\frac{1}{2}$ in. travel dial gauge reading to 0.0001 in. To avoid having to touch the Y-shift control of the oscilloscope after the load calibration had been made, it was necessary to provide an extra variable input voltage which could temporarily be plugged into the oscilloscope Y-input in place of the output from the load amplifier. This allowed the spot to be traversed vertically across the screen at each step of the deflection calibration.

It was arranged that, after pushing the specimen through the anvils, the pendulum snapped a pencil lead fixed in its path, thus breaking a relay circuit and energising two solenoids. These solenoids were arranged to pull back the contacts so that they were not struck by the pendulum on its return swing (see Plate 4).

Time-Base

The oscilloscope time-base was of a type in which the spot movement was triggered by breaking a pair of contacts. Another pencil lead was wired in the time-base trigger circuit and arranged on the impact machine so that it was broken by the pendulum just before impact. By suitable adjustment of the lead position and of the oscilloscope X-shift control, it was possible to arrange for the spot just to have arrived on the screen at the moment of impact. To calibrate the time-base a

sine wave signal from an oscillator, which was accurate to within $\pm 2\%$, was fed into the Y-plates and the time-base triggered by hand. A signal of 2,000 c/s was normally found to be convenient.

The amplifying and recording gear is shown schematically in Fig. 20. All records were taken using the same Kodak R.P.30 recording paper that was used for the slow bend tests, but it is suggested that some faster film such as R. 60 would probably be more suitable.

Calibration Routine

Although the basic calibrations required were in themselves perfectly straightforward it was very easy to forget to alter a particular instrument setting or wiring connection when changing from one calibration to another. With this in mind a calibration routine was devised and it was found possible to use only those controls with certain pre-set positions (such as the gain controls on the drift-corrected amplifiers). Thus this routine could conveniently be employed at any stage of the testing programme without the risk of altering any of the calibration values. The calibration routine is described in Appendix 'F', Part (1).

Temperature Control

A special locating tool was made to transfer the specimen from the cooling bath to the machine and this ensured that the specimen was positioned both quickly and accurately. Both the

specimen and the locating tool were immersed in the bath, and the temperature change in the specimen in the short time interval between leaving the bath and the moment of impact was thus minimised. Tests were carried out with a thermocouple attached to the specimen and the temperature change in a given time was recorded; by this means the actual temperature at impact could be determined. This temperature correction was not large, there being about a 2°C . rise in temperature before impact when a specimen was transferred to the testing machine from a bath at -75°C .

5.2. Testing Procedure

The initial dimensions of all the specimens to be tested were measured in the usual manner, and the alignment of the testing machine was checked to ensure that a specimen which was accurately located on the anvils would be struck exactly opposite the notch. The calibration routine was then performed.

The specimen could then be placed on the locating tool and both immersed in the heating or cooling bath. Before the pendulum was raised to its release position, the power to the solenoids was switched on and it was checked that the contacts were pulled back, because it was possible to damage the contacts if the pendulum was moved in the reverse direction while they were in place. Once the pendulum was raised to its maximum position and held by the release catch the pencil leads could

be inserted, cutting the power to the solenoids and moving the time scale spot off the screen. When the required temperature had been reached the brightness of the two spots was increased to the maximum value with the camera shutter still closed, and using the locating tool the specimen was rapidly transferred from the bath into the machine and the pendulum immediately released. As the pendulum descended the camera shutter was opened manually and then closed again directly after impact. The machine reading was recorded before the pendulum was stopped with the brake. The specimen was then retrieved, dried, examined and measured.

A series of specimens of both steels were tested in this way over a range of temperatures to cover the transition range of the steels. Simultaneous diagrams of load-time and of load-deflection were obtained and the integrated area of the load-deflection diagrams were used to give a measure of energy absorbed in the tests.

5.3. Results and Discussion of Results

It has not been possible to present direct reproductions of the actual diagrams obtained because these were rather weakly defined, but reproductions (to a reduced scale) of some of the tracings made on the Hilger Measuring Projector are shown in Figs. 28 and 29. Although load-time diagrams were obtained in every case they were not usually employed to determine the indicated energy absorbed and were therefore not

always traced. For this reason they are missing on some of the reproductions. No scale is given because the scale employed was not the same for all tests and calculations were made on individual batches. The energies obtained from these diagrams together with other relevant data, such as the indicated machine energy, % crystallinity, etc., are given in Appendix 'A' Tables 7 and 8. The general crystallinity and energy transition curves are shown in Figs. 45 to 48 and these will be discussed in relation to the other tests in Section 8.

Increasing interest has recently been paid to the interpretation of Charpy impact test results and in the past few years a few results (69 to 74) have been published of loading records taken to supplement the normal energy or crystallinity measurements. These results have, however, all been essentially load-time diagrams obtained by measuring the load on either the striker (69, 70, 72 and 73) or the anvils (71 and 74). To obtain energy values from these records it was necessary to convert the load-time record into one of load-deflection. It has been usual (69) to make the assumption of a pendulum velocity decreasing uniformly from the known impact to the observed exit value. The simultaneous records of load-time and load-deflection presented by the author allow a check to be made on the assumptions made with regard to the variation of pendulum velocity.

(1) A Comparison of Load-Time and Load-Deflection Diagrams

The diagrams obtained on a ductile specimen of steel 'V' were chosen for this investigation (testing temperature 0°C.). The load-time diagram was divided into ten equal time zones, the velocity being assumed constant in any one zone. From a knowledge of the initial and final angles of the pendulum and of the friction and windage losses involved, it was possible to calculate the impact and exit velocities. The velocity at the mid-point of each time zone was then calculated on the assumption of a linearly decreasing velocity and hence for a given zone the time scale could be converted into one of deflection. The initial portions of the two diagrams were not well defined but were seen to be of the same form; a point of coincidence was therefore taken just after the "upper yield drop" and the load-time diagram was replotted on a deflection basis from this origin (see Fig. 30). The error produced by assuming an arbitrary origin at this point is not great because the initial portion of the diagram accounts for less than 10% of the total energy indicated by the diagram, and thus any error in the value of this portion will not greatly affect the total indicated energy absorbed.

It is seen that the area of the load-time diagram, replotted on this basis, is larger than that of the load-deflection diagram. A check was made and, by taking pairs of corresponding points on the two diagrams, the average striker velocity over the region of the replot was calculated and

found to be of the same order as the exit velocity (4.50m./sec.). It was decided to assume some form of shock loss of energy at impact and to replot the load-time diagram on a deflection base, this time assuming a constant velocity at the exit value. This curve is shown dashed in Fig. 30. A similar replot was performed on the diagram for a more brittle specimen and is shown in Fig. 31. Both results show that the assumption of an instantaneous loss of velocity followed by a constant velocity push through gives a degree of coincidence with the load-deflection record that is superior to the assumption of linearly decreasing velocity. As a direct check on this assumption it would have been very desirable to have obtained a deflection - time diagram from the instrumented machine, but this only became fully apparent at a later stage when the results were being analysed and it was not practicable to re-assemble the equipment at this stage.

The only displacement - time diagrams for the impact notch-bend test of which the author is aware, are those obtained by Tanaka and Umekawa (70), who present diagrams of this type corresponding to energy absorptions of 0, 9.5 and 20 kg.m. on a 30 kg.m. machine. No deflection scale is given with these diagrams and their interpretation is far from clear. If, however, enlargements of the curves corresponding to 0 and 9.5 kg.m. energy absorptions are superimposed as in Fig. 32, then it might be suggested that an abrupt change of velocity is revealed.

(2) A Comparison of Energies from the Machine Reading and from the two Diagrams.

Fig. 34 shows that the area under the load-deflection curve represented about 67% of the energy absorbed as indicated by the testing machine (after subtraction of friction and windage losses). This was true at all values of crystallinity and the same percentage was obtained on both steels. This factor is in fair agreement with Cotterell (69) who used the assumption of linearly decreasing velocity to obtain energy values from his load-time curves. Augland (74), however, used a machine of large capacity and, assuming a constant velocity, obtained good agreement between the energy obtained from the load-time curves and the machine energy. Good agreement was also obtained by Fearnough (73) who used a correction suggested by Augland to convert load-time curves obtained on a standard capacity machine. This correction was as follows:-

$$W_A = W_t \left(1 - \frac{W_t}{4W_0} \right) \quad (5.1)$$

where W_A was the corrected energy, W_t the energy derived from the load-time curve assuming a constant velocity (Presumably the mean velocity), and W_0 was the initial energy of the striker. It would appear to the author that the above correction does involve some approximations, which could introduce errors when large values of energy absorption are obtained on a standard capacity machine.

It will be shown later in this section that certain dynamic effects can introduce errors in measurements obtained with ordinary resistance strain gauges. Such errors cannot, however, account for the differences between the results obtained by the various investigators, since most of the other investigators have also used these gauges either directly for load measurement, or indirectly for the dynamic calibration of the load cell.

In view of the different relationships presented between the values of energy absorbed, as obtained by the various methods, it was decided to investigate more closely this aspect of the test. Table 20 shows the various values of energy absorption obtained on the specimen of steel 'V', the diagrams for which are shown in Fig. 30. (For Table see following page.)

It would appear from Table 20 and from Fig. 30 that, after the initial period of vibration, it is a fair approximation to assume that the velocity remains constant at about the exit value. This can, of course, only be an approximation because it is not possible for the pendulum to do work without losing kinetic energy and hence velocity. It would, however, indicate that a shock loss of energy takes place at impact and the results will be examined on this basis.

TABLE 20

ENERGY	kg.m.
Energy absorbed as indicated by the testing machine after subtraction of friction and windage losses.	$W_m = 9.79$
Energy absorbed obtained from load-deflection diagram.	$W_d = 6.55$
<u>Energies absorbed obtained from load-time diagram</u> (a) <u>Using the measured area of the diagram and assuming:</u> (i) Velocity constant at the mean velocity (5.0 m./sec.). (Using Augland's correction (ii) Velocity constant at the exit velocity (4.50 m./sec.) (b) <u>Replotting the load-time diagram to a base of deflection on the assumption of:</u> (i) Velocity decreasing uniformly with time (diagram divided into 10 time zones) (ii) Velocity constant at the exit velocity.	$W_t = 7.21$ $W_A = 6.76$ $W_e = 6.50$ $W_f = 7.25$ $W_g = 6.54$

Let us assume that the velocity - time curve is of the form shown in Fig. 33, i.e. impact at '1' followed by a shock loss to 'a' and then a linear decrease in velocity to the exit point '2'. Let W_1 , W_a , W_2 be the energies possessed by the pendulum at the points '1', 'a' and '2' respectively.

Dashed velocities will be used to refer to the centre of gravity of the pendulum whilst undashed velocities will refer to the striker edge, which is opposite the centre of percussion. The two sets of velocities are related by the geometry of the pendulum so that $V = hV'$. (For the machine in question $h = 1.085$.)

The machine energy, $W_m = W_1 - W_2$

It will be assumed that the various losses which occur in the machine from many causes can be represented by two factors, firstly the shock loss $W_s = (W_1 - W_a)$ and secondly losses during the progress of the test from 'a' to '2', represented by K^* . Thus if W^* is the true work done on the specimen in this region then $W^* = (W_a - W_2) - K^*$

Consider the linear portion of the curve from 'a' to '2'.

$$dW = Pdx = P \frac{dx}{dt} dt = PVdt$$

$$dW_t = V_m Pdt \text{ (assuming a constant mean velocity, } V_m)$$

$$\text{Therefore } \frac{dW_t}{dW} = \frac{V_m}{V} \text{ at any stage} \quad \text{-----} \quad (5.2)$$

It is not known how the secondary loss K varies from 0 to K^* in the period from 'a' to '2', because the exact causes of such a loss are not known. As an approximation it would

seem reasonable to assume a linear variation with velocity so that at any point in this region where the striker velocity is V then

$$K = \left(\frac{V_a - V}{V_a - V_2} \right) K^* \quad (5.3)$$

(The alternative assumption, K varying with $(\text{velocity})^2$, has been found to have a very small effect on the following calculations.)

Thus the true work done on the specimen from 'a' to this point is

$$W = \frac{1}{2} \frac{m}{h^2} (V_a^2 - V^2) - \left(\frac{V_a - V}{V_a - V_2} \right) K^* \quad (5.4)$$

$$\text{Now } W_a = \frac{1}{2} m (V_a')^2 = \frac{1}{2} m \frac{V_a^2}{h^2} \quad (5.5)$$

where m is the mass of the pendulum

Dividing equation (5.4) by equation (5.5) gives:-

$$\frac{W}{W_a} = 1 - \frac{V^2}{V_a^2} - \frac{(V_a - V)}{(V_a - V_2)} \frac{2h^2}{mV_a^2} K^*$$

$$\text{Therefore } dW = W_a \left(-\frac{2V}{V_a^2} + \frac{2h^2 K^*}{mV_a^2 (V_a - V_2)} \right) dV$$

$$\text{or } dW = W_a \left(-\frac{2V}{V_a^2} + SK^* \right) dV \quad (5.6)$$

$$\text{where } S = \frac{2h^2}{mV_a^2 (V_a - V_2)}$$

Hence from equation (5.2)

$$dW_t = W_a \frac{V}{V} \left(-\frac{2V}{V_a^2} + SK^* \right)$$

$$\begin{aligned}
 \text{Therefore } (W_t)_{a \rightarrow 2} &= W_a V_m \int_{V_a}^{V_2} \left(\frac{-2}{V_a^2} + \frac{SK^*}{V} \right) dV \\
 &= W_a V_m \left[\frac{-2(V_2 - V_a)}{V_a^2} + SK^* \log. \frac{V_2}{V_a} \right] \\
 &= W_a V_m \left[\frac{2(V_a - V_2)}{V_a^2} - SK^* \log. \frac{V_a}{V_2} \right]
 \end{aligned}$$

Substituting for W_a from equation (5.5) and for S gives:

$$(W_t)_{a \rightarrow 2} = \frac{mV_m}{h^2} \left[(V_a - V_2) - \frac{h^2 K^*}{m(V_a - V_2)} \log. \frac{V_a}{V_2} \right] \text{-----} (5.7)$$

Now V_m is a quite arbitrary diagram conversion factor and is not related to the mean velocity in the region 'a' to '2'.

On the assumption that $V_m = 5.0$ m./sec. we have a value of 7.21 kg.m. for W_t (see Table 20). This does, however, include the indicated energy absorbed during the period of severe vibrations. The numerical value of this energy was small and W_t may be taken as approximately 7.1 kg.m. for the region 'a' to '2'. Substituting these values in equation (5.7) together with $m = 22.45$ kg., $h = 1.085$ and $V_2 = 4.50$ m./sec. gives:-

$$0.73 = \left[(V_a - 4.50) - 0.0524 \frac{K^*}{V_a - 4.50} \log. \frac{V_a}{4.50} \right]$$

where K^* is in kg.m.²/sec.².

It is not possible with the available data to obtain an explicit solution to the above equation, but it is possible to

define the limits within which W_s and K^* can vary.

The first possibility is that the losses are entirely from shock causes and that K^* is zero. In this case $V_a = 5.23$ m./sec. and $W_a = 26.60$ kg.m. This makes the shock loss, $W_s = 29.45 - 26.60 = 2.85$ kg.m. and makes the true work done on the specimen, $W = 6.94$ kg.m. which is starting to compare with the value of 6.55 kg.m. obtained from the load-deflection diagram. In other words, if this were true, there is a shock loss of 2.85 kg.m. at impact which would be dissipated in vibration of the pendulum etc. followed by a straightforward test with the velocity decreasing linearly from 5.23 to 4.50 m./sec. (i.e. a mean velocity of 4.86 m./sec.), with no other losses occurring.

The other extreme is the assumption that W_s is zero and that all the losses are contained in the K^* term. This means that $V_a = V_1$ and $W^* = (W_1 - W_2) - K^* = 9.79 - K^*$. Substituting these values in equation (5.7) and remembering that W_t must now be taken as 7.21 to include the initial region gives

$$K^* = 24.8 \text{ kg.m.}^2/\text{sec.}^2 = 2.52 \text{ kg. (force) metres}$$

Therefore $W^* = 9.79 - 2.52 = 7.27$ kg.m.

(c.f. the value of $W_f = 7.25$ kg.m. obtained by replotting the load-time diagram to a deflection base on the assumption of a linearly varying velocity.)

If the comparison of the load-time and load-deflection diagrams in section 5.3. (1) is correct, then the former

assumption would seem more nearly correct and it would seem likely that K^* is small. Similarly, any attempt to explain the difference between the machine energy and the indicated energies from the diagrams from the basis of dynamic errors in the strain gauges giving a load signal which was too small, encounters the same difficulty. If the load signal were too low then the indicated energies from the diagrams were too low, and hence W_s should be smaller and V_a larger. This would mean that a higher mean velocity would be required as a basis for the replot and this is not in agreement with the velocity required for correlation of the two diagrams.

It is suggested that the best basis for the transformation of diagrams of load-time into those of load-deflection is the assumption of a constant velocity at the exit value. This is, however, presented as a working approximation and not as a true representation of the physical behaviour. It is also possible that the magnitude and distribution of losses may well be different in another type of machine.

(2) Records Obtained with Gauges on Only one Side of the Striker

A few tests were made using only the gauges on one side of the striker. A typical record of this type is shown in Fig.35 and it is seen that the resultant diagram is very different from the usual diagram produced by all four gauges; in fact the recorded stress in the striker goes well into the

tensile region at some stages. This effect has also been noticed by Fearnough (73), who has obtained simultaneous records of the load-time relationship in each of two gauges, one on either side of the striker. The cancellation of this effect when the two gauges are placed in opposite arms of a bridge or wired in series, suggests that there is a strong bending stress superimposed on the direct stress system.

Cotterell, however, used only one gauge yet did not obtain this effect. This surprising difference may of course be attributed to the type of machine employed and, although the pendulum-head of the author's machine was of a type which resembled the Losenhausen machine of Cotterell far more than it did the Avery of Fearnough, it was found that a rather high degree of torsional vibration was possible in the shaft. In this respect it was similar to the Avery machine which is somewhat notorious for its lack of rigidity. The final portion of these diagrams appears to be a series of oscillations of decreasing frequency (2.4 to 1.7 kc./sec.) superimposed on a decay curve. It is suggested that some form of torsional vibration was set up in the pendulum shaft causing bending loads to be applied to the striker nose.

(3) Comments on the Stress Waves and Oscillations set up in the Specimen and in the Recording System

The stress waves and other oscillations obtained in the records may now be examined, but at this stage it should be

noted that the records obtained were designed to give an overall picture of the loads during the test and not specifically to examine the stress wave system at impact. Thus the times involved in the initial period of fluctuations could not be measured with great accuracy. The records were enlarged on a Hilger Measuring Projector and were traced if required. Time measurements were made to an apparent accuracy of 10^{-5} sec. The oscilloscope time base, however, was only accurate to within $\pm 10\%$ and the calibration trace to within $\pm 2\%$. Also the X and Y Plates of the oscilloscope were assumed to be mutually perpendicular and any lack of alignment could cause a quite considerable error in the small time values measured for the initial stages. Thus it is doubtful whether the time values could be measured to an accuracy greater than about 1.3×10^{-5} sec.

To simplify the analysis the record obtained using gauges on only one side of the striker (Fig. 35) will be employed. Although dummy gauges were included in the circuit for temperature compensation, they were fixed transversely at the full thickness section (see Fig. 19). Thus the magnitude of the strain produced in the dummy gauges, whether by a steady stress or transient stress wave, was very much less than that at the active gauge. Also, unless there were any extra transverse stresses produced during the test, only the cross-sensitivity of

the dummy gauges need be considered. It therefore seems safe to conclude that the dummy gauges had little effect on the bridge output, except to compensate for any temperature changes. Thus we may consider Fig. 35 to be the record of the stress history at a single active gauge.

(a) Initial Stress Wave Effects

First consider a compressive stress wave which is produced in the striker by the initial impact; this wave travels along the striker until it reaches the interface A-A between the hardened insert and the main body of the pendulum. The insert is a tight fit in the body but it is possible that some kind of air gap could exist. It will be assumed that partial reflection of the wave occurs. Thus a compressive wave of reduced magnitude continues towards the free face B-B (Wave 1) and a tensile wave (Wave 2) of an unknown magnitude is reflected towards the nose. The calculated time taken between the initial compressive wave passing the active gauge and the reflected tensile wave returning to the gauge again is approximately 2×10^{-5} sec. This agrees with the measured time of 2.3×10^{-5} sec. between 'a' and 'b' on the load-time record (Fig. 35). The reflected tensile wave, 2, reaches the nose of the striker and, depending on the contact with the specimen is partially or completely reflected as a compression wave. There is sufficient time for this wave to travel to A-A, return to the nose, and once again travel to A-A before the initial

part of the compressive wave has returned to A-A after reflection as a tensile wave from B-B. In this time interval wave 2 will have superimposed fluctuating stress waves of decreasing magnitude and increasing complexity upon the compressive stress wave. No such fluctuations are in fact observed.

If the interface were in fact perfectly continuous then the initial compressive wave would continue unaltered to B-B where it would be reflected as a tensile wave of the same magnitude, thus the stress at the gauge should drop to zero about 7.6×10^{-5} secs. after peak 'a'. This is not indicated by the diagram.

Another effect which must be noted is that due to the shape of the striker nose. If a compressive wave is propagated along the axis of a conical specimen then, because of the change in section, it is followed by a wave of tension (75). In a similar fashion a compressive wave approaching the striker nose may be expected to develop a "tension tail". This phenomenon adds considerably to the difficulty of interpreting the recorded waves.

It would seem pertinent to mention at this stage three points which have been raised by Bloss (76) with regard to the use of resistance strain gauges under dynamic conditions.

- (1) The gauge is of a finite length and should the wave length of the stress wave be shorter than this, then a falsely low

reading will be obtained. (2) Installation techniques, especially the cement thickness, may affect the gauge response appreciably. (3) Some types of gauge generate an e.m.f. which is proportional to strain rate and this voltage may introduce a significant error under impact loading conditions.

Some investigators have measured the loads in the specimen supports and not in the striker. In one case (71) initial fluctuations are present, but the load drop is not so severe, whilst in the other (74) quite severe fluctuations are recorded, but there does not seem to be any decrease before maximum load although the record suggests that the spot may have hesitated during the initial increase. These differences between recording methods only serve to reinforce the fact that the recorded waves are those in the load cell in question and not those in the specimen.

(b) Oscillations at Later Stages in the Diagram

With regard to the oscillations immediately following the maximum load, Cotterell considers that they are entirely mechanical and that they are gradually damped out if appreciable yielding takes place. The author's records confirm this and it is interesting to note that the oscillations are more pronounced on the standard diagrams than on those with gauges on only side. The measured frequency of these oscillations is 15 kc./sec. which is

exactly the value which Robertson (77) quotes as the natural frequency of an unbroken specimen, and also agrees with Cotterell's experimentally determined value.

Some records were made to a reduced time scale and the vibrations remaining after the load had dropped to zero were clearly seen (Fig. 36). These corresponded to about 1.25 kc./sec. and are probably connected with the natural frequency of some part of the striker, because they are present both after fractures which are ductile and also those which are largely brittle in nature.

(c) Transient Stresses Within the Specimen

It should be noted at this point that the maximum force between striker and specimen occurs at the moment of first contact or soon after (78). This does not, however, mean that the bending moment or stress is necessarily a maximum at this moment.

In the neighbourhood of the impact point the transient stress distribution is three dimensional; the waves are not only propagated along the beam but are also reflected from all surfaces. The presence of the notch further complicates the wave distribution. At greater distances from the impact point the stress will be more uniformly distributed across the width and the normal variation of bending stress in the depth direction will be exhibited. This pattern will extend to the contact region if the load is removed. Experiments have been

performed by dropping steel balls on to beams and measuring the contact force and outer fibre strain at various points with respect to time (79). The maximum central bending strain has been shown to increase linearly with velocity of loading. This agrees with the observation of Cotterell that the initial pulse due to the first impact is proportional to the impact velocity. An important fact which has often been neglected is that there are also shear waves present in the specimen which travel at $\sqrt{\frac{G}{E}}$ x velocity of direct stress waves. Thus, as is shown in Fig. 37, the specimen is effectively shorter than its true length until the shear wave has travelled to the supports. This means that although the central force may reach its maximum value the bending moment may not do so because of the reduced effective moment arm. These shear waves will be reflected in their turn, but it is a matter of conjecture how great a reflection will occur at the supports and how much at the free end of the specimen. It may be concluded then that the stress system in the specimen is extremely complex at the moment of impact and soon after, and that any attempt to estimate the stresses from a load cell on the striker or supports is likely to be very misleading.

(4) The Variation of Diagram Shape Over the Transition Range

With reference to Figs. 28 and 29 it may be seen that as in the slow notch-bend tests the sudden drop in load, which is

taken to correspond with the passage of a brittle crack through the test piece, occurs at successively earlier stages in the test as the temperature decreases. At the lower temperatures fracture is seen to occur on what would appear to be the rising elastic line, although not before the indicated drop in load through stress wave effects (e.g. V247 at -73°C . in Fig.29b).

If we compare P94 ($+48^{\circ}\text{C}$. 70% crystalline) in Fig. 28 with V244 (-16°C . 40 % crystalline) in Fig. 29, it can be seen how a given energy level (in this case approximately 4.15 kg.m. indicated energy absorbed from the load-deflection diagrams) can be obtained from different types of record even for two classes of mild steel.

SECTION 6 - NOTCH TENSILE TESTS

6.1. Apparatus

The 10 ton Denison Testing Machine was also used for these tests and the specimen assembly, including the spherically seated shackles, the load cell, and the temperature bath is shown in Fig. 21. The spherical seats from a pre-existing piece of apparatus were employed in an attempt to obtain axial loading, but as will be seen later from the results of the static yield stress tests, true axiality is very difficult to obtain. The temperature tank was made in two sections, each capable of being drained independently. The centre section could thus be drained temporarily to remove a fractured specimen and insert a new one without the tank unit as a whole losing all its heat, thus saving some time in reaching the next required temperature. For the higher temperature tests electric heating elements were clamped around the tank and the double tank arrangement had the added effect of giving a more even temperature distribution in the centre section. Low temperatures were obtained using petroleum-ether, solid carbon dioxide, and liquid nitrogen as for the slow notch-bend tests. Two thermocouples were attached to the specimen at either end of the parallel portion to detect any temperature gradient, and the test was only performed when uniform conditions had been achieved.

The mechanical recorder was again used for these tests

but as is seen in Fig. 21, a load cell was included in series with the shackles, and this was calibrated directly against the testing machine. The same amplifying and recording gear was used to give records of true load against time and the rate of cross-head movement was adjusted to give the required rate of loading. As before, a time-marker was fixed to the mechanical recorder, thus allowing the response of the mechanical recorder to be checked. No appreciable error was found to occur up to the point of fracture, and since for both ductile and brittle fractures this occurred in an unstable fashion an instantaneous reduction of load to zero at this point could be assumed.

6.2. Testing Procedure

The general testing procedure followed was very similar to that in the slow notch-bend tests. Some preliminary tests were made to determine a rate of pumping which would produce the same nominal rate of elastic stressing at the root of the notch as in the slow notch-bend tests. These tests of course required the recording of load-time diagrams to obtain a measure of the loading rate, but for the routine tests only the mechanical recorder was required.

6.3. Results and Discussion of Results

The energies, loads and overall extensions obtained from the autographic load-extension records are listed in

Appendix 'A', Tables 13 and 14, together with the corresponding % crystallinity values. The % crystallinity and energy transition curves are shown in Figs. 49 and 50 respectively and these will be discussed, together with the notch-bend results in Section 8.

Reproductions of the autographic records obtained on steel 'P' are shown, to a reduced scale, in Fig. 38. It can be seen from these diagrams that a more obvious yield is indicated than in the notch-bend tests which also employed this recorder. This yield usually took place in two stages, designated "yield 1" and "yield 2". These do not correspond to the upper and lower yield stresses as indicated on a normal unnotched tensile load-extension curve. "Yield 1" was the first point of obvious non-linearity (which was usually a quite distinct point), whilst "yield 2" corresponds to a short plateau, which was sometimes followed by a slight drop in load.

The most outstanding feature of these records is the instability exhibited soon after maximum load, irrespective of whether the fracture was fibrous or crystalline. A similar instability was observed in the records obtained on steel 'V'. It was presumed that this was a machine stiffness phenomenon, and it was verified that the instability point occurred at a stage when the slope of the diagram, as the load fell away, approached that of the machine stiffness line. The fracture

which took place at this point was very fast, and no appreciable difference was observable between the load-time records taken for fractures which were fibrous and those which were crystalline in appearance. In other ways the diagrams were straightforward, the yield load increasing as the temperature was decreased. The maximum load also increased with decreasing temperature until fracture took place in the region of increasing load. Thus a result of the instability is that the indicated energy absorbed to fracture may be greater in a specimen exhibiting a largely crystalline surface appearance than in one with an entirely fibrous surface (compare specimens P2 and P7). This does not, of course, mean that the true work done at the fracture surface was greater. It merely means that the strain energy available in the system at the instability point was more than sufficient to do this work and any surplus was dissipated in other forms. The amount of the available strain energy used in fracturing the specimen relative to the amount dissipated would be greater in the case of a ductile specimen, but this could not be divined from the diagram. The effect of machine stiffness will be discussed more fully at a later stage, but this accidental instability serves to illustrate the caution with which such indicated energy values must be employed.

SECTION 7 - YIELD STRESS TESTS

7.1. Apparatus

Special axial loading shackles were fitted to a 10 ton Buckton single lever testing machine for these tests. This type of machine was chosen for two main reasons. Firstly, because it was one of the few machines in which three extensometers set at 120° around a tensile specimen would be visible without obstruction by tie-rods etc., and secondly because the inherent stability of the loading system allowed plenty of time for the extensometer readings to be taken. The shackles have been fully described elsewhere (80), but consist basically of two mutually perpendicular knife-edges together with a ball thrust bearing. The complete assembly is shown in Plate 5.

The type of specimen used is shown in Appendix 'C' Fig. C.3. The enlarged upper portion was necessary for the low temperature tests to avoid yield occurring in the warmer section not immersed in the bath rather than at some point within the gauge length. It was necessary to use an adaptor to make the threaded end of the specimen fit the bottom shackle, and it was found convenient to make this adaptor integral with the cooling tank.

A special triple extensometer unit was constructed to measure the extensions on three mutually perpendicular 4 in. gauge lengths at any desired temperature within the range

+ 120°C. to -100°C. and this is fully described in Appendix 'G'.

Temperature Control

A similar system to that used in the notch-tensile tests was employed, except that the desirability of a double tank unit had not been discovered at this stage. Thermocouples were attached at each end of the gauge length and no test was performed until the two thermocouples gave the same reading. Initially the temperature difference between the two points was often as much as 10°C., but once the desired temperature had been attained it was not difficult to maintain it without a temperature gradient along the specimen.

7.2. Testing Procedure

A series of preliminary tests was performed at room temperature to check both the accuracy of the extensometers and also the axiality provided by the shackles. It was expected that despite the special shackles there would still be a certain amount of bending induced in the specimen, and before any test was performed the load was cycled several times between about 10% and 80% of the expected yield load in an attempt to make the loading conditions as uniform as possible. The first results revealed different values of extension under each extensometer, as would be expected if any bending were present. In addition, however, neither were the individual load-extension plots linear, nor did the points plotted for

load decreasing fall on the same line. The mean value of the three extensions, however, did give a good straight line plot against load, there being no detectable difference between the points plotted for load increasing or decreasing. The mean line also gave a consistent value of 30.5×10^6 p.s.i. for the modulus of elasticity for the specimen of steel 'V'. In view of these results it was suspected that the shackles might be responsible. In order to check this, the extensometer unit was rotated through 120° and the test repeated after prior cycling. The unit was then rotated through 120° once more, again cycled and a further test made. The results of the three tests showed that it made little difference which extensometer was used to record the extension on a given plane; much the same load extension curves were obtained in each case.

It was concluded that the degree of bending induced by the shackles varied as the load was increased, causing the individual load-extension curves from each extensometer to deviate from the expected straight line. Various adjustments of the shackles were then tried and, although it was never possible to eliminate non-linearity in the initial portions of the curves, the three extensometers could be made to give straight line plots for the majority of their length.

These tests were designed to determine the static lower yield stress over a range of temperatures, and also if possible

to give an estimate of an upper yield stress at these temperatures. Two methods have been suggested by which upper yield stress values can be obtained from tests of this type. The first was Morrison's method (see Appendix 'G' part (2)) to calculate the position of the plane of maximum strain. It was suggested that if one extensometer were set on this plane, a linear load-extension line would be obtained up to the upper yield point, which would be indicated by a sudden large increase in extension. The maximum strain and the plane on which it occurred were duly calculated, and it was found that with experience the shackles could be adjusted to limit the amount of bending strain to about $\pm 4\%$. It was unfortunate, however, that when the shackles were adjusted for the maximum axiality the plane of maximum strain occurred on that side of the specimen facing the pillar of the testing machine, thus preventing any reading from being taken.

The second method, suggested by Warnock and Brennan and outlined in Appendix 'G' part (3), was therefore used for any estimations of the upper yield stress.

With the shackles adjusted as so described, the tank temperature was adjusted in the usual manner, after which time great care was taken to keep the temperature as nearly constant as possible. About fifteen minutes were allowed for conditions to become steady during which time the load was cycled several times. After this some scale readings were taken as a check on

the axiality and, provided that this was satisfactory and that the thermocouples were giving similar constant readings, the test was started. The load was carefully increased in small steps, a few minutes being allowed after the application of load before the scale readings were recorded. This was continued until a sudden yield occurred. Tests of this type were performed at various temperatures over the range $+118^{\circ}\text{C.}$ to -93°C. At the lowest temperatures for each of the two steels the load was decreased as soon as unstable yielding began. The temperature was then decreased still further and the load again slowly increased until yield again occurred. It was thought that the amount of plastic flow which occurred during the first yield was insufficient to cause work-hardening and that the second yield was therefore a valid point. The latter results are shown in parentheses in Appendix 'A' Table 15.

7.3. Results and Discussion of Results

At the time that the above tests were planned no other results were available giving the variation of yield stress of the two steels with temperature or strain rate. It was thought that a knowledge of the variation of static yield stresses over a range of temperature would be useful in any future analysis of the results. The author was aware that considerable doubt might be cast upon the upper yield stress values as determined by Warnock and Brennan's method, but no more suitable apparatus was then available. When the tests

were actually being performed, however, the author was notified that a set of static and dynamic tests on the Admiralty series of steels was planned and these results have very recently been made available by Taylor (81). They give static and dynamic, upper and lower yield stress values over a range of temperature (see Appendix 'A' Tables 17 and 18). It was hoped that the variation with stress rate would also be available, but this did not prove possible and only a small change in the rate of dynamic stressing could be achieved (Appendix 'A' Table 19).

The upper yield stress values are largely governed by experimental technique, in fact the experimentally determined upper yield stress may be as much as 100% greater than the lower yield stress if sufficient care is taken (82, 83). (This would not be predicted by Tarnock and Brennan's method.) It would thus seem more reasonable to employ Taylor's results for the purposes of any analysis since they were all obtained on the same Hopkinson bar machine. The author's results are listed in Table 15 but, because of the decision to employ Taylor's results, the somewhat tedious calculation of upper yield stress has only been made in two cases, merely to draw a comparison.

A few room temperature tests at low rates of loading were also performed on miniature Houndsfield specimens in an Instron testing machine and the yield stress values obtained are given

in Appendix 'A', Table 16. Slight evidence of an upper yield stress was obtained in two cases only. Although the machine tested at constant rates of cross-head movement, there was often a deviation from a linear relationship before a true yield occurred. It was thus not always possible to quote a rate of stressing for these tests.

To examine the experimental results of the notch-bend and notch-tensile tests in the light of some of the different hypotheses, however, the yield stresses (especially the upper yield stress) are required over a wider range of stress rates. We have already noted that there are in existence various theoretical predictions of the upper yield stress (see appendix 'B'). There may be disagreement as to the validity of some of the physical concepts involved, but they have been found to give fair agreement with experimentally determined values over quite a wide range of temperatures and stress rates. Let us take equation (B.3):-

$$\sigma_{yd} = \sigma_o \left((1 + \lambda) C \dot{\sigma} / \sigma_o \right)^{\frac{1}{(1+\lambda)}}$$

Now λ may be expressed in the form $\lambda = \frac{E_o}{kT}$

Let us take $\sigma_o = 1.68 \times 10^5$ p.s.i. and $E_o = 0.435$ e.v. (which was the value obtained by Hendrickson and Wood (21) on a 0.17% carbon steel). Also take kT at 20°C. as $1/40$ e.v.

Substitution of Taylor's experimental values into equation (B.3) allowed C to be evaluated. It was necessary

to assume a value for the stress rate in the static tests by taking an arbitrary time to yield of $\frac{1}{2}$ minute. There was of course a fair degree of scatter, but the following mean values were determined.

For steel 'P', $C = 10^{-11}$ sec.

For steel 'V', $C = 1.5 \times 10^{-9}$ sec.

The theoretical upper yield stress against temperature curve at the two stress rates was then calculated using the above constants, and is compared with the actual experimental points in Figs. 39 and 40. The theoretical curves are seen to be fair approximations to the experimental points in most cases but tended to predict values which were rather too low at the lower temperatures. This is in agreement with the findings of Hendrickson and Wood. It was decided to add an arbitrary correction factor to the curve. This was taken to increase linearly with decreasing temperature from 0 at -80°C . to +12% at -200°C . The corrected theoretical curve is shown dashed in Figs. 39 and 40. Bearing in mind the experimental accuracy obtainable, the corrected curve may be said to be a fair representation of the observed values. A similarly corrected version of equation (B.3) will therefore be used to obtain any upper yield stress values which may be required later.

SECTION 8 - GENERAL DISCUSSION

8.1. A Correlation of Transition Temperatures

A very large amount of research effort has been devoted towards a correlation of the various tests both with each other and also with service failures. For example, different modifications have been made to the standard Charpy specimen in attempts to develop an inexpensive way of predicting behaviour in wide plate tests. It will be apparent, however, from Section 2, that the number of different factors that are altered in changing from test to test is often quite large and an apparent correlation may well be the fortunate collation of various effects. The author's results will be presented together with the results of Tipper and Van der Veen tests on the same steels to show as far as possible the effects of changing some of the individual variables. This was intended as a comparison between some of the standard engineering tests for brittle fracture and thus, although care was taken in the preparation of specimens, no special attempt was made, for example, to prevent the formation of the residual machining stresses. To obtain a strict comparison in the light of some of the hypotheses it would have been desirable to have performed some controlled atmosphere heat-treatment to keep the grain size/specimen dimension ratio constant and also to relieve residual stresses. The tests being compared would not then be the standard engineering tests, however unsuitable these

may be thought. It was decided to accept the factor of a constant grain size irrespective of specimen size, but another variable which was unintentionally allowed to remain was the machine stiffness factor, which may have a stronger influence than was thought (see Section 8.2.).

The charts, Tables 21 - 23, show how the transition temperature as judged by a given percentage crystallinity criterion are affected by the three variables (1) strain rate, (2) size, (3) stress system (i.e. notch-tension or notch-bending). In these charts the concentric semi-circles represent the different sizes of specimen, the size increasing with the radius of the semi-circle from half size Charpy notch-bend to full plate thickness specimens. The centre sector shows the notch-bend tests at a constant slow rate of elastic straining. In the left hand sector the elastic strain-rate is virtually unchanged, but the stress system is changed from notch bending to notch-tension. In the right hand sector the stress system is unchanged, but now the rate of elastic straining is changed in two steps to the impact state. To distinguish the two steels the values for steel 'V' are placed at the right hand side of any compartment and are underlined. The data shown in these charts is also presented in a modified form in Table 24, where the effects of the individual variables are separated.

20% CRYSTALLINE TRANSITION TEMPERATURES

VALUES FOR STEEL V
ARE UNDERLINED

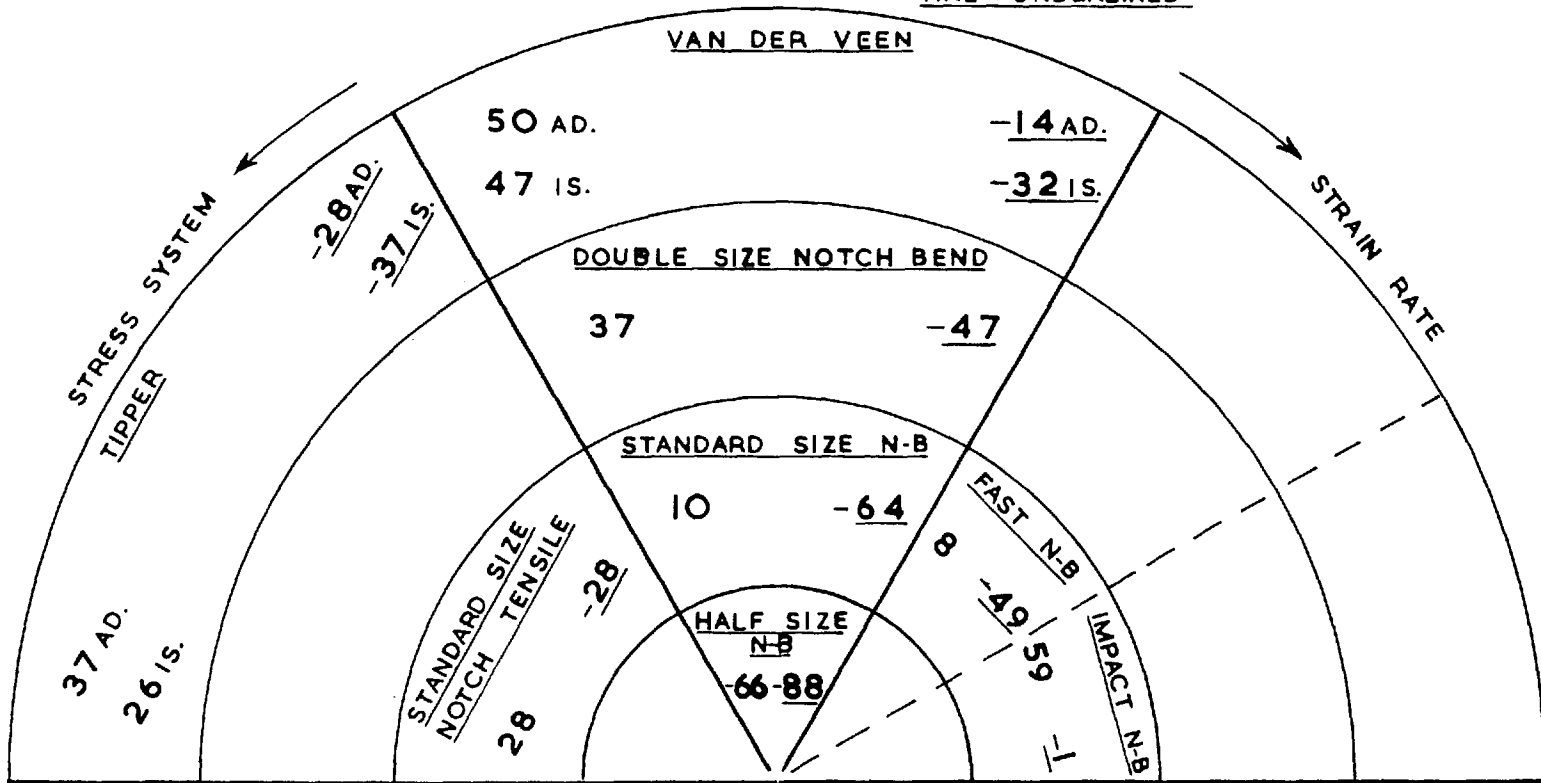
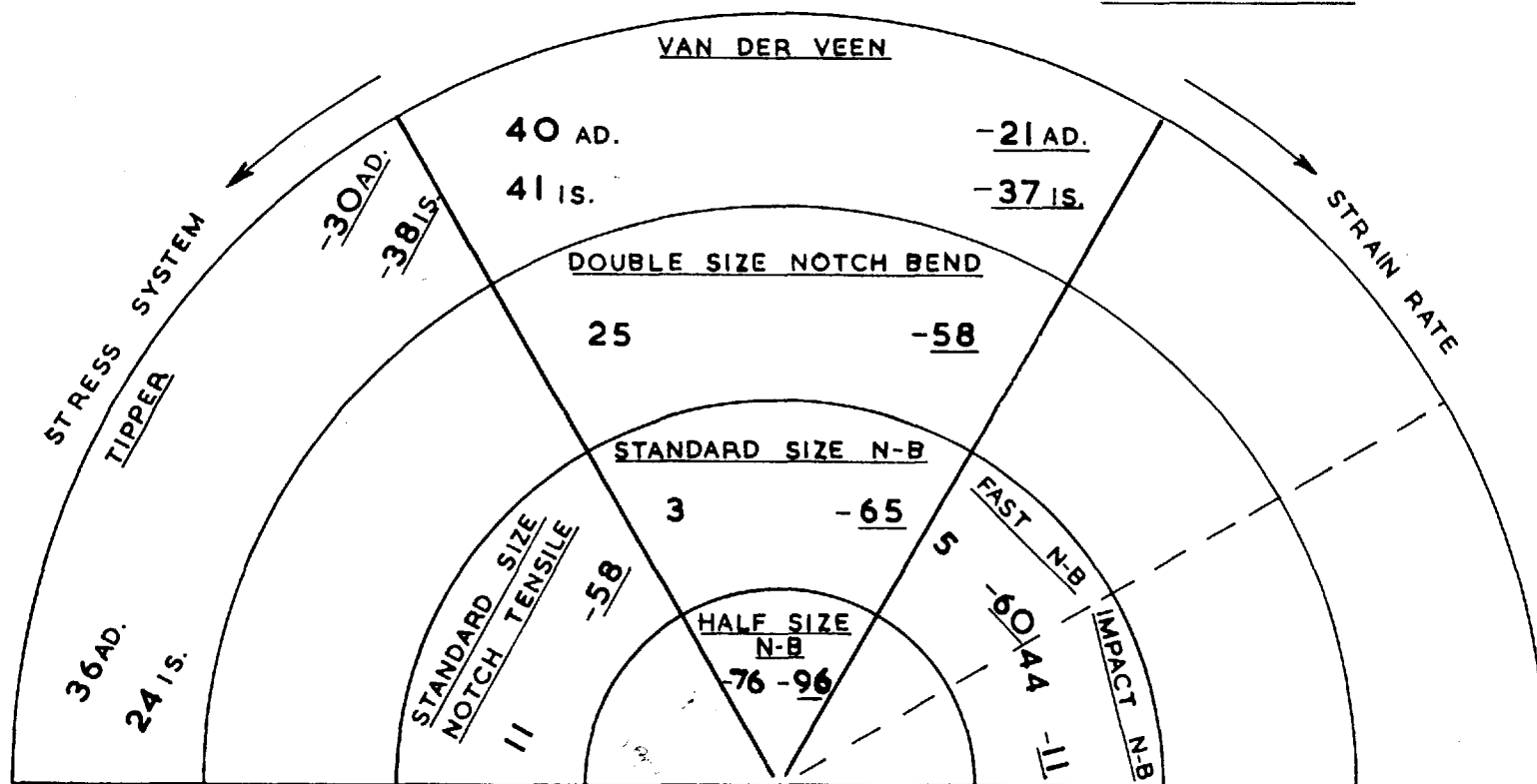


TABLE 21

50 % CRYSTALLINE TRANSITION TEMPERATURES

VALUES FOR STEEL V
ARE UNDERLINED

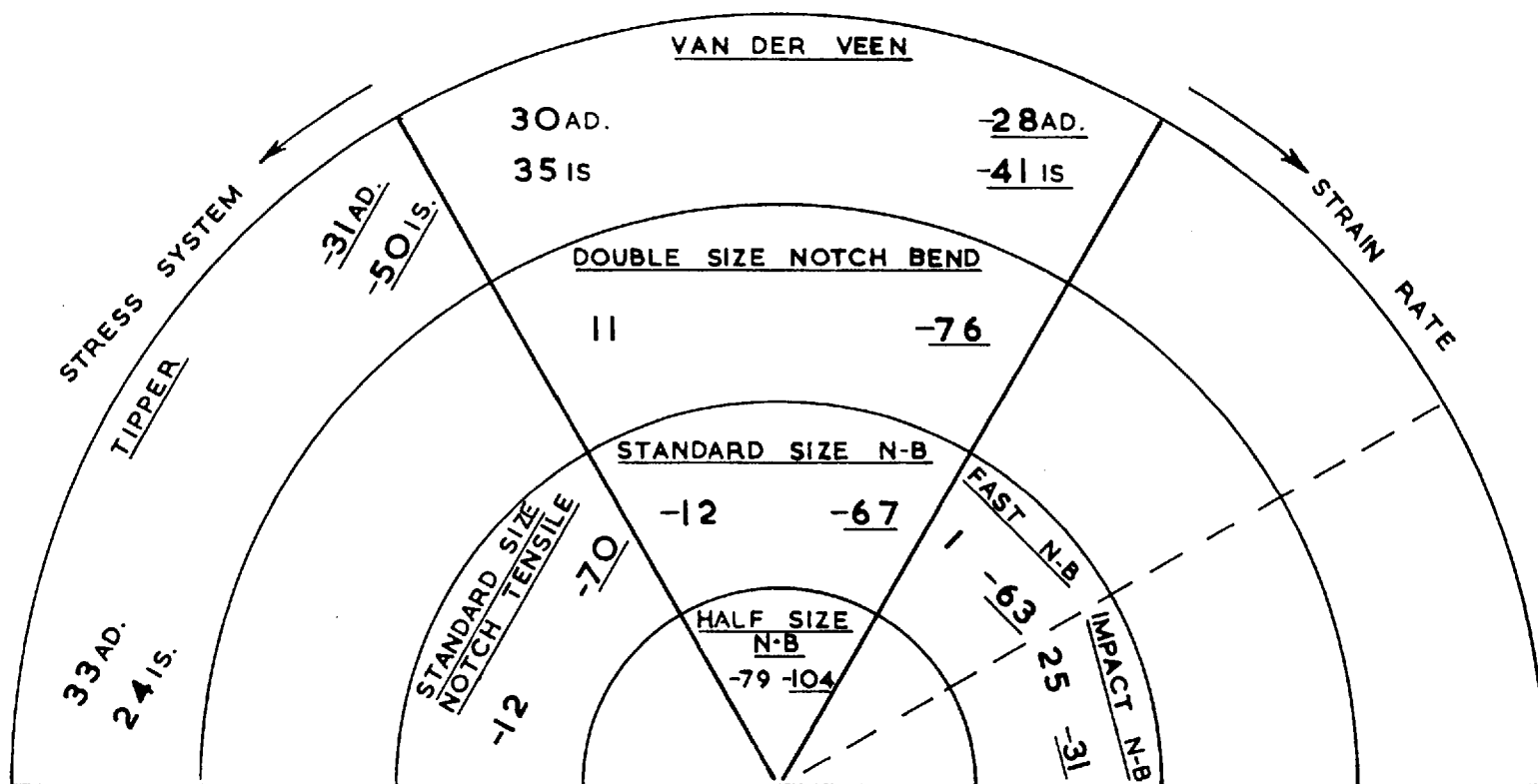
TABLE 22



80% CRYSTALLINE TRANSITION TEMPERATURES

VALUES FOR STEEL V
ARE UNDERLINED

TABLE 23



Before any discussion is made on the results shown in the charts the degree of accuracy available must be clearly understood. It is not always possible to define a precise transition temperature for a given material from the results obtained in a given test. Those tests which do not exhibit a sharp clear-cut transition tends to show scatter in the transition region and since normally it is not practicable to take more than a few points in this region the final transition curve drawn may vary by several degrees depending on the points taken. This is particularly true at the "elbows" at either end of the transition range, it is seen from Fig. 42, for example, how one or two extra points could alter the shape of the curve in the brittle region quite considerably and with it the 30% crystalline transition temperature.

The results for the Tipper and Van der Veen tests have been taken from a report by the Admiralty Advisory Committee on Structural Steel (67). The data available in this report was collected from eight different sources, although in fact only three laboratories carried out the Tipper test and two the Van der Veen test. Thus, although nominally the same steels were used as in the author's test it is quite possible that variations did exist from plate to plate. Also some laboratories performed the tests "adiabatically", cooling the specimen to a suitable temperature below the desired fracture temperature and then recording the actual temperature at fracture; whilst others performed them "isothermally", the

specimens being immersed in a suitable bath. In general the transition temperature for tests performed isothermally were of the order of 10°C . lower than the corresponding transition temperature for adiabatic tests. Since the author's tests were all performed isothermally, the isothermal values of transition temperature for the Tipper and Van der Veen tests probably form a fairer basis of comparison, and the isothermal values have been used in Table 10 unless otherwise stated.

It has been stated that the elastic strain rate at the root of the notch is the same for all tests in the centre and left hand sectors. This is true for all of the author's tests and approximately so for the Van der Veen test, but for the same strain rate to be achieved in the Tipper test the rate of loading should have been increased quite considerably. At moderate rates of loading, however, the effect on the transition temperature is not great, as can be seen from Table 24 by the change from slow notch-bend to fast notch-bend. Thus the Tipper test results are probably only about five degrees lower than the value for a true comparison, at least as far as strain-rate is concerned.

Another point to be considered is the fact that the notch in the Van der Veen test is a pressed notch whereas all the others are machined. Van der Veen has shown, from a limited number of tests, that it would appear that the transition temperature obtained with machined notches is approximately

10°C. lower than those obtained with pressed notches (113), and this should be borne in mind when using the Van der Veen results for any comparison.

The Combined Effect of the Various Factors on the Transition Temperature

(1) The Strain Rate Effect

The change of strain rate from impact to slow notch-bend caused a marked decrease in transition temperature, the numerical value of this drop decreasing with the value of the % crystallinity criterion; i.e. the change in transition temperature was greater when judged by a 20% crystallinity criterion than by an 80% crystalline one. This was true for both steels although the numerical values of this decrease were a little larger in the case of Steel V. The results obtained for the change from fast to slow notch-bend are not entirely consistent because the uncertainty as to the exact position of the curve, particularly at the two ends, tended to mask the small change involved. It seems fair to say, however, that for the steels tested a change in elastic strain rate at the root of the notch from 2.02×10^5 to 2.7×10^6 p.s.i./sec. caused only about 5°C. decrease in transition temperature.

It is perhaps unfortunate that the Charpy test, which has developed from a quite arbitrary test to study impact properties, is so widespread in its use. It was pointed out in Section 5 that the stress system produced by the impact is extremely complex and the nominal rate of elastic stressing

of 17×10^9 p.s.i./sec. derived in Appendix 'E' can only be a rough approximation. It may not be justifiable to consider the change from fast to impact notch-bend as a simple strain rate effect because, at least in the early stages, the rate of increase in stress is governed by the stress wave condition as are also the complexity of stress and the effective length of the specimen. It would seem that a controlled study of the rate effect in notched specimens might be more suitably performed on a long notch-tensile specimen in which an analysis of the stress conditions would be more feasible.

(2) The Size Effect

There is less consistency in the results for the effect of size, which are shown in Figs. 41 and 42. The results on steel P show that the % crystallinity criterion chosen does not make a large amount of difference in most cases. But the results on steel V are not consistent, the double and standard size curves crossing in the high crystallinity region (see Fig. 42). It has already been suggested that this anomaly may have been caused by testing an insufficient number of specimens in the transition region. The effect of size would appear to be more severe in the smaller specimens as is shown in Figs. 51 and 52, where the transition temperatures as judged by various crystallinity criteria are plotted against the specimen size. The Von der Veen specimen was not geometrically similar to the other specimens and so an arbitrary dimension of 42.2 mm. equal to $(\text{cross-sectional area})^{1/2}$

was chosen for this plot. But the general shape of the curve would have remained unchanged had either the actual width or depth been used instead. If a correction factor of $10^{\circ}\text{C}.$ were to be subtracted from the Van der Veen transition temperatures to account for the effect of the pressed notch then it is seen from Tables 21 - 24 that there is little difference between the double size slow notch-bend results of the author and the Isothermal Van der Veen test results. This would imply that the curves in Figs. 51 and 52 tend to become almost horizontal at the larger dimensions.

For those steels which exhibit a size effect, it should theoretically be possible to perform tests, at a given temperature, on a series of geometrically similar specimens of different sizes. This was in fact suggested in a discussion (114) on Docherty's slow-bend tests (63). It is possible to plot curves of this type using existing results. If an arbitrary temperature is chosen on the curves of % crystallinity against temperature (Figs. 41, 42, and also from Ref. 67), then the % crystallinity corresponding to each size of specimen may be recorded. These "size transition" curves are shown in Figs. 53 and 54. These curves show how the transition size, as judged by any given % crystallinity criterion, increases with increasing temperature. The exact shape of the curves is not known with accuracy, for only four points are available for each curve one of which (the Van der Veen) is only approximate. The type of curve

expected, however, is indicated and Fig. 53 would appear to suggest that the effect of temperature on the transition size is more severe at the higher temperatures. In the same way that Figs. 51 and 52 were derived from curves of crystallinity against temperature, such as Fig. 47, it should theoretically be possible to derive curves of say 50% crystalline transition size against temperature from Fig. 53. On reflection, however, it will be seen that this would in fact be the same curve as Fig. 51. Thus the shape of Fig. 51 say could equally well be described as saying that the effect of size on the 50% crystalline transition temperature is more severe at the smaller sizes, or that the effect of temperature on the 50% crystalline transition size is more severe at the higher temperatures.

8.2. The Machine Stiffness Effect

The effect of the stiffness of the testing machine on the transition temperature has been shown to be small in many cases, but the available evidence is somewhat conflicting (108,109). It is quite possible, however, to obtain ductile instability in slow notch-bend tests (110) in a similar manner to that described for notch-tensile specimens in Section 6. Under these circumstances the transition temperature, as judged by a certain value of indicated energy absorbed, may be considerably altered. When the results of some slow notch-bend tests performed on specimens of steels 'F' and 'V' in a Hounsfield Tensometer (111) were compared with those of the

author it was found that, whilst the crystallinity transition temperatures were unaltered, the energy - temperature curves were entirely different. For steel 'V' the overall energy level was higher and the transition temperature nearly 20°C . lower than in the author's tests, whilst for steel 'P' the reverse was true (i.e. the overall energy level was lower and the transition temperature 15°C . to 20°C . higher). Thus since, as far as the author is aware, there was no evidence of ductile instability it seems that the testing machine may, under certain circumstances, have a considerable effect. This raises doubts as to whether the comparatively small change in transition temperature between slow and fast notch-bend tests might be due to the masking effect of a change in effective machine stiffness, by the introduction of the ring-type load cell for the fast tests.

8.3. A Comparison of Loads at General Yield in the Different Tests.

Before any criterion of brittle fracture is considered the yield loads obtained in the various tests will be examined and the relationships between them determined. The following subscripts will be used throughout:-

- snb. = Standard size notch-bend
- hnb. = Half size notch-bend
- dnb. = Double size notch-bend
- snt. = Standard size notch-tensile

The same shape of notch is employed in all the specimens and therefore the elastic stress distribution beneath the notch

is likely to be the same in each case and the yield stresses will be given by:-

$$(\sigma_{nb})_{yd} = \frac{k'PL}{4Z'} \quad \text{and} \quad (\sigma_{nt})_{yd} = \frac{k'P}{A} \quad \text{-----} (8.1)$$

where P = Applied Load.

L = Distance between specimen support.

Z' = Section modulus of the notched cross-section.

A = Cross-sectional area at the notched cross-section.

k' = A constant depending on the yield criterion chosen.

If the appropriate specimen dimensions are substituted in the above equations then we obtain the following ratio for the yield loads:-

$$P_{hnb.} : P_{snb.} : P_{dnb.} : P_{snt.} = \frac{1}{4} : 1 : 4 : 7.5 \text{-----} (8.2)$$

If we consider the alternative case of the loads required to produce a fully plastic notched cross-section then:

$$P_{hnb.} : P_{snb.} : P_{dnb.} : P_{snt.} = \frac{1}{4} : 1 : 4 : 5 \text{-----} (8.3.)$$

The various yield loads have been plotted against temperature and are shown in Figs. 56 to 65. Fracture load curves are also shown but these will be discussed later. Smooth curves have in general been drawn through the plotted points for yield loads but there is doubt in some cases as to the best representation of the plotted points. The yield loads corresponding to the different temperatures were read off these curves and are listed in Table 25 (page 137). The yield loads obtained at a given temperature in the different tests could then be compared. This has been done in Figs. 66 & 67, where the various notch-bend yield loads and the static lower yield stress are plotted on a base of the standard notch-tensile yield load.

It may be seen that these curves are difficult

TABLE 25

The Variation of Yield Load with Temperature.

Temp. °C.	Yield Load Steel P Tons					Yield Load Steel V Tons				
	snt.	hnb.	snb.	dnb.	Static L.Y.S. 10 ⁴ psi	snt.	hnb.	snb.	dnb.	Static L.Y.S. 10 ⁴ psi
50	2.60	-	-	2.17	2.94	-	-	-	2.48	3.54
40	2.65	-	-	2.20	2.95	-	-	-	2.52	3.56
30	2.69	-	.404	2.24	2.98	3.125	-	-	2.58	3.59
20	2.73	.131	.42	2.28	3.00	3.22	.123	-	2.62	3.62
10	2.82	.135	.44	2.33	3.01	3.31	.131	-	2.68	3.64
0	2.92	.138	.46	2.43	3.025	3.40	.138	-	2.72	3.67
-10	3.03	.1425	.486	2.58	3.05	3.50	.146	.56	2.78	3.71
-20	3.17	.1475	.515	2.72	3.09	3.60	.153	.57	2.86	3.76
-30	3.31	.153	.542	2.86	3.14	3.70	.161	.59	3.00	3.82
-40	3.42	.160	.575	3.00	3.19	3.90	.167	.615	3.16	3.90
-50	3.67	.1675	.610	3.15	3.25	4.11	.176	.64	3.28	4.00
-60	3.91	.177	.645	3.29	3.35	4.33	.184	.675	3.43	4.11
-70	-	.187	.690	3.44	3.46	4.56	.192	.710	3.57	4.25
-80	-	.198	.740	3.58	3.62	4.78	.200	.750	3.71	4.41
-90	-	.210	.795	3.72	3.85	5.00	.207	.798	3.85	4.60
-100	-	.222	-	-	4.20	5.22	.215	.846	3.99	4.83

to interpret for, whilst the points for hnb. and snb. specimens lie within the limits for the elastic and fully plastic relationships, those for the دنب. specimens lie on or above the fully plastic line. As a direct comparison of the three sizes of bend specimen the respective yield loads have been plotted, on a log. basis, in Fig. 68. The expected yield load ratios are shown as full lines and it is seen that not only is the expected ratio hnb. : snb : دنب. = $\frac{1}{4}$: 1 : 4 not followed (the experimental ratios being 0.276 : 1 : 4.25), but also a non-linearity is displayed which follows the same trend in both cases.

The static lower yield stress has also been plotted in Figs. 66 and 67 and it is seen that there does not appear to be any obvious relation between this stress and the yield loads in either the notch-bend or the notch-tensile tests. If we are to relate the uniaxial yield stress with the yield loads then we must consider (i) The value of the strain rate at the root of the notch and (ii) The effective stress concentration factor that is operative.

To obtain an estimate of strain rate, let us take the case of a specimen in which the curved arcs, which mark the beginning of general yield, have just formed. If the geometry of the standard size specimen is considered then for a total angle of bend θ^0 , initially neglecting any indentation by the striker or by the anvils, the deflection of the striker may be

shown to be:-

$$\delta = 10 \left(1 - \frac{(1 - 2 \sin. \frac{1}{2} \theta)}{2 \cos. \frac{1}{2} \theta} \right) \text{ mm.} \quad (8.6)$$

or, for small angles of bend

$$\delta = 10 \theta \text{ mm} \quad (8.7)$$

Now the deflection at yield is of the order of 0.010 in., say $\frac{1}{4}$ mm.

$$\therefore \theta = 0.025 \text{ rad.} = 0.143^\circ$$

and from equation (2.22) the C.O.D., 'b' is given by

$$b = 2.49 \theta \times 10^{-3} \text{ in.} = 3.56 \times 10^{-4} \text{ in.}$$

Therefore, for a striker displacement of 0.050 in., there is a C.O.D. of 3.56×10^{-4} in. and, since this displacement is calculated over a gauge length which is approximately equal to the notch root diameter, then the strain is 0.00686. Thus for a rate of striker travel of 1.5 in./mm. the strain rate at yield is $0.00686 \times \frac{1.5}{0.010}$ per min. $\div$ 1 per min.

Some account must also be taken of the indentation of the striker, and, as a very approximate estimate, it will be assumed that, at this stage of the test, half the striker movement may be accounted for in indentation and half in bending the specimen. Thus the estimated rate of straining at the root of the notch at yield is approximately 0.5/min. If, in Table 16, the strain is assumed to be equal over the gauge length of the miniature Houndsfield specimen ($\div$ 0.65 in.) then from a knowledge of the rate of cross-head movement, the strain rates at yield may be calculated. It can be seen that the increase in the yield stress at 0.5/min. over the static

value will only be about 10%. Even in the dynamic yield stress tests of Taylor (81), where the rate of plastic straining at the lower yield was estimated at 1,920/sec., the increase in the lower yield stress over the static value was 38% for steel 'P' and only 15% for steel 'V'.

It is most unlikely that the elastic stress concentration factor is still effective at the point of general yield, and instead the rigid-plastic analysis of Alexander and Komoly (40) may be used to predict the general yield load.

For the V-notch Charpy specimen, taking into account the circular notch-root they obtained a value for the constraint factor of $L' = 1.2979$.

$$\text{where } L' = \frac{2M'}{ka^2} \quad \text{-----} \quad (8.4)$$

and M' = The Collapse Moment/unit width

a = Depth of specimen below the notch root

$$k = \frac{\sigma_{yd}}{2} \text{ (Tresca) } \text{ or } = \frac{\sigma_{yd}}{\sqrt{3}} \text{ (Von Mises)}$$

The specimen geometry may be substituted in the above equation to give:-

$$\frac{P_{yd} \text{ (Tons)}}{k \text{ (p.s.i.)}} = A, \text{ say } \text{-----} \quad (8.5)$$

Where A has the values 0.72×10^{-5} (Half Size), 1.88×10^{-5} (Standard Size) and 11.5×10^{-5} (Double Size).

In general terms it may be stated that values of the general yield load calculated in this manner were in fair agreement with the experimental values for specimens which were

just in the fully ductile range (e.g. standard size steel 'V' at $-40^{\circ}\text{C}.$). At temperatures above this the predicted load tended to be too low, and at lower temperatures the predicted loads were too high.

8.4. An Examination of the Results with regard to the Hendrickson, Wood and Clark Hypothesis

It is necessary, first of all, to consider what transition would be predicted if this hypothesis were in fact correct. They have suggested that the material may be considered brittle if the plastic zones are confined to the vicinity of the notch root and ductile when the zone becomes unstable and spreads across the specimen. There is no doubt that such a transition does take place and Knott and Cottrell (112), using Fry's reagent on a high nitrogen steel, have obtained clear pictures of such deformation patterns. They found, however, that this transition took place at a much lower temperature than the 20 ft. lb. transition temperature in the Izod impact test, which was predicted by Hendrickson, Wood and Clark. Knott and Cottrell performed pure bending tests on notched specimens, and plotted the load at yield and the load at fracture against temperature. They found that the temperature at which these two curves crossed coincided with the temperature of incipient instability of the plastic zone.

Similar curves have been plotted from the author's results and are shown in Figs. 56 to 55. The load at the first brittle crack has been plotted for the fracture load and those cracks occurring after the maximum load account for the way in

which this curve falls away rapidly at the higher temperatures. It is, of course, not possible to obtain values for the yield stress at temperatures below the point of intersection of the two curves (unlike Knott and Cottrell who were able to turn the specimen over so as to put the notch in compression), but, whenever sufficient points have been taken in the low temperature region, any point of intersection could still be determined. The two steels used did not respond to Fry's reagent but the surface deformation patterns of a few of the specimens were examined using the Schlieren technique. This is a reflected light technique which requires polished specimens and reveals the deformed regions as either light or dark areas (depending on the kind of stop used). Although the specimens were not as highly polished as might be desired, the resultant pictures (see Plate 6) reveal clearly how the plastically deformed zones, shown as light areas, become more and more confined to the vicinity of the notch as the temperature is decreased over this range. The white bands at the top and bottom of some of the specimens are only produced by a slight bevelling of the corners during machining. The surface patterns may be taken as indicative of the deformation at the centre of the specimen even if the actual pattern differs somewhat. Comparisons are available of the Fry's etch pattern at the centre and the Schlieren surface photographs (110).

Thus, considerable doubt is thrown on the deformation patterns which Hendrickson, Wood and Clark obtained by means

of microhardness tests. For plastic deformation to be revealed by a change in hardness the metal in that region must not only have yielded but also work hardened, and yet one of the assumptions made in the derivation of the theory was that no work hardening took place before plastic instability. It has also been shown that, after the plastic arcs have spread across the specimen, then further deformation can produce "wings" which spread from the region of the notch root up to the notched surface. These wings are thus an indication of gross plastic flow, and yet such a wing is shown in the diagram of Hendrickson, Wood and Clark which is meant to illustrate restricted deformation (Ref. 23 Fig. 7).

The author's curves of the yield and fracture loads do not follow exactly the same form as those of Knott. If we examine those for the standard and double size specimens of steel 'P' (Figs. 57 and 58) it is seen that, after falling away with decreasing temperature, the fracture curve appears to rise again. In fact, in Fig. 58 the two curves just fail to intersect at the minimum point.

Whether or not this is true for the standard size specimens as shown in Fig. 57 depends on whether or not yield has occurred in the test at -106°C . It was thought that they were probably coincident and the point is so plotted, but if we consider that the ratio of the yield load at -100°C . to that at 0°C . is approximately the same as for the half size specimens then it is seen that yield has probably not occurred. Thus it

seems likely that the standard and double size curves have much the same form. A similar kink is obtained in the curve for notch-tensile specimens (although this time it does appear that the two curves intersect) and one is also suggested by the dashed curve in Fig. 59, for the fast notch-bend.

In all these cases the discontinuity in the curve may be shown to coincide approximately with the final disappearance of any fibrous region, which usually means the shear thumb-nail. In the case of steel 'V' it could be argued that a similar kink is exhibited in the curve for the fast notch-bend tests (Fig. 64), but this could with equal justification merely be attributed to scatter. The remaining tests on this steel do not provide sufficient points to determine whether or not the curve rises again. The correspondence with the disappearance of the fibrous region may still be valid.

It is very much to be regretted that the desirability of obtaining the surface deformation patterns was not realised when the specimens were being made so that suitably polished surfaces could have been prepared. The surface finish on some of the double size specimens was, however, adequate at least to indicate the type of deformation although photographing the patterns was difficult. Plate 7 shows three of the double size specimens of steel 'P'. General yield was clearly indicated in specimen P.10 at -85°C . and this is just discernible as white arcs in the photograph. For P.11 at -99°C . it was found convenient to show the deformation zones as dark patches and it can clearly be seen that yield zones are

just starting to spread across the specimen; experience would suggest therefore that general yield has not occurred at the centre of the specimen. At -125°C . only localised yield has occurred. The deformation patterns shown in Plates 6 and 7 would then seem to indicate that this kink in the fracture load curve reveals the point of plastic instability, irrespective of whether or not the fracture and yield loads actually intersect.

It is interesting to compare the shape of these fracture curves with the tensile fracture stress curves of Hahn et al. (13) (see Fig. 3) and with the comparison of similar curves with nominal fracture stress curves by Knott and Cottrell (see Ref. 112, Fig. 6). No metallurgical examination has been made, however, to investigate the possibility of the specimen fracturing as soon as it starts to yield by deformation twinning.

Returning to the Hendrickson, Wood and Clark hypothesis, it will be noticed that no size effect is predicted. By a theoretical analysis they calculate the nominal bending stress/yield stress ratio at instability as $\frac{\sigma_{nb}}{\sigma_{yd}} = k_1$ and at this stress the maximum stress is given by the equation $\frac{\sigma_m}{\sigma_{yd}} = k_2$. For the transition point the maximum stress is, by hypothesis, the critical brittle fracture stress σ_c . The values $k_1 = 0.8$ and $k_2 = 2.41$ are likely to be much the same for the Charpy specimen as for the Izod, but their theoretical derivation is independent of size. Thus the only remaining term, by means of which a size effect can be introduced, is the yield stress.

For the standard size specimens of steel 'P' it is suggested that the instability point, as indicated by the bottom of the kink in the fracture load curve of Fig. 57, occurs at approximately -106°C . The upper yield stress at this temperature, for the assumed rate of elastic stressing at the notch root of 2.02×10^5 p.s.i., has been calculated by the methods of Section 7 as 8.52×10^4 p.s.i. If k_2 is then assumed to be 2.41 the value of $\sigma_m = \sigma_c$ is 2.06×10^5 p.s.i. which is in good agreement with the value of Handrickson, Wood and Clark. (It has been suggested by Knott and Cottrell that the critical tensile stress criterion for fracture is valid, but at higher temperatures general yield followed by work hardening is necessary before this critical stress can be reached.)

Similar calculations were made for the double size specimens and the upper yield stress, at the temperature corresponding to instability, was found to be 7.15×10^4 p.s.i. Now this difference between the yield stresses at instability in the two sizes of specimen can be explained purely on the basis of the effect of specimen size upon the yield stress. From the work of Richards on the yield stress in unnotched beams (65) we would expect the yield stress to decrease by 15.4% for a change of specimen size from 10mm. to 20mm. If the upper yield stresses, at a given temperature and stress rate, vary in the same ratio with size, then this would predict an upper yield stress of 7.21×10^4 p.s.i. Similarly the

predicted upper yield stress in the half size specimens would be 9.86×10^4 p.s.i. The temperature at which such a yield stress would be obtained has been calculated to be about -130°C . It may be seen from Fig. 56 that an instability point at this temperature would be quite consistent with the results already obtained.

With regard to the effect of strain rate, it can be shown that, at the suggested rate of elastic stressing in the impact test (1.7×10^9 p.s.i./sec.), an upper yield stress of 8.52×10^4 p.s.i. would be attained at a temperature of about -80°C . The surface-finish of the specimens was not sufficiently good to permit examination by the Schlieren method, but it may be seen from Table 7 that the specimen first becomes fully crystalline at about -72°C . Thus it might be suggested that, within the limits imposed by ignorance of the exact condition of stressing, an instability point has again been predicted.

8.5. Some Comments on the Explanation of Size-Effect on the Basis of a Constant Width of Fibrous Frame.

It has been suggested by Boyd (115) that there is, at a given temperature a constant width of "fibrous frame", irrespective of the specimen. Boyd envisaged a symmetrical frame (i.e. all four sides of equal width), a state of affairs that was not found in the author's tests. The appearance of the impact specimen approached this most closely but it was felt that it was only justifiable to compare the three sizes of slow bend specimen.

Because of the unusual way in which the % crystallinity curves for steel 'V' cross over (see Fig. 42) it was decided to use the specimens of steel 'P' for the investigations. In these specimens the fibrous frame consisted of the well-known "Thumbnail" together with a somewhat wider fibrous band along the struck face. There were also fibrous edges to the frame which tended to be wedge shaped, decreasing in width towards the notch. The borders of the frame were irregular and it was not possible to measure them to an accuracy greater than 0.1mm. It can be seen from Fig. 70 that both the fibrous thumbnail and the fibrous edge may be said to be approximately constant at a given temperature, irrespective of whether a standard or double size specimen is considered. Considerable scatter was exhibited and the plot has not been continued into the region of large shear lips, because in this region the crystalline surface is only present in isolated patches.

In the half size specimens an entirely different surface appearance was obtained (see Fig. 69), only the fibrous thumbnail being present. Even at comparatively low values of % crystallinity the crystalline region was exhibited as a band along the struck face. It was not possible to effect any correlation of the half size specimens with the standard and double sizes by means of the shear thumbnail size. It may be stated, however, that the change from a completely fibrous surface to a full width crystalline band took place at a temperature approximately 10°C. below that at which the fibrous edge finally disappeared in the larger specimens.

Attempts were made to estimate the variation of the % crystallinity transition temperature with size on this basis, but no real agreement was obtained with the experimental results. It would seem from the work of Bluhm (34) that the width of the fibrous edge is likely to be constant at a given temperature, but that in the Charpy specimen this effect is masked by the end effects at initiation and arrest.

8.6. Brief Comments on the Crack Opening Dislocation Concept

It was shown in Section 2 that the C.O.D. may be expressed in terms of the angle of bend θ as:-

$$b = 2.49 \theta \times 10^{-3} \quad (\text{equation 2.22})$$

and in Section 8.3 it was shown how an estimate of θ , and hence 'b' could be made from the value of deflection. It is difficult to know what allowance to make for the indentation of the specimen, but it is interesting to make a comparison of the theoretical energy absorbed, as estimated through these concepts, with the measured values.

If fracture mechanics are to be used to interpret data from V-notch Charpy specimens then it would seem that the type of specimen most likely to conform to these ideas is one possessing a fully crystalline surface appearance and preferably also one which fractures before general yield. If we consider one of the fast notch-bend specimens, P 193 (-100°C . then it is seen from Fig. 59 that not only did fracture take place below the suggested instability temperature, but also

below the clearly indicated intersection of the yield and fracture load curves.

If indentation is neglected then the deflection at fracture (0.017 in.) may be used to calculate the angle of bend as 1.246° . Now Wells (38) has suggested the following relation for the angle of bend at fracture:-

$$\theta^\circ = 0.47 U$$

Where U is the energy absorbed in ft. lb.

For the value of $\theta = 1.246^\circ$ this would predict an energy absorption of 2.65 ft. lb. or 0.366 kg.m. This may be compared with the measured energy absorbed of 0.29 kg.m. The calculated deflection corresponding to the latter value of energy is 0.007 in. which would imply a value of 0.010 in. indentation.

An alternative approach is to use the relation between the upper yield stress and K_{1c} suggested by Krafft and Sullivan (116). This is:

$$M = K_{1c} \sigma_{uys}^{1.5}$$

Where M is a proportionality constant, which for steel 'P' would be of the order of 0.95×10^{12} . Now for the temperature and stress rate in question σ_{uys} may be calculated as 9.02×10^4 p.s.i., giving K_{1c} as 14.05 Ton/in.² $\sqrt{\text{in.}}$. This value may be substituted into the relation $K_{1c} = 8 \sqrt{U}$ Ton/in.² $\sqrt{\text{in.}}$ (where U is in ft. lb.), which was suggested by Wells, giving $U = 3.09$ ft. lb. or 0.428 kg.m.

If the value of indentation of 0.010 in. is correct then the true angle of bend at fracture is 0.988° , which from equation (2.22) makes 'b' = 0.0025, which is of the same order as the critical displacement suggested by Cottrell (see p.53).

SECTION 9 - CONCLUSIONS AND RECOMMENDATIONS FOR FUTURE WORK

Experimental evidence has been produced which allows the relative importance of some of the individual variables, that control the transition temperature in a given test, to be estimated. The results suggest that, for the type of notch-bend specimen considered, the variation of the nominal rate of stress within moderate limits (say up to 10^7 p.s.i./sec.) does not have a strong effect on the transition temperature. It is only at rates of loading approaching impact conditions that a marked change in transition temperature seems likely. It is suspected, however, that under these conditions extra complications may be introduced into the stressing system other than pure rate effect.

The effect of specimen size is shown to be considerable, but the experimental evidence points towards a possible limit to this effect in the larger sizes. It has been shown that the stiffness of the testing system, originally thought to be of minor importance, may have a strong effect on the transition point, whether or not ductile instability is present. It is suggested that should this programme be continued, in order to complete some of the other links indicated by the vacancies in Tables 21 to 23, then efforts should be made to ensure that any variation in machine stiffness is minimised.

A fracture criterion of the Hendrickson, Wood and Clark type appears to be valid, but it is shown that the transition

point predicted in notch-bend specimens is not the 20 ft. lb. transition temperature which they suggested. The transition point is in fact at the fully brittle end of the curve, and corresponds with a discontinuity in the fracture load against temperature curve. The discontinuity itself has been shown, by means of Schlieren photographs of the deformation patterns, to coincide with the onset of fracture at general yield. This part of the investigation could only be performed on steel 'P' and the valuable confirmation of these findings might be obtained if the tests on steel 'V' were extended into lower temperature regions. It is further suggested that any future specimens, which are intended for tests in this low temperature region, should be polished, at least on one side.

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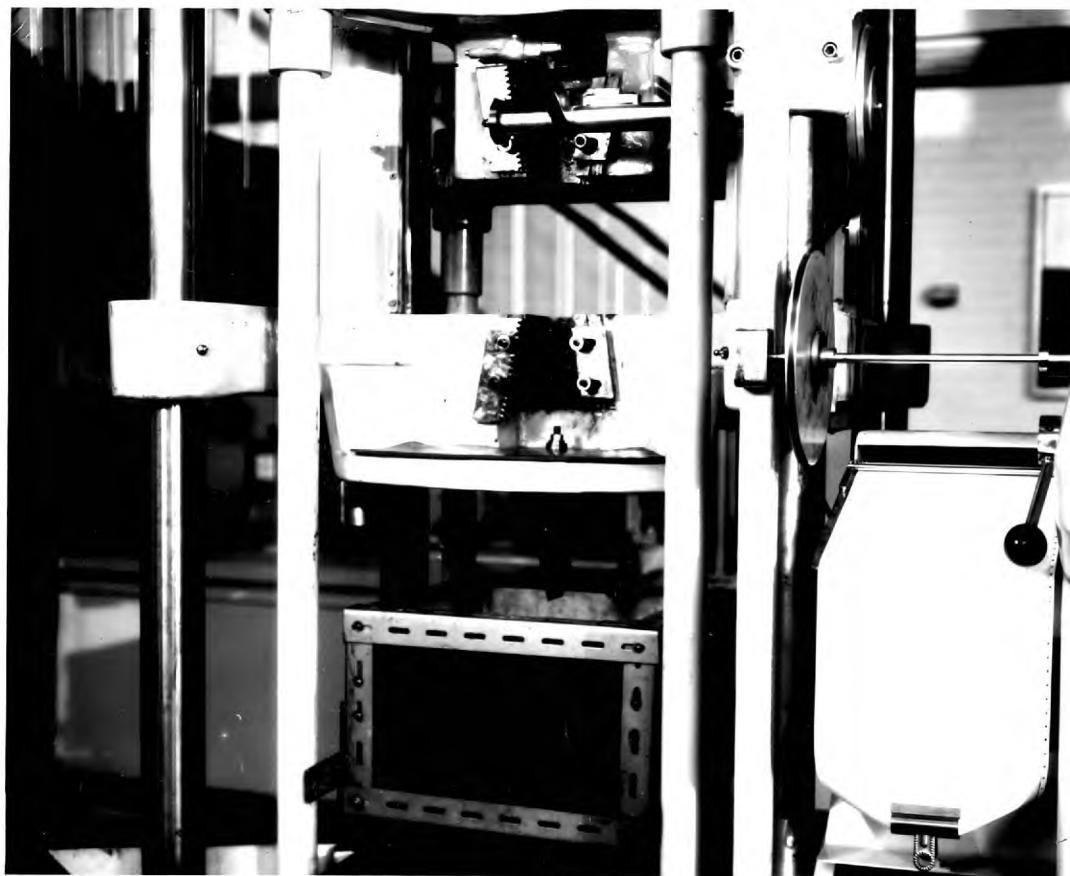
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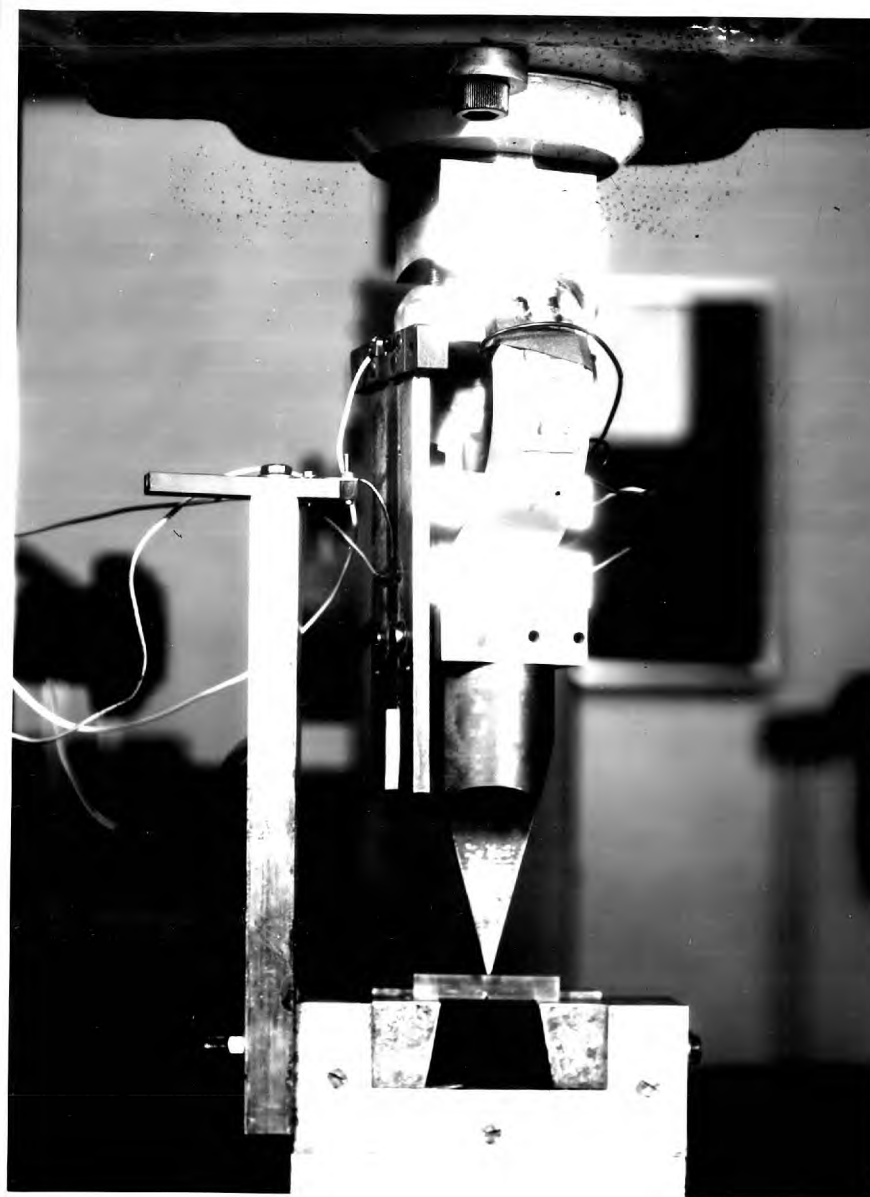
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(PLATE I)



THE OVERALL LOADING SYSTEM IN THE SLOW NOTCH-BEND TESTS
SHOWING THE AUTOGRAPHIC RECORDER



STANDARD SIZE SLOW NOTCH-BEND JIG
SHOWING THE RING-TYPE LOAD-CELL AND THE
CARBON STRIP DEFLECTION PICK-UP

(PLATE 2)



CHARPY IMPACT MACHINE - PENDULUM HEAD

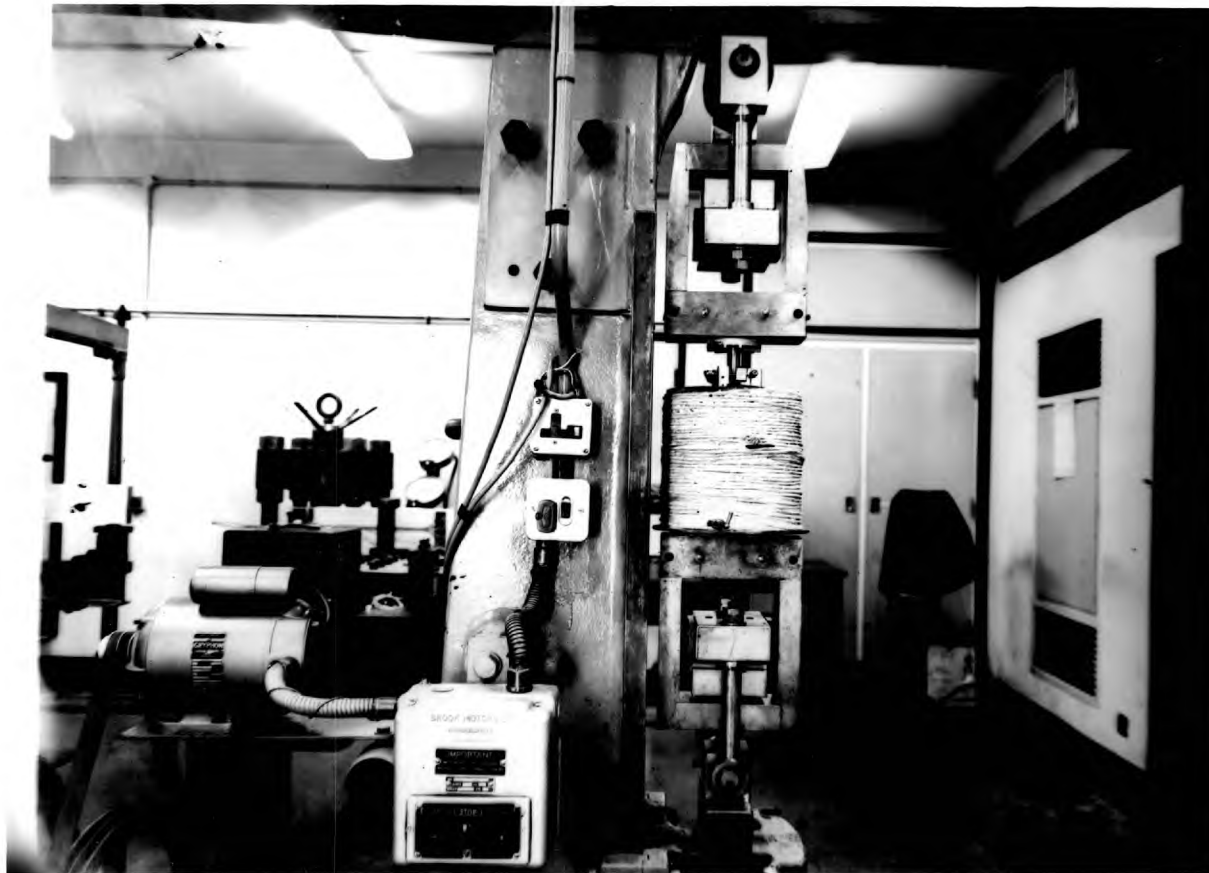
(PLATE 3)

(PLATE 4)



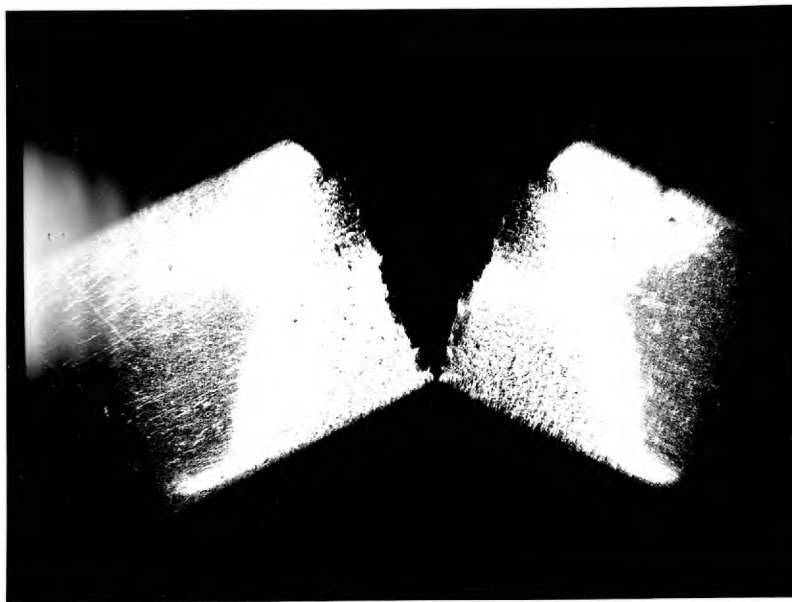
SPRING-LOADED CONTACTS AND SOLENOIDS ON THE IMPACT MACHINE

(PLATE 5)

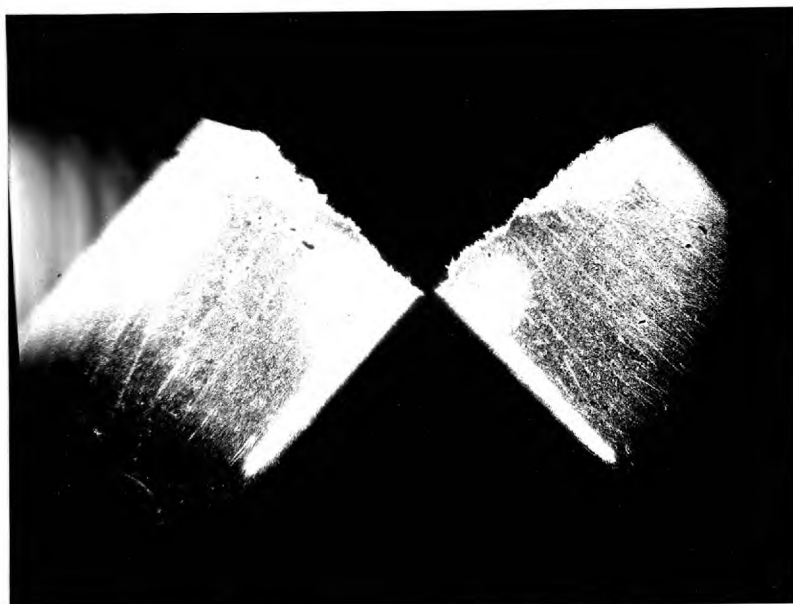


SHACKLE ASSEMBLY FOR THE YIELD STRESS TESTS

THE DEFORMATION PATTERNS OBTAINED
ON STANDARD SIZE SLOW NOTCH-BEND SPECIMENS
PHOTOGRAPHED USING THE SCHLIEREN TECHNIQUE

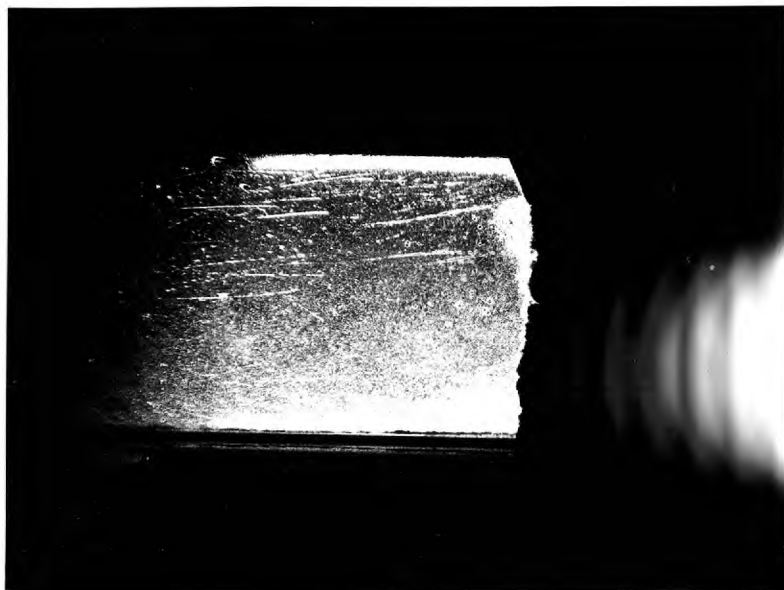


(a) P.301 - 82°C. 98% crystalline

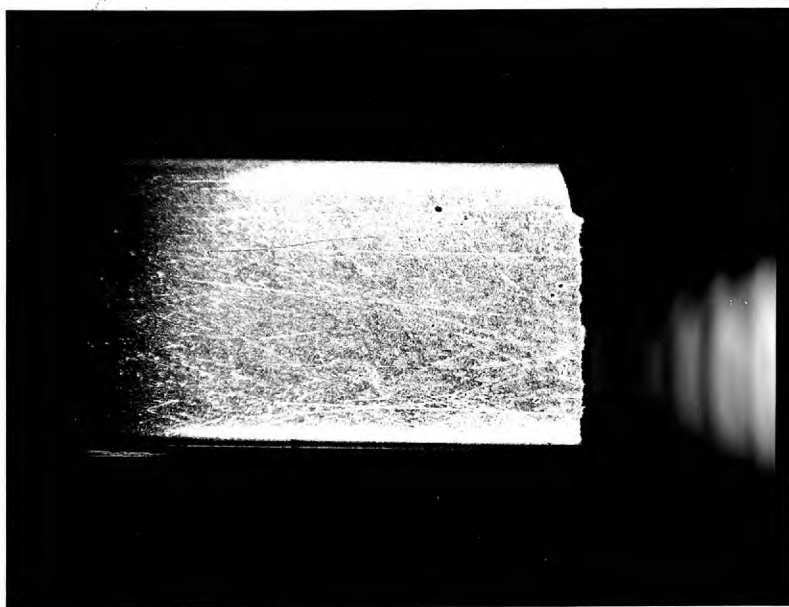


(b) P.302 - 106°C. 99% crystalline

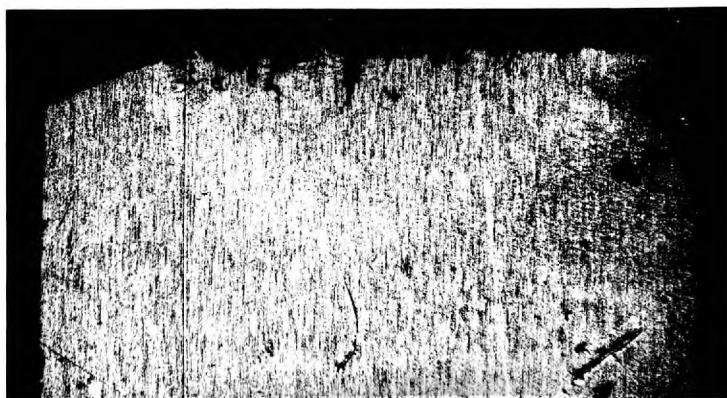
(PLATE 6)
(continued)



(c) P.306 - 118°C. 100% crystalline



(d) P.307 - 196°C. 100% crystalline

P 10 -85°C .P 11 -99°C .P 12 -125°C .

THE DEFORMATION PATTERN IN SOME
DOUBLE SIZE SPECIMENS

(PLATE 7)



FURTHER SCHLIEREN PHOTOGRAPH
STANDARD SIZE NOTCH TENSILE SPECIMEN
P.12 -115°C . 100° crystalline

(PLATE 8)



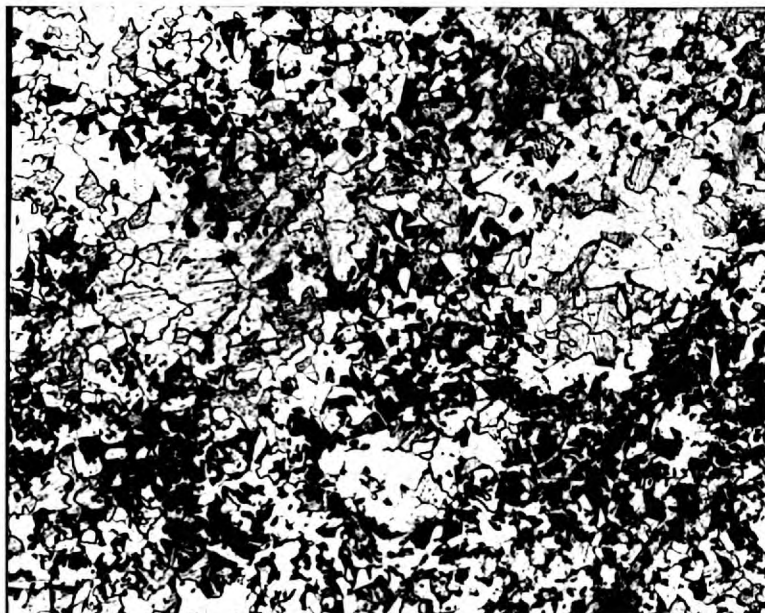
(a) Top



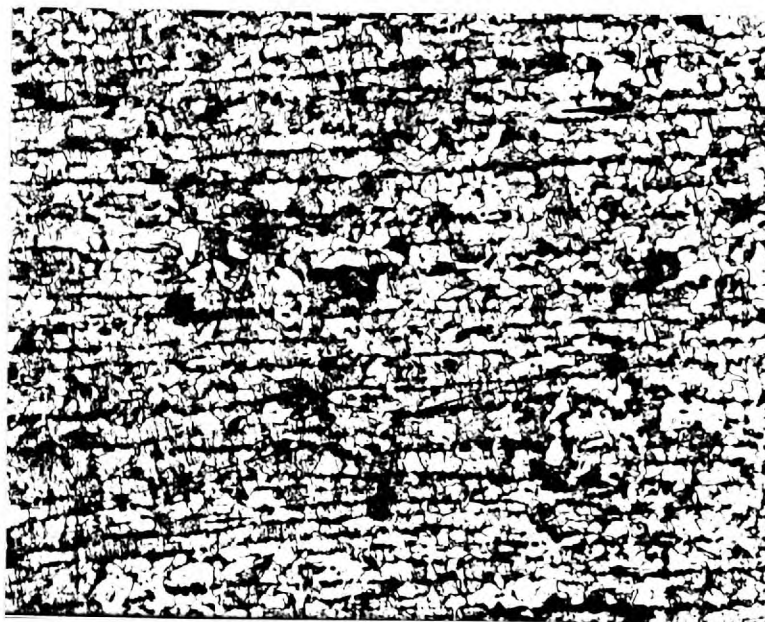
(b) End

MICROSTRUCTURE OF STEEL 'P'
x 100

(PLATE 9)



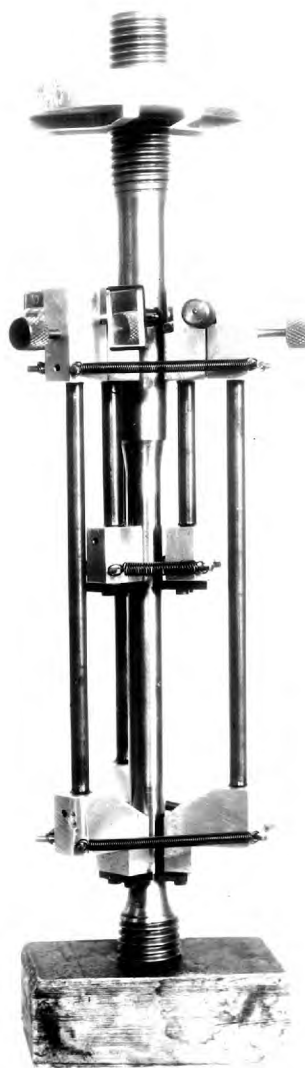
(a) Top



(b) End

MICROSTRUCTURE OF STEEL 'V'
x 100

(PLATE 10)



TRIPLE EXTENSOMETER ASSEMBLY

(PLATE 11)

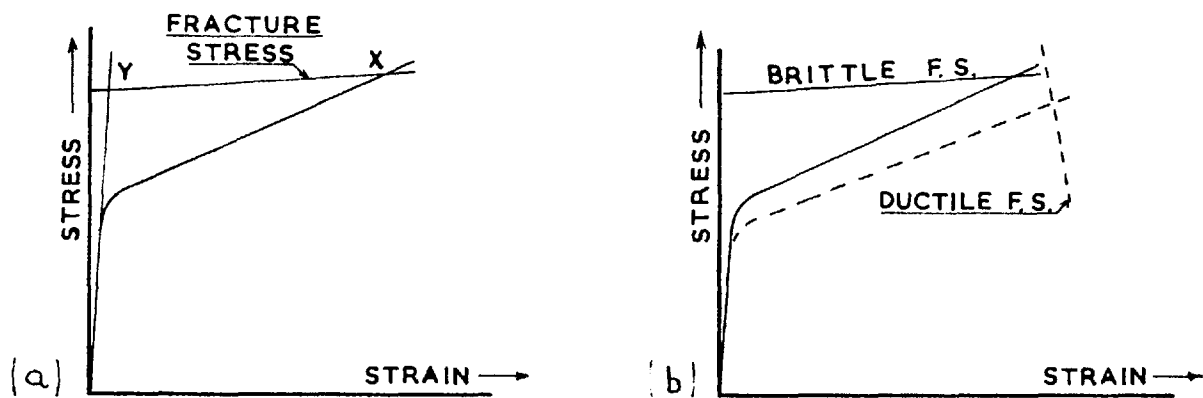


FIG. 1 FLOW STRESS FRACTURE STRESS RELATIONSHIPS

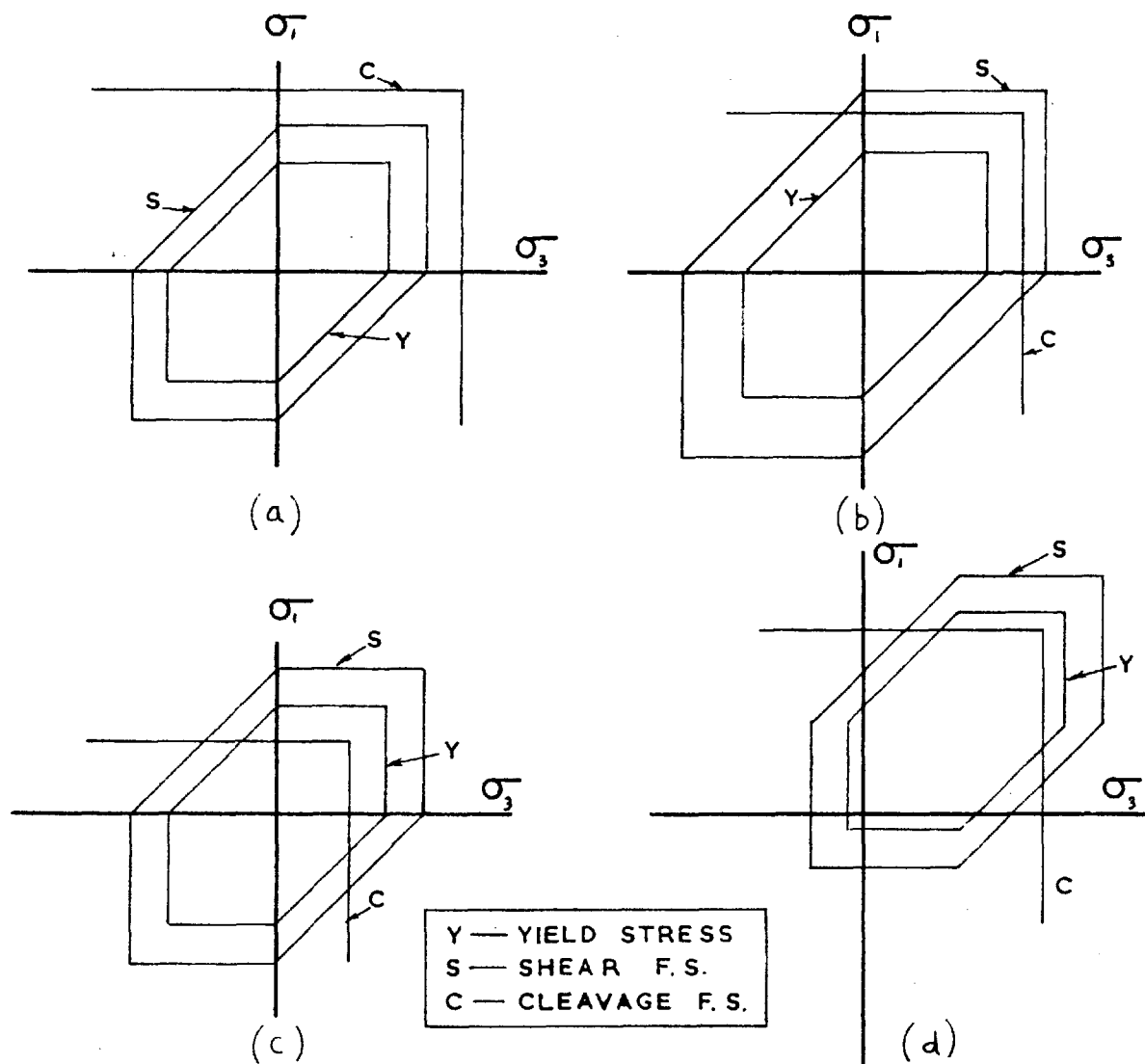
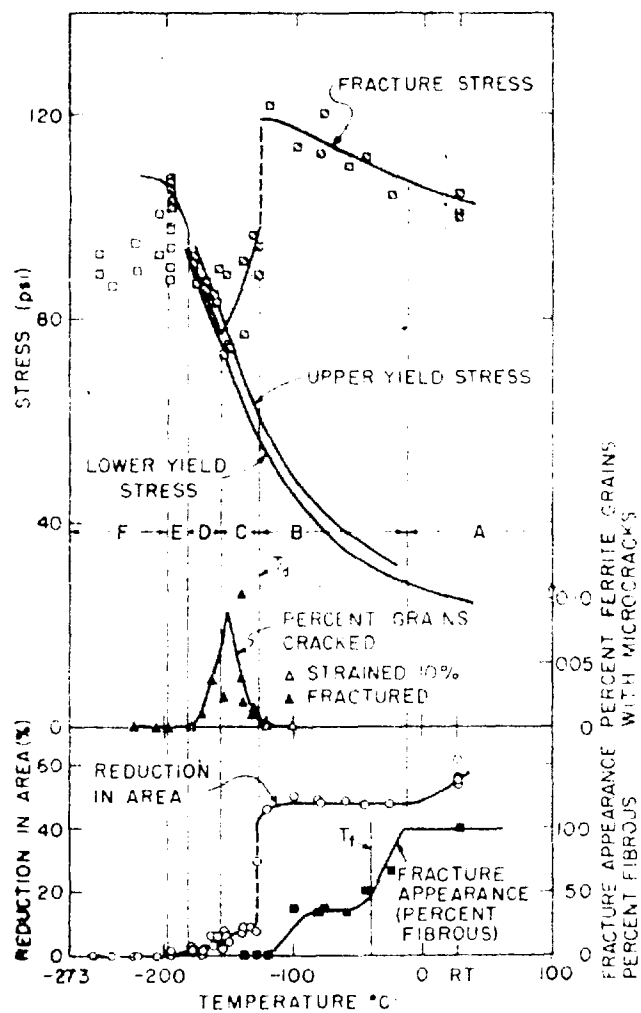
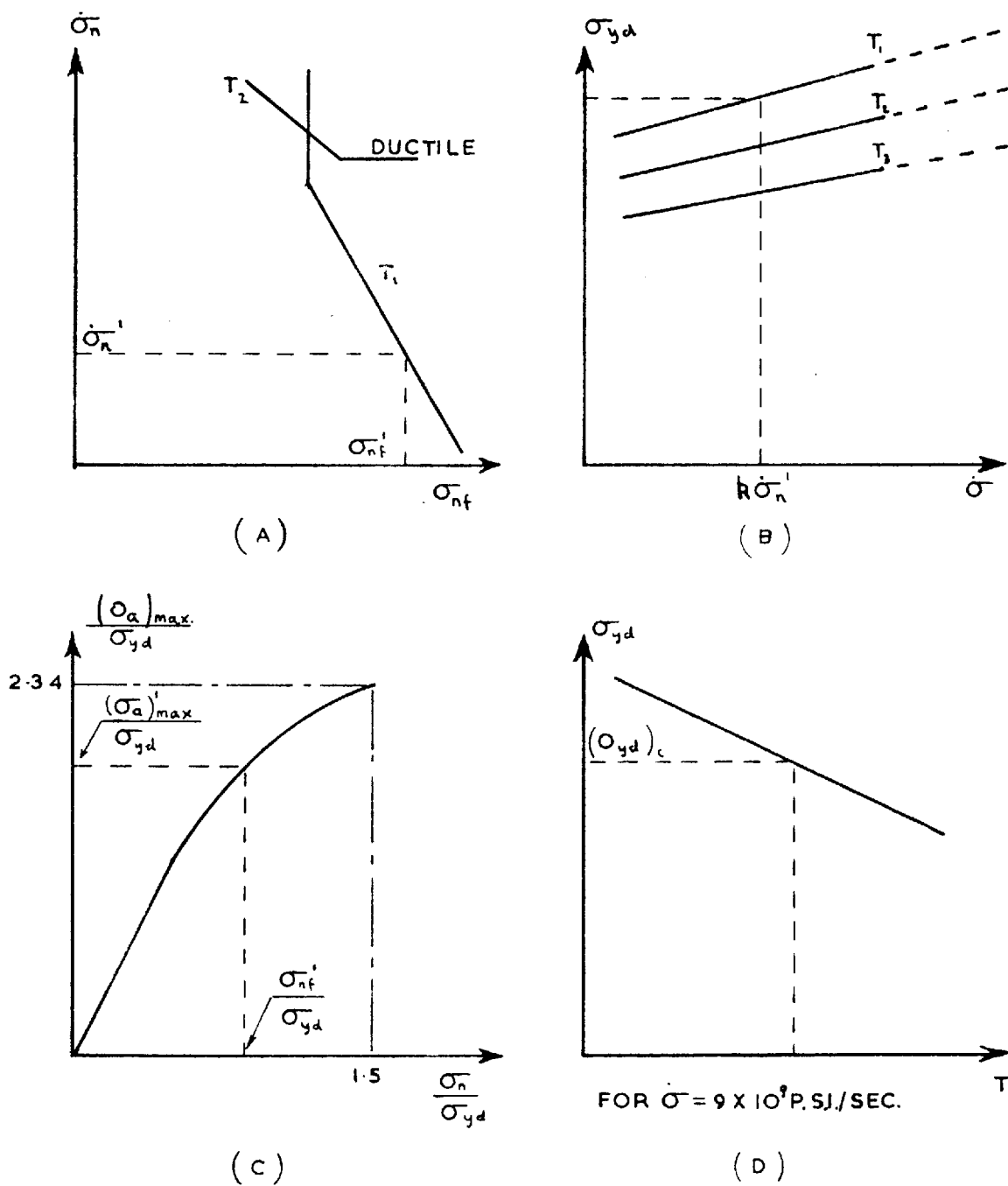


FIG. 2 SOME POSSIBLE RELATIVE POSITIONS OF THE FLOW & FRACTURE SURFACES ON THE σ_1 - σ_3 PLANE.



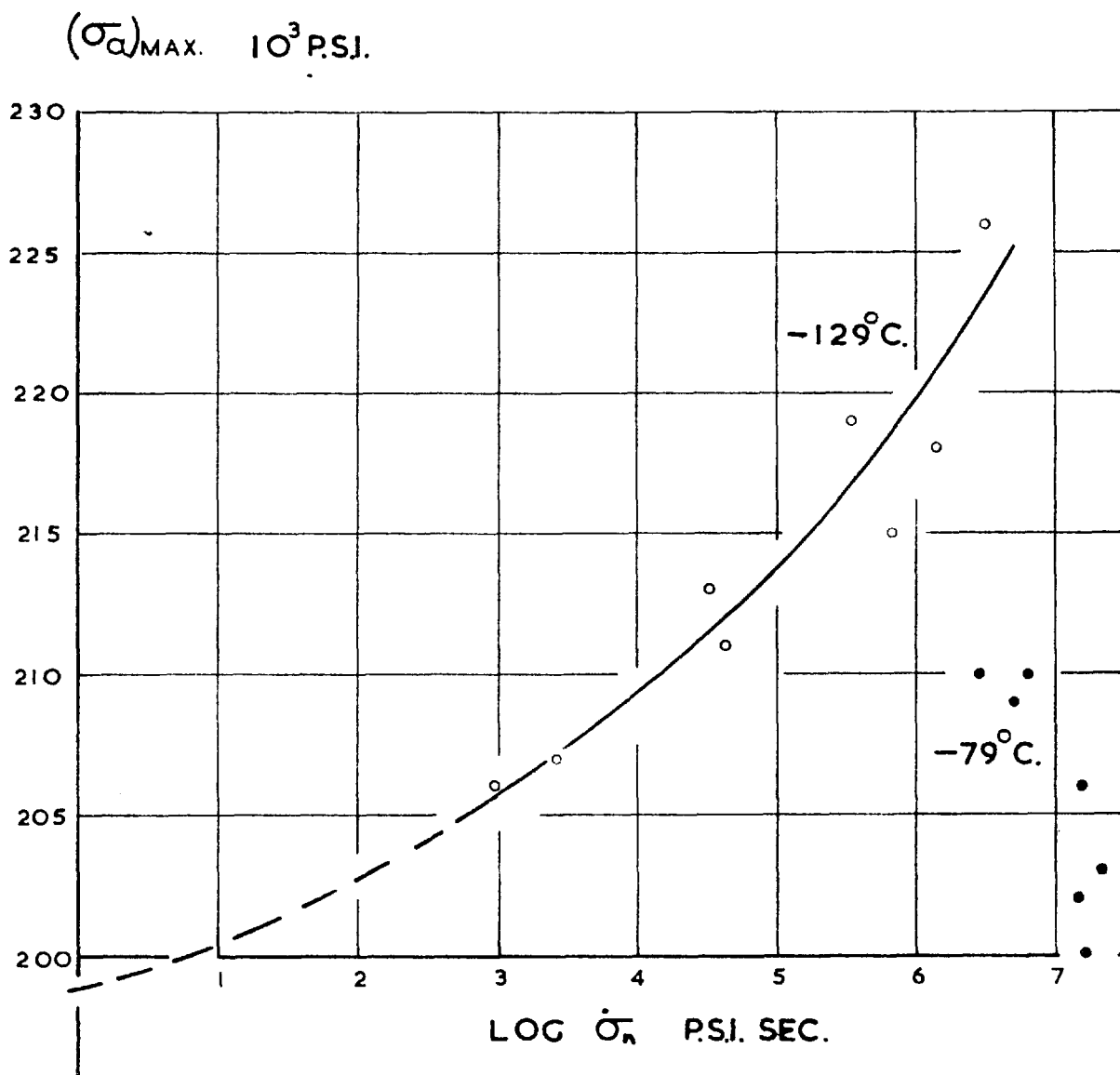
VARIATION OF TENSILE PROPERTIES WITH TEMP.
[HAHN ET AL]

FIG. 3



THE METHOD EMPLOYED BY HENDRICKSON CLARK & WOOD
TO CALCULATE THE CRITICAL FRACTURE STRESS (22)

FIG. 4



VARIATION OF THE CRITICAL FRACTURE STRESS
WITH THE APPLIED STRESS RATE [H,W&C]

FIG. 5

THE DEPENDENCE OF ' G_c ' ON THE POSITION
AND SHAPE OF THE ' R ' CURVE

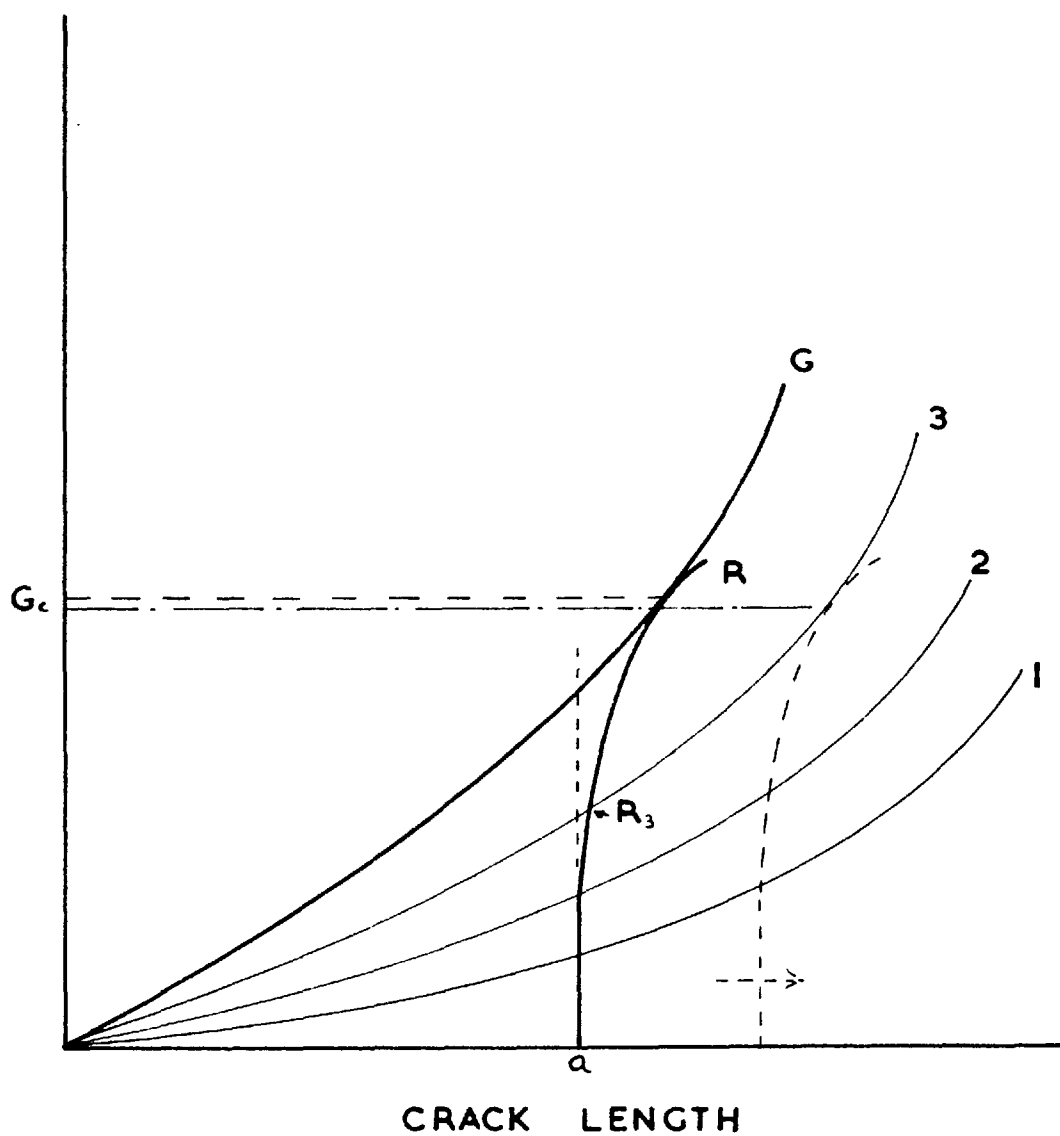
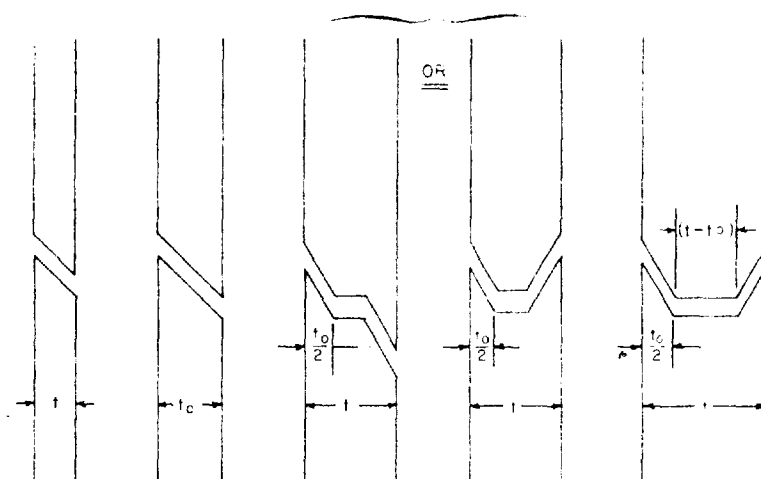


FIG. 6



Fracture Model for Analysis.

THE VARIATION OF SHEAR LIP AS SUGGESTED
by BLUHN

FIG. 8

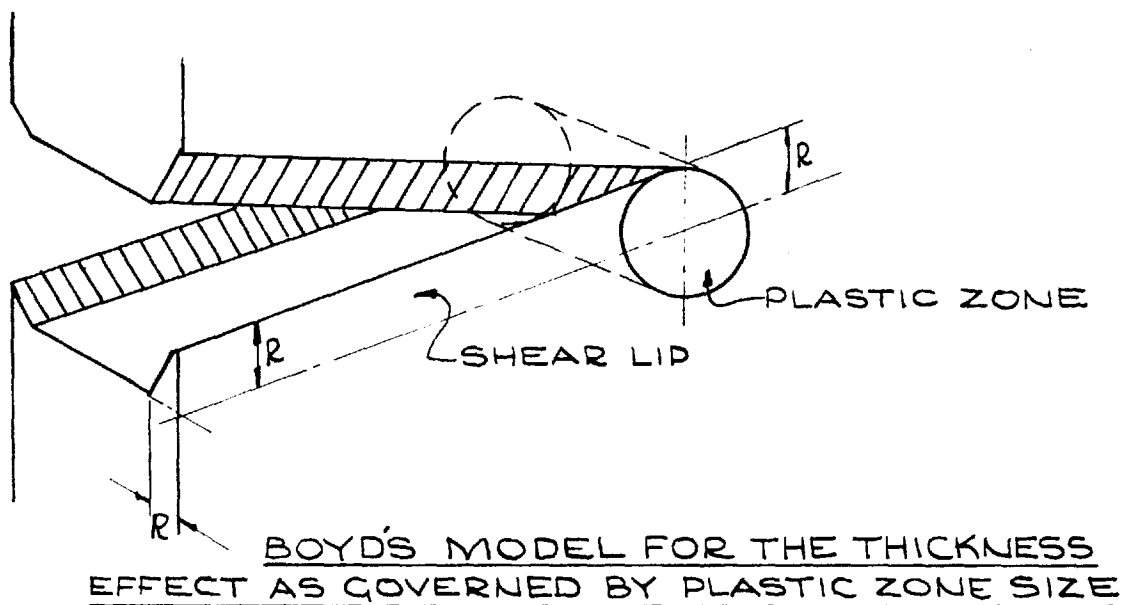
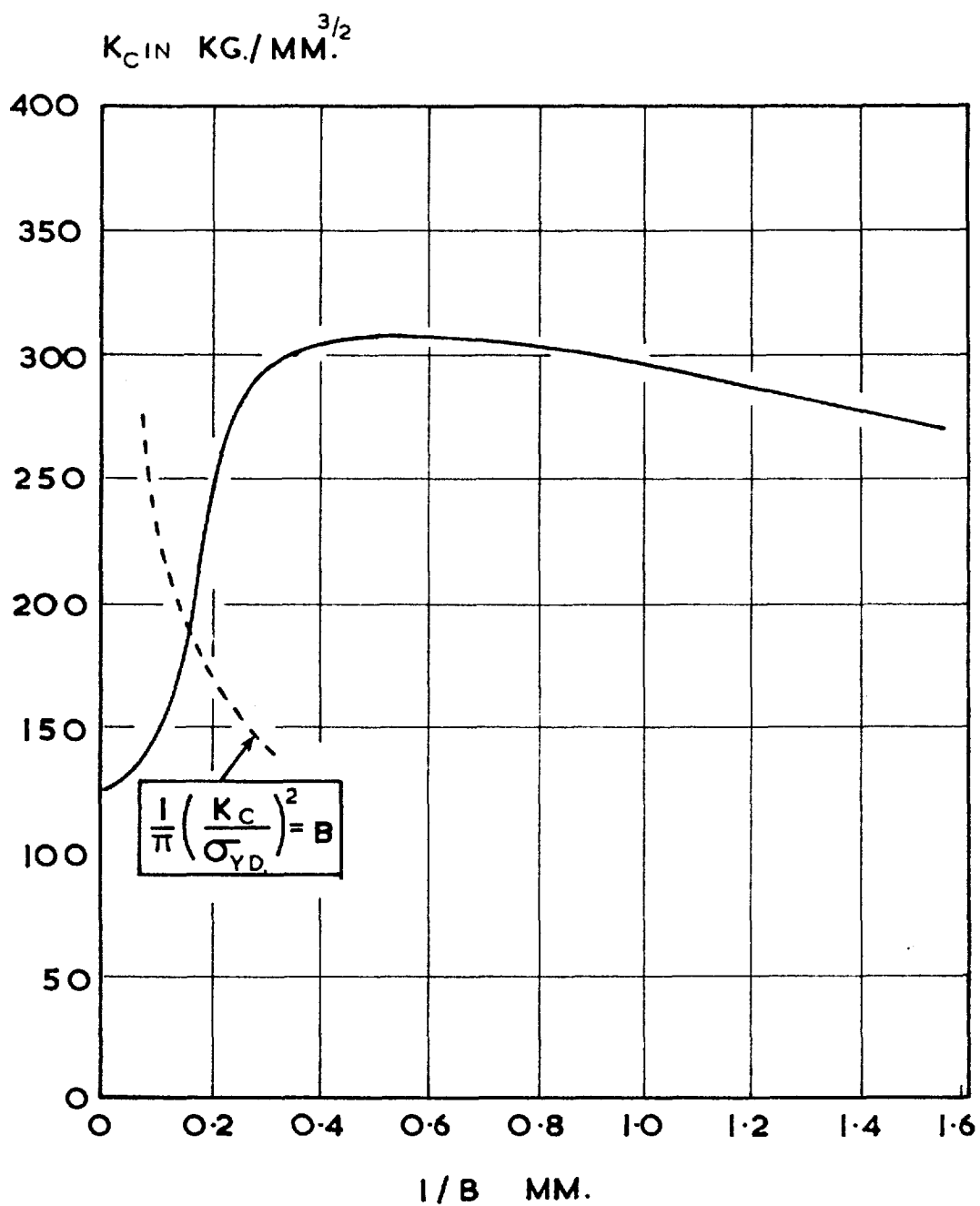


FIG.7



THE VARIATION OF K_C WITH PLATE THICKNESS
FOR ALUMINIUM

FIG.9

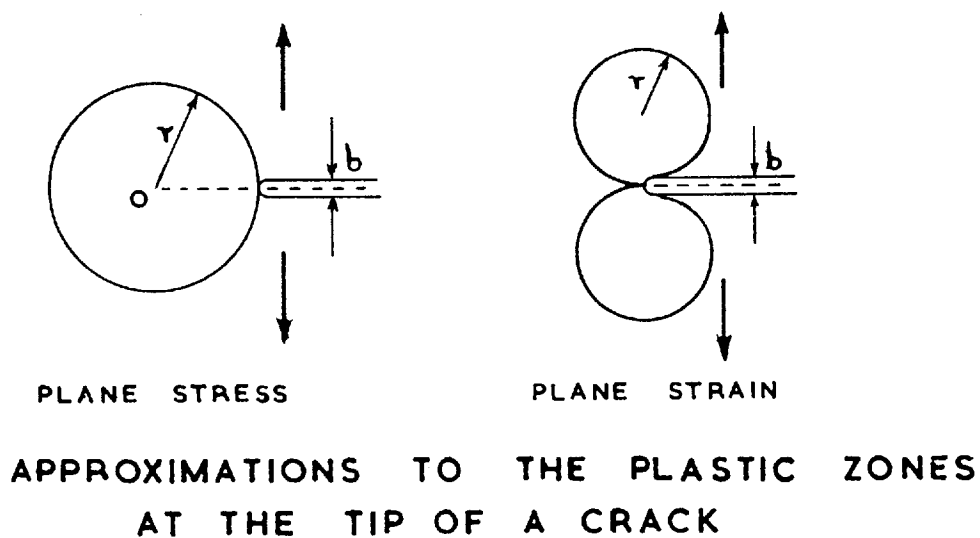
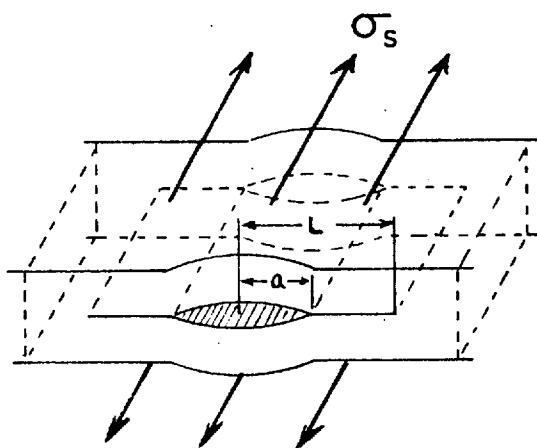
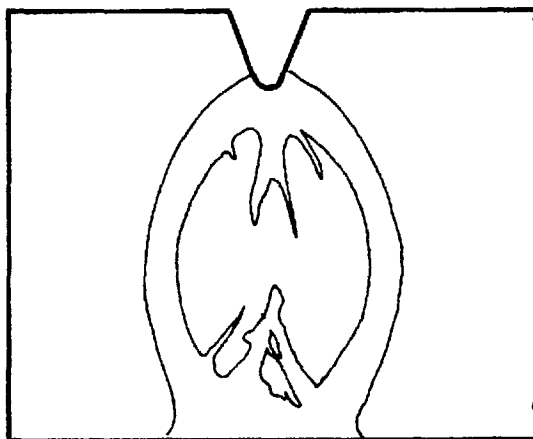


FIG. 10

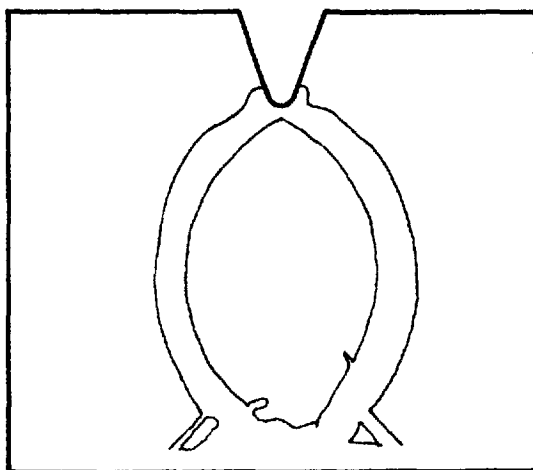


SPREAD OF PLASTIC ZONES FROM
A SHEARED SLIT

FIG. 11



(a)



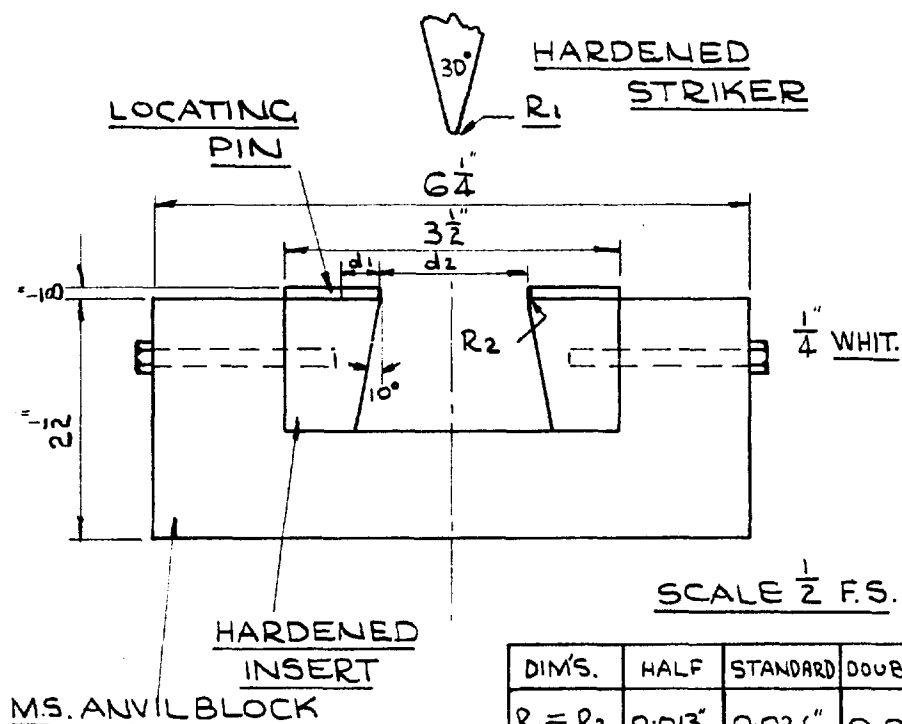
(b)

DEFORMATION PATTERNS OF 'V' NOTCH SPECIMENS
BENT UNDER CONCENTRATED LOAD.
FRYS. ETCH.

(a) SURFACE

(b) CENTRE

FIG. 12



DIM'S.	HALF	STANDARD	DOUBLE
$R_1 = R_2$	0.013"	0.026"	0.052"
d_1	5 mm	10 mm	20 mm
d_2	20 mm	40 mm	80 mm

NOTE: DOUBLE SIZE INSERTS
REQ. A LARGER ANVIL
BLOCK

SLOW-BEND JIG DIMENSIONS

FIG. 13

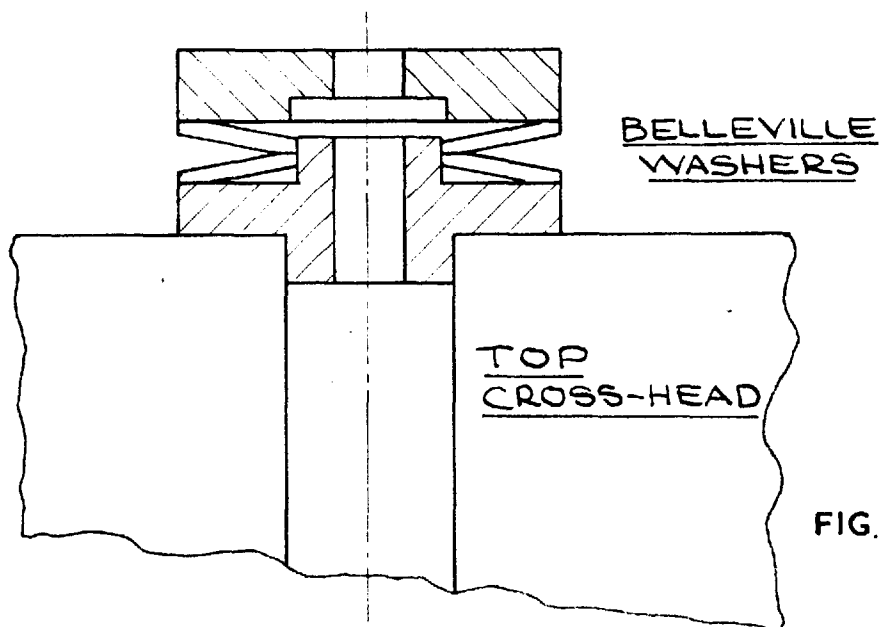
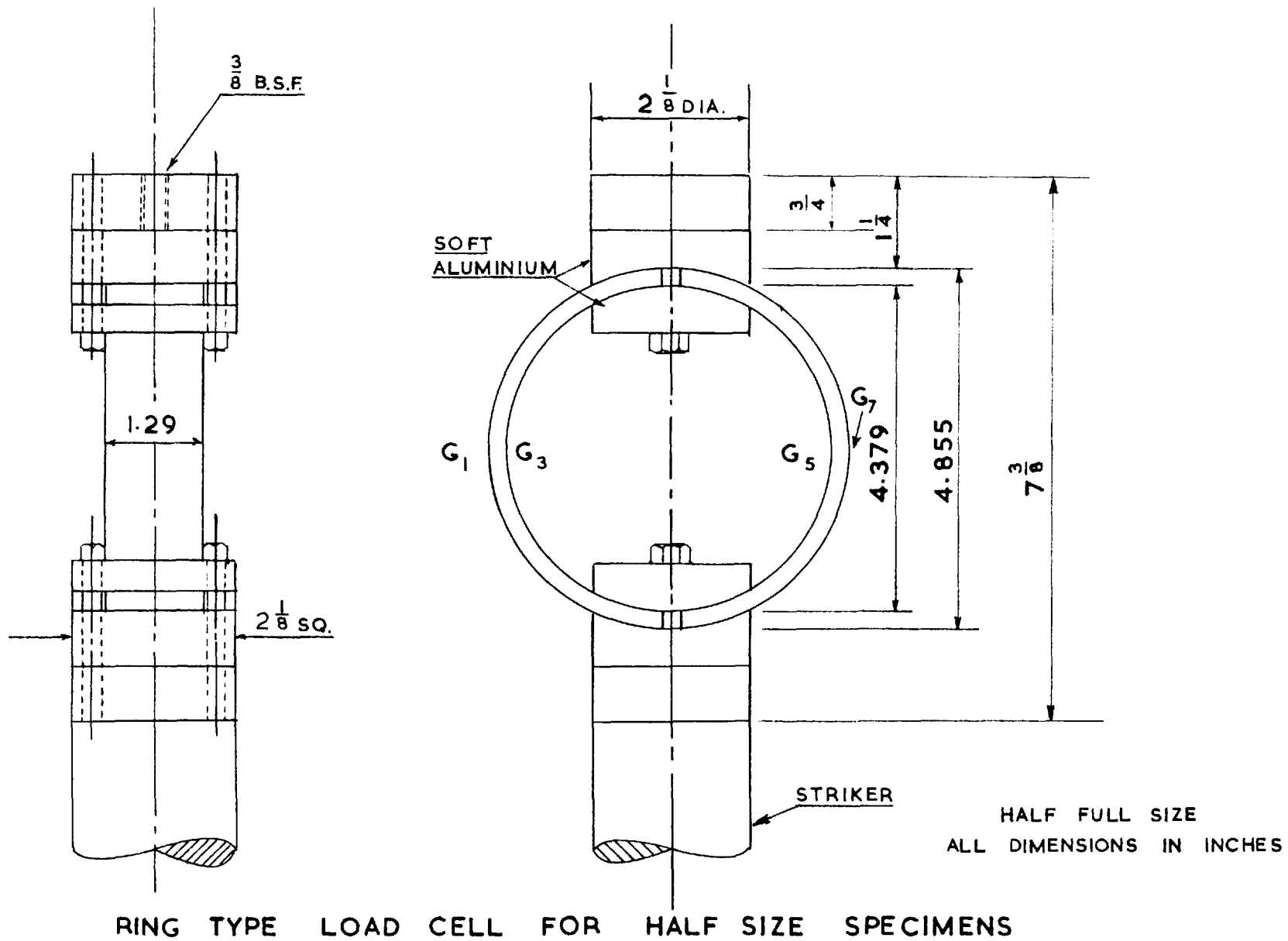


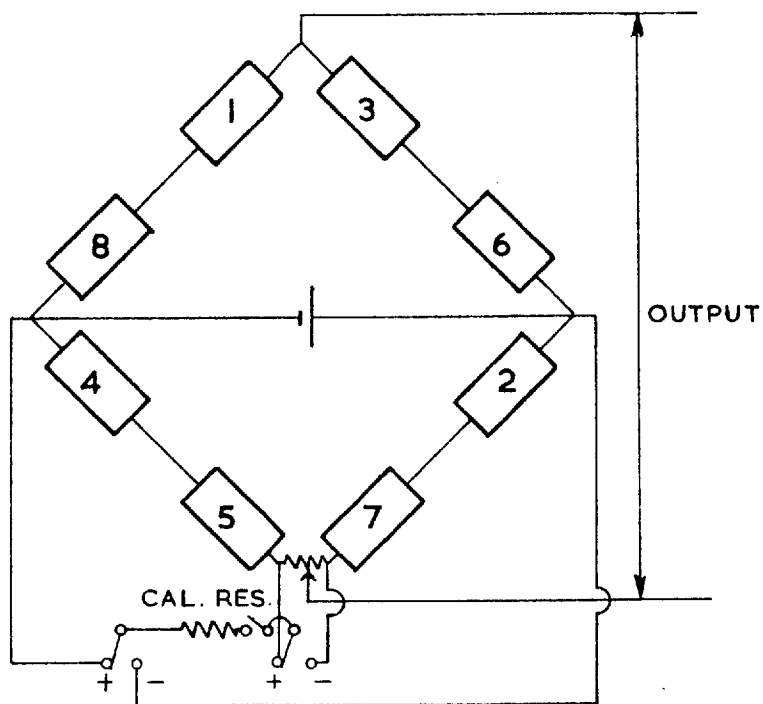
FIG. 14

METHOD OF PRE-LOADING THE
CYLINDRICAL LOAD CELL

SCALE F.S.

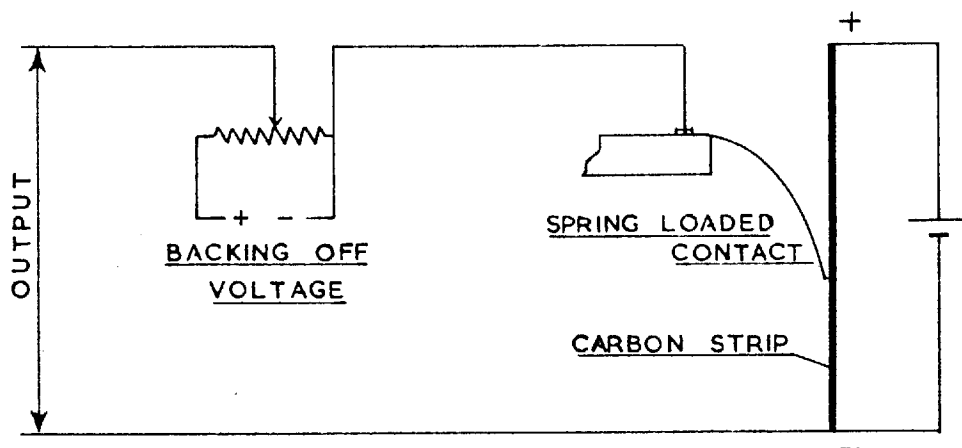
FIG. 15





BRIDGE CIRCUIT FOR THE RING TYPE
LOAD CELLS

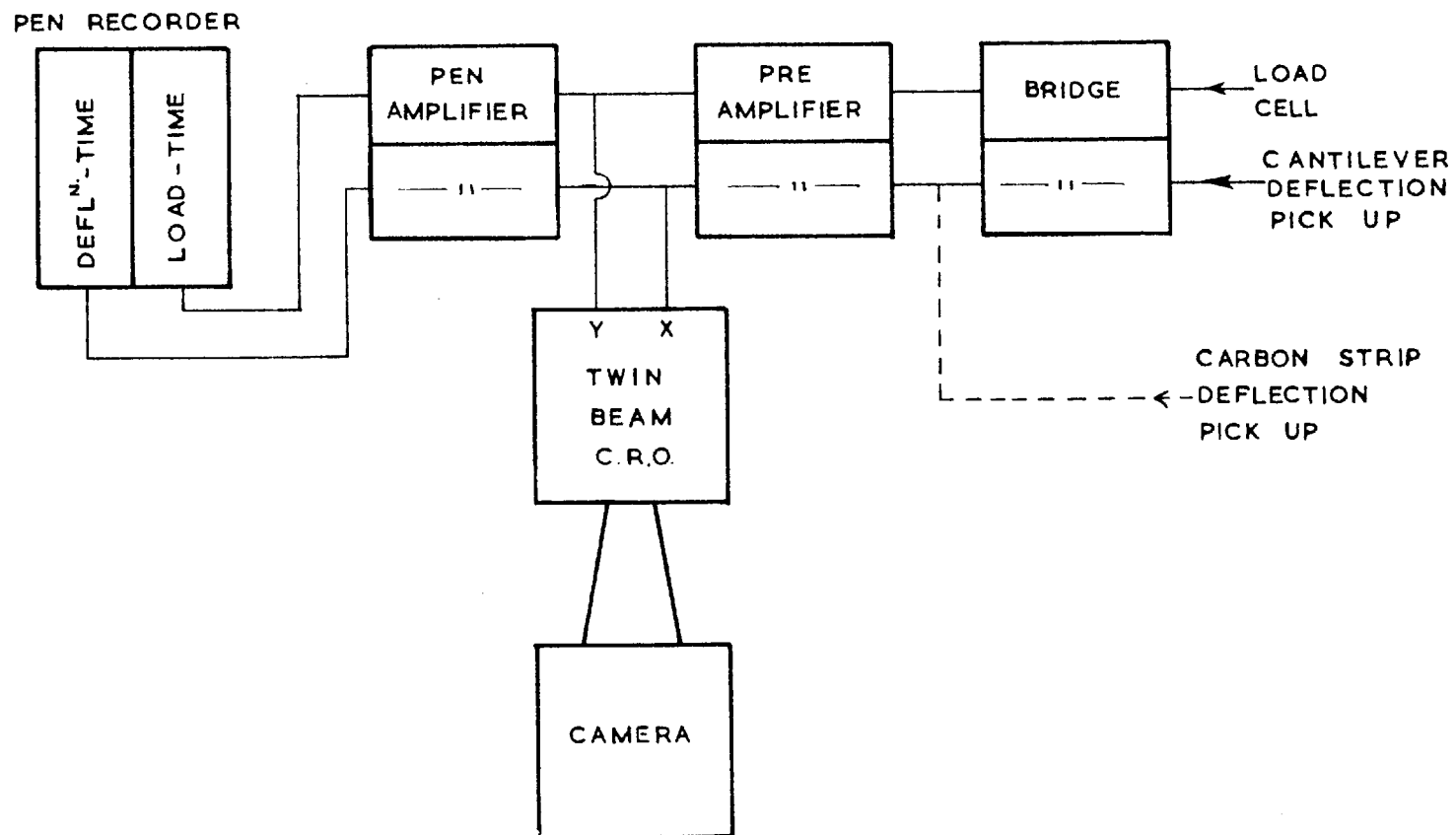
FIG. 16



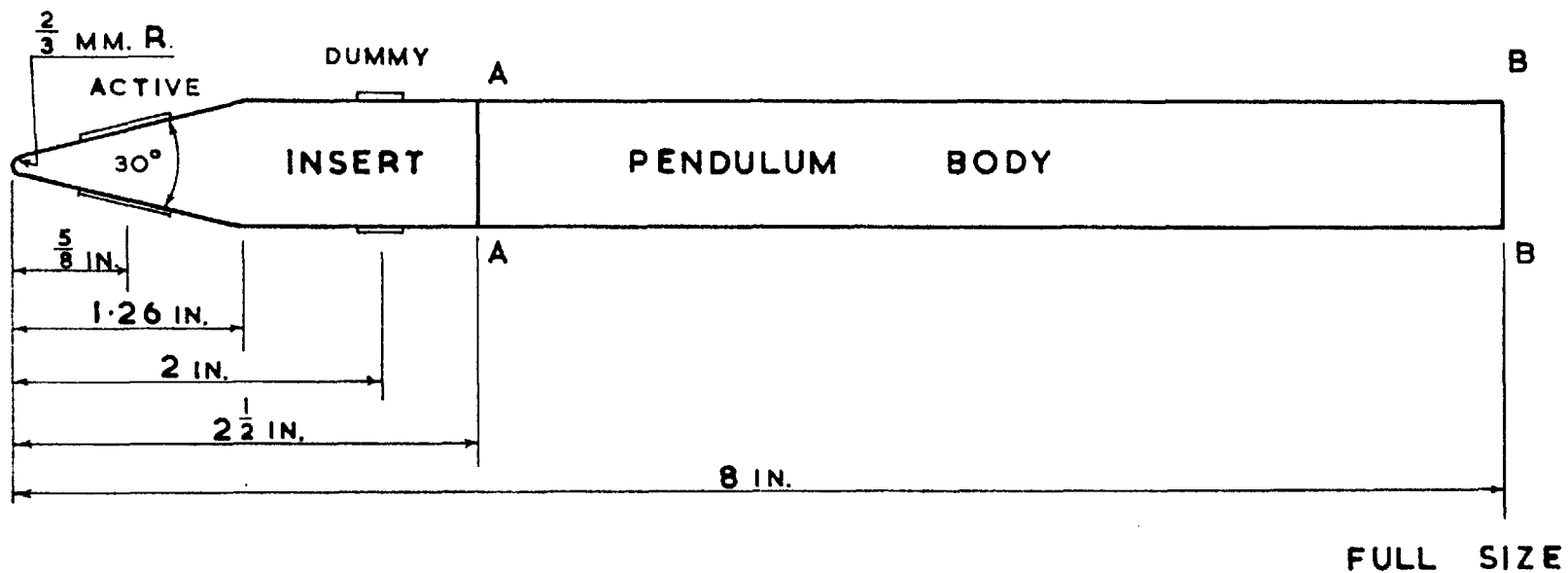
CIRCUIT FOR THE CARBON STRIP
DEFLECTION PICK UP

FIG. 17

FIG. 18



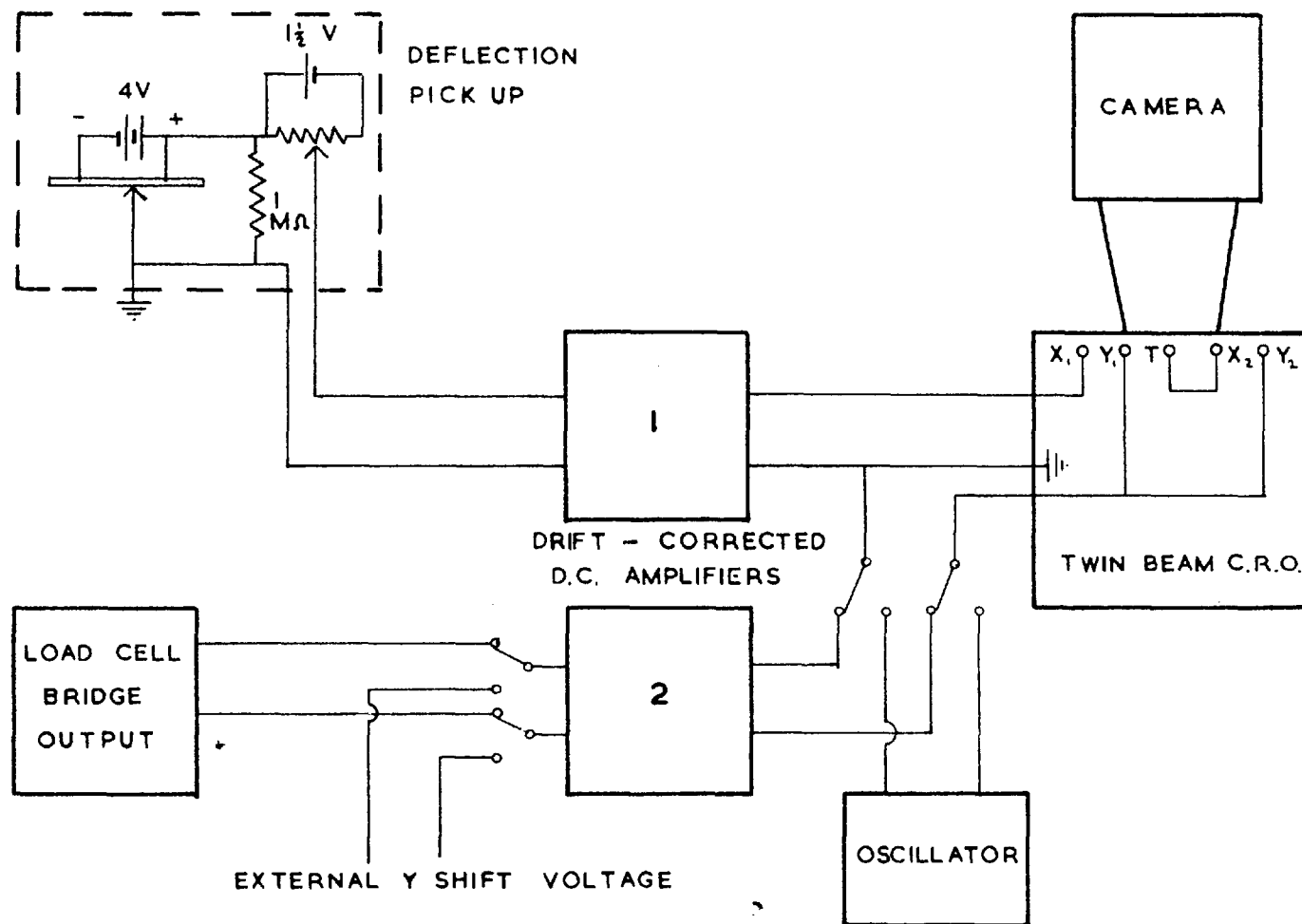
BLOCK DIAGRAM OF THE AMPLIFYING AND RECORDING SYSTEM IN THE SLOW NOTCH BEND TESTS



PLAN CROSS-SECTION OF THE CHARPY MACHINE STRIKER

FIG. 19

FIG 20



AMPLIFYING AND RECORDING GEAR FOR THE INSTRUMENTED CHARPY IMPACT TEST

SHACKLE ARRANGEMENT FOR
NOTCH-TENSILE TESTS

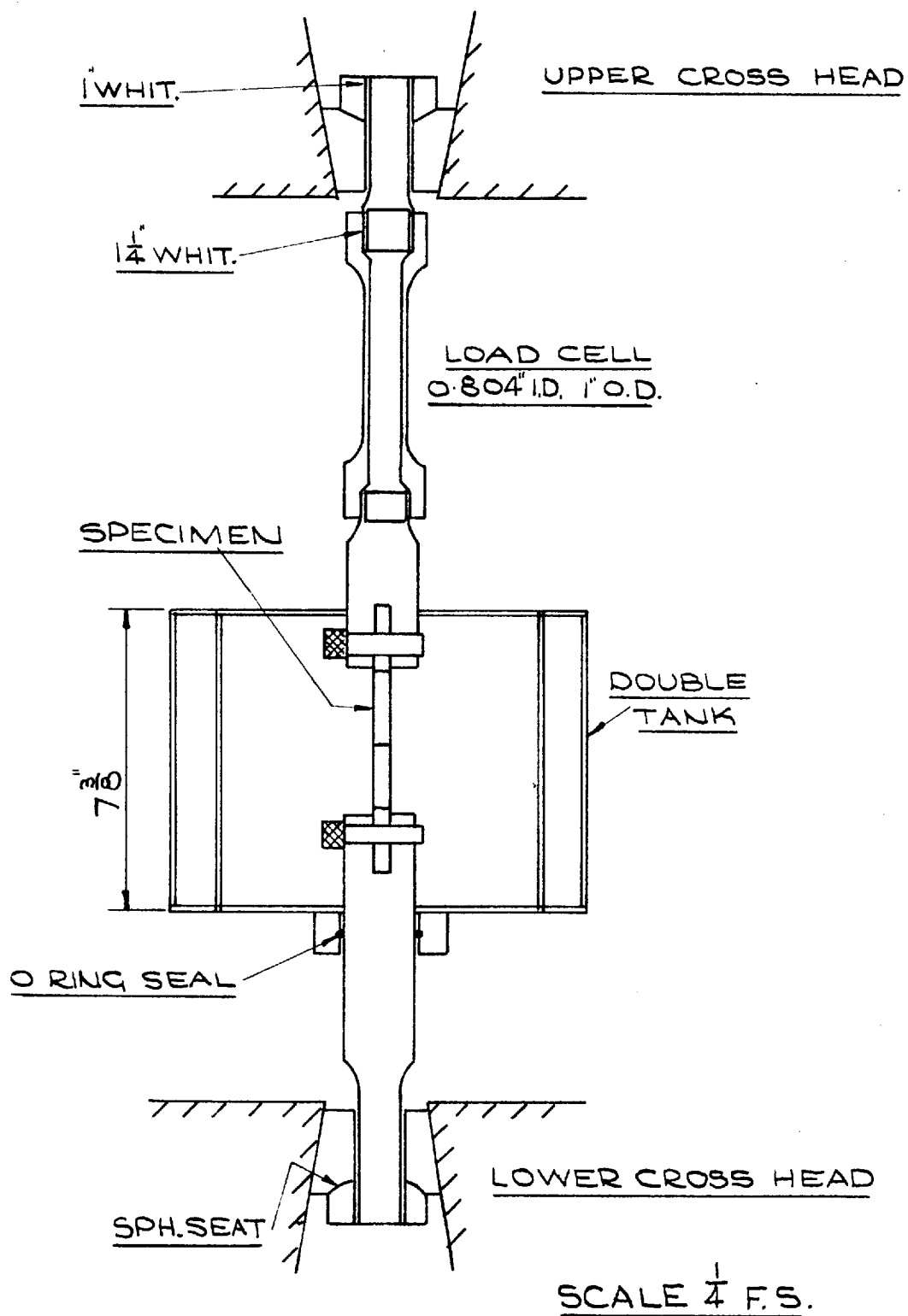


FIG. 21

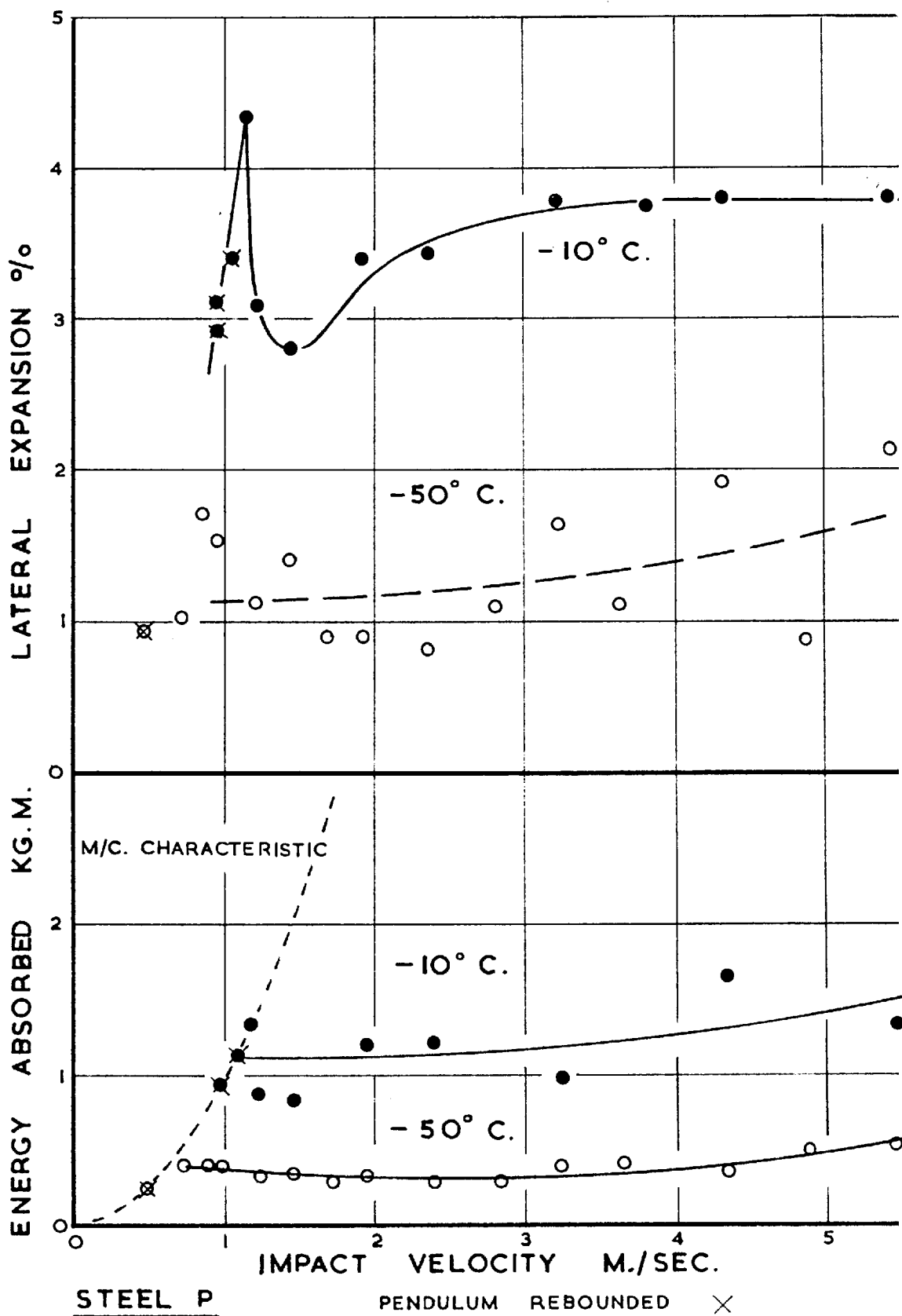
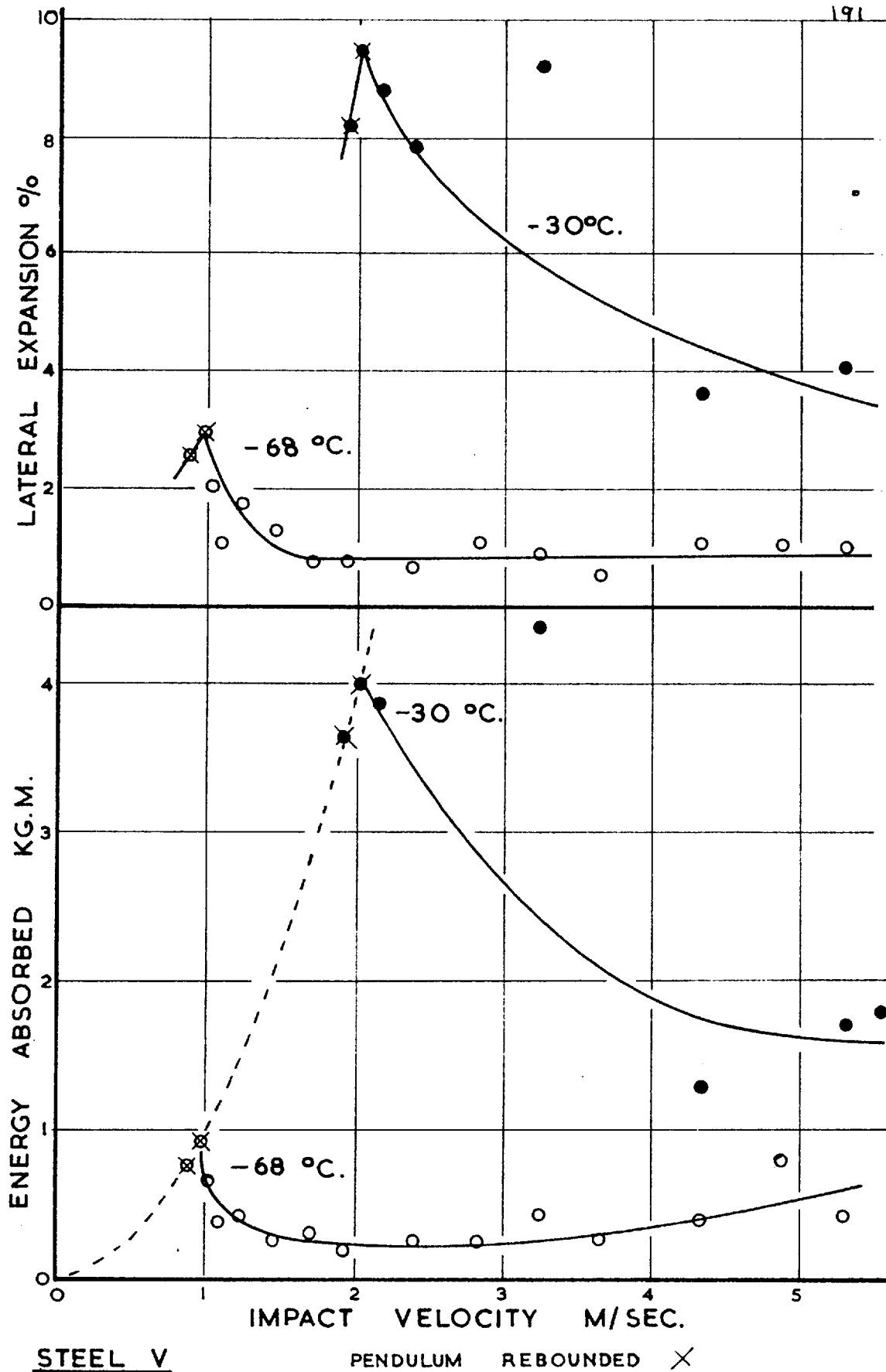


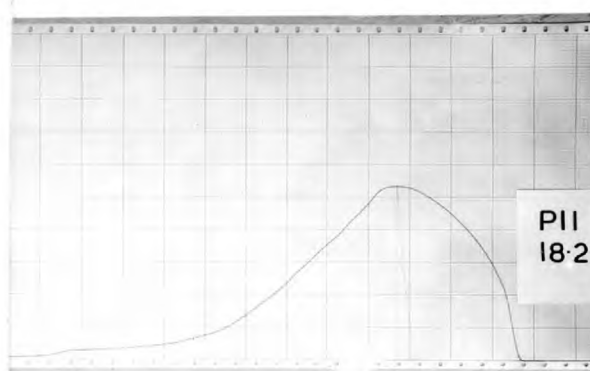
FIG 22 EFFECT OF VARYING IMPACT VELOCITY



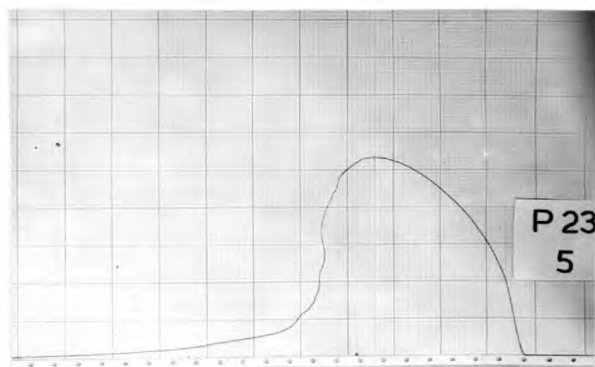
STEEL V

PENDULUM REBOUNDED X

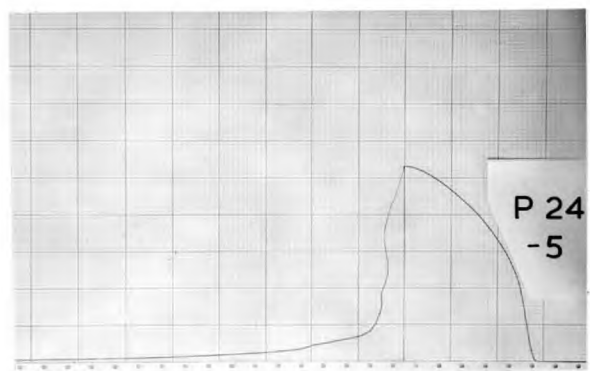
FIG.23 EFFECT OF VARYING IMPACT VELOCITY



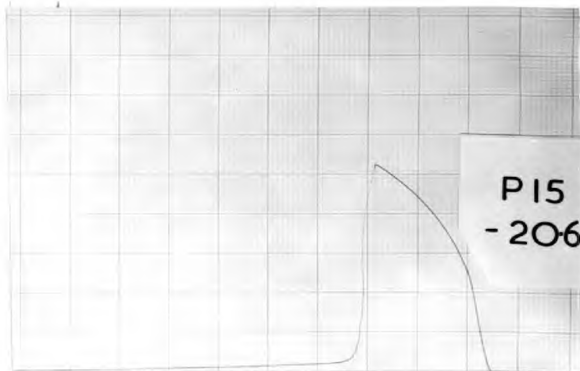
1 %



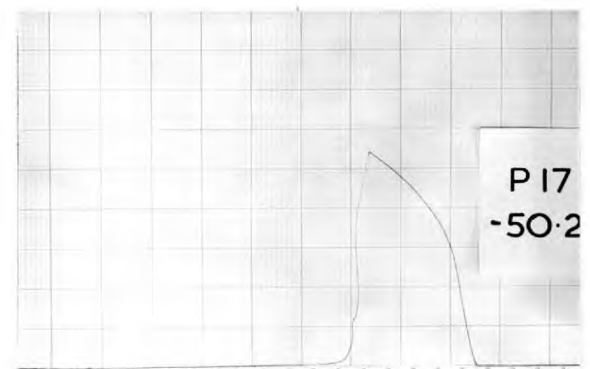
45 %



65 %



85 %



95 %

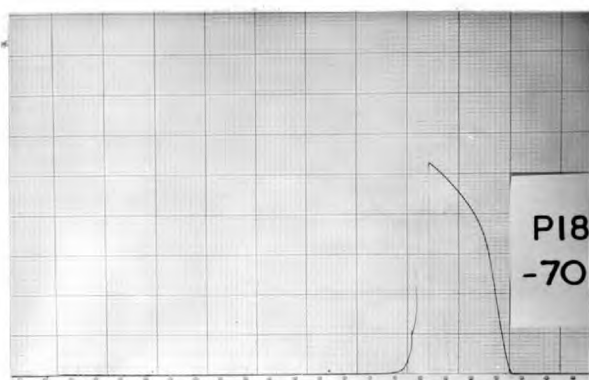
TESTING TEMPERATURES ARE
GIVEN ON THE DIAGRAMS IN °C

% CRYSTALLINITIES ARE NOTED
BELOW.

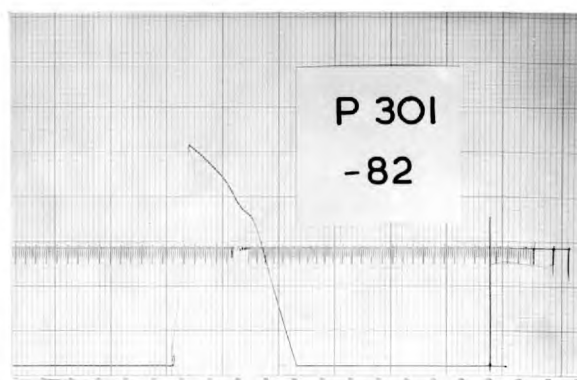
SCALE: *large*
 ↑ 1 GRAPH DIV. = 0.2 TON
 ← " " = 0.05 IN.

FIG. 24.

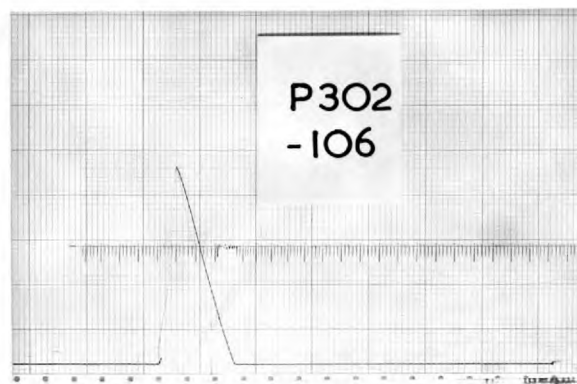
LOAD DEFLECTION DIAGRAMS FOR THE
SLOW NOTCH-BEND TESTS
(TO A REDUCED SCALE)



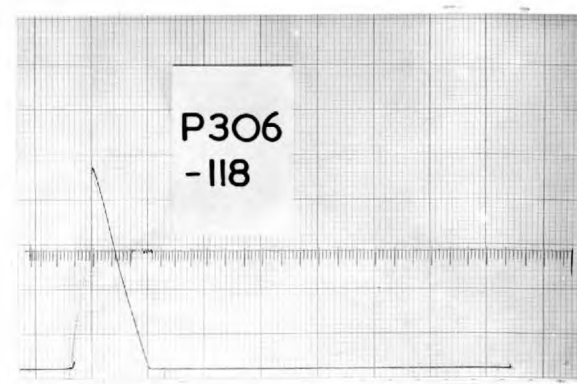
100 %



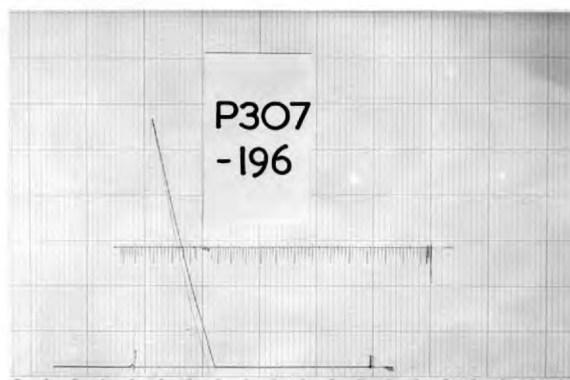
98 %



99 %



100 %



100 %

FIG 24 CONTINUED

FIG. 25

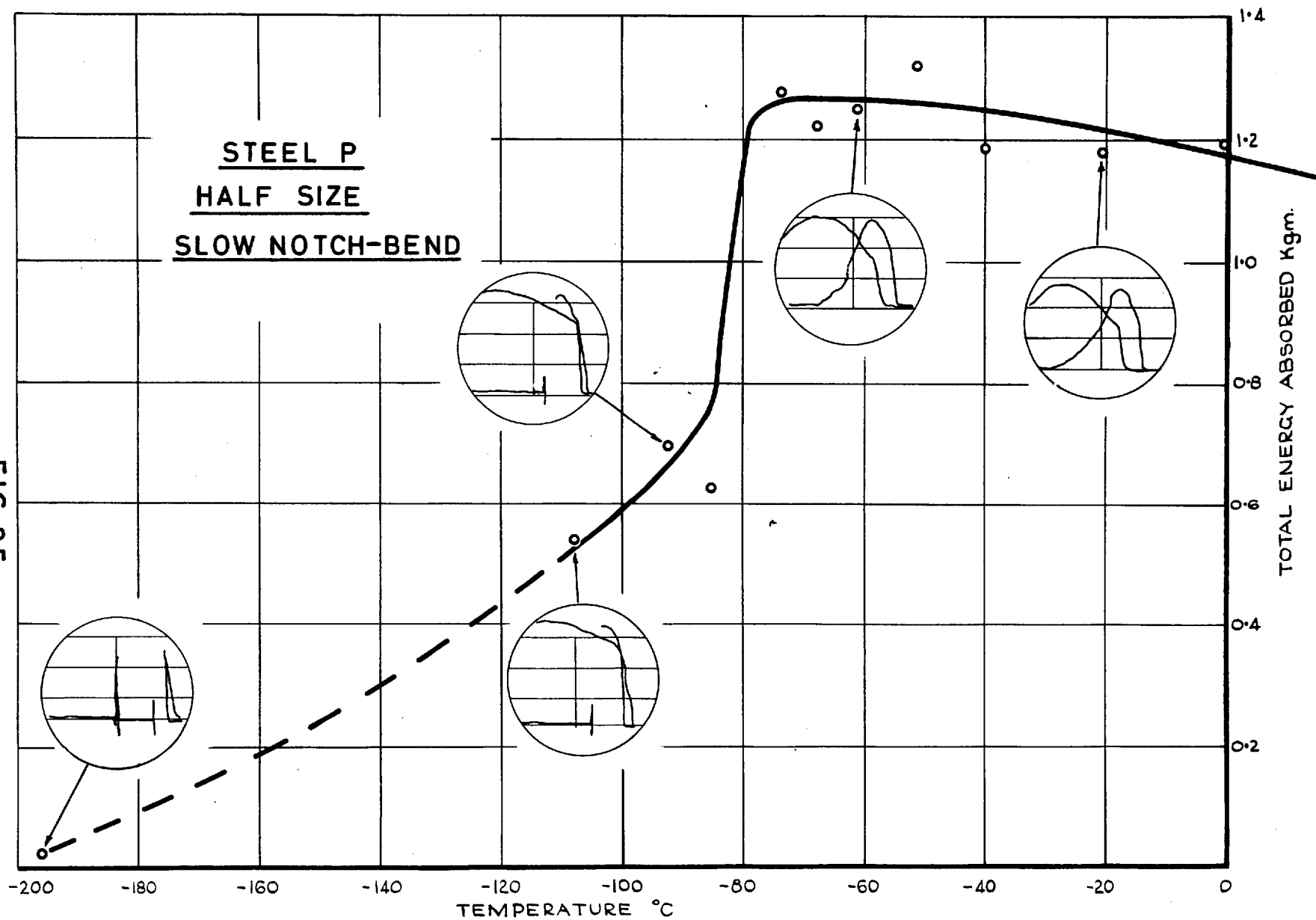
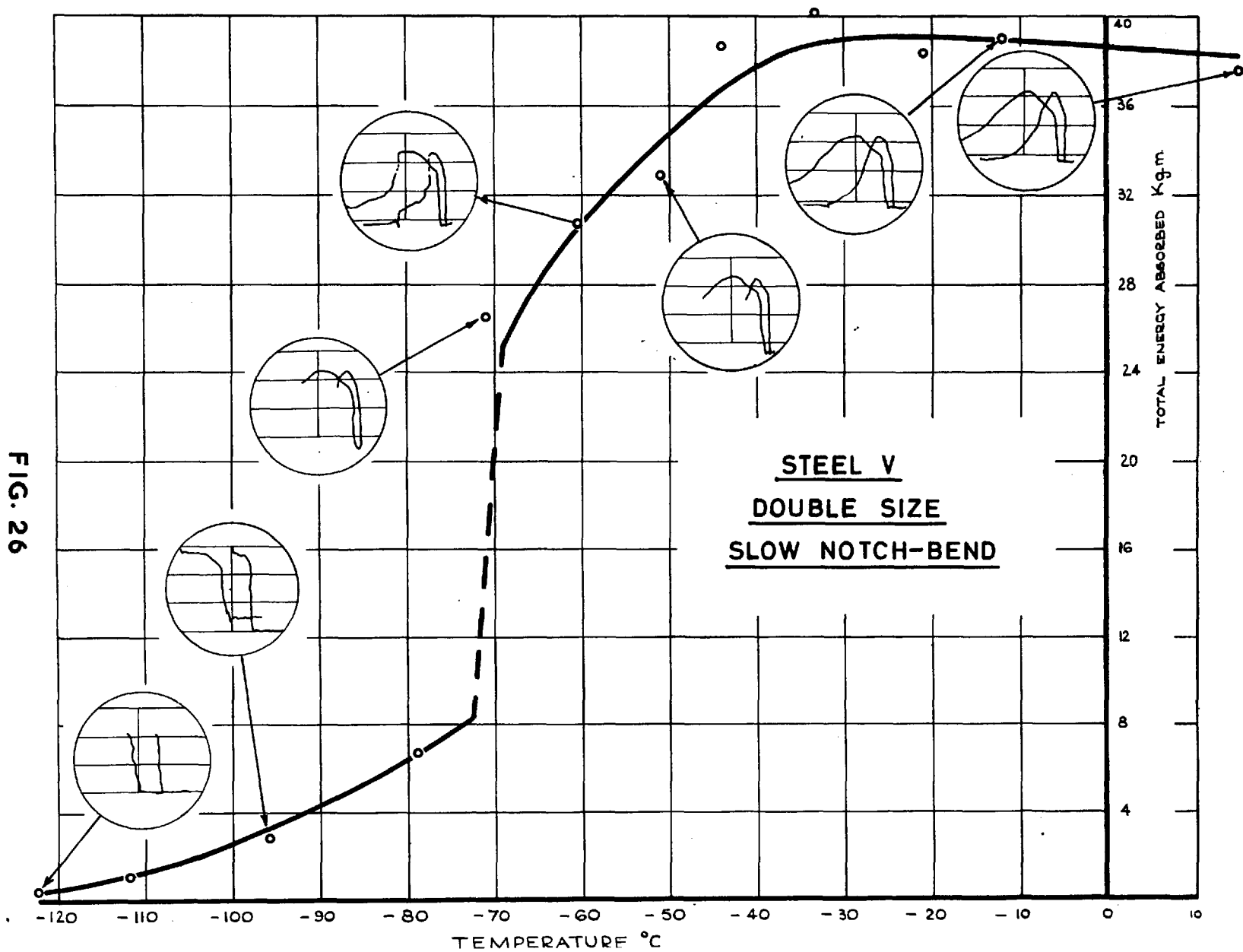


FIG. 26



EXPANSION-CONTRACTION RELATIONSHIP
STEEL 'P'

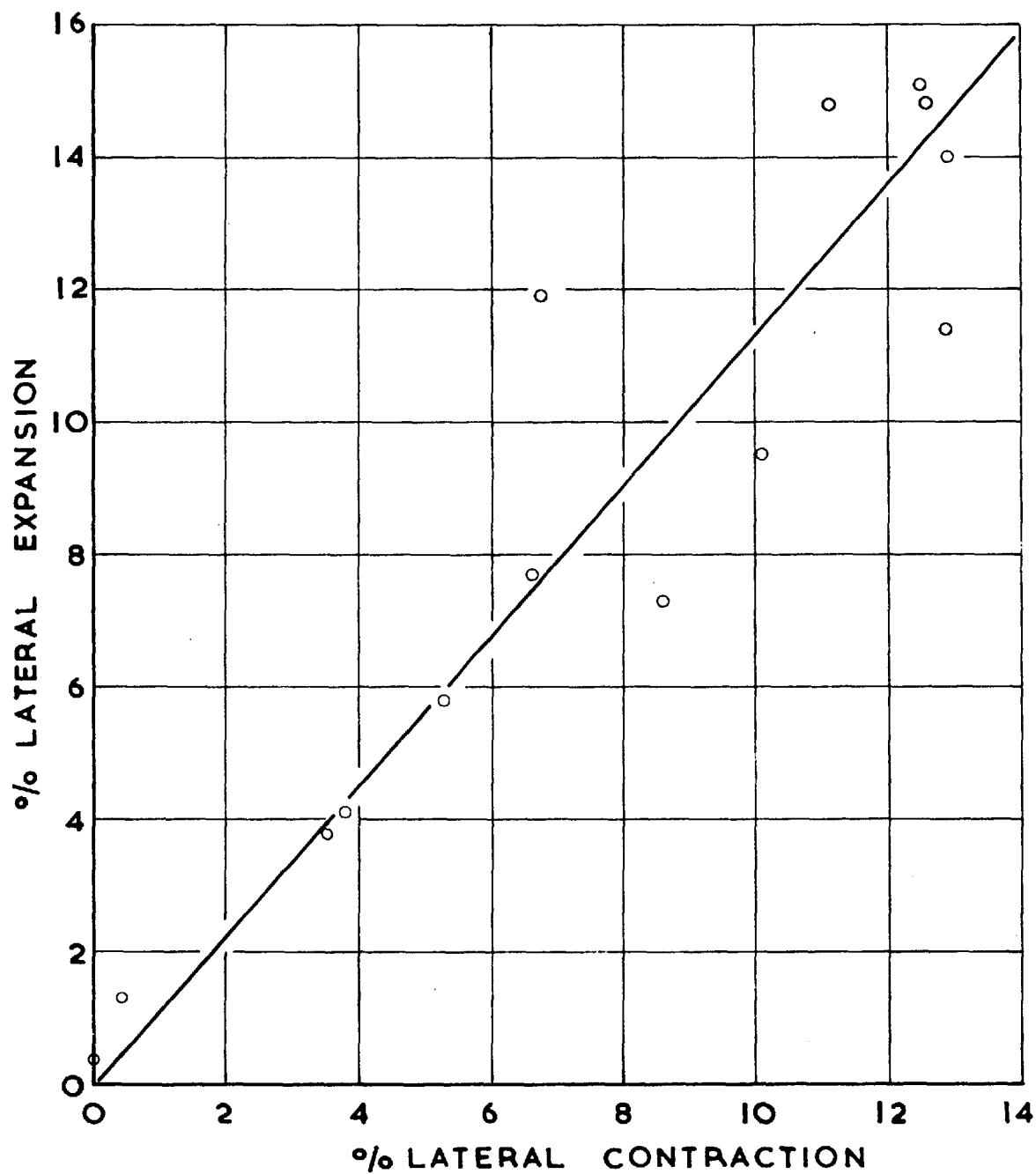
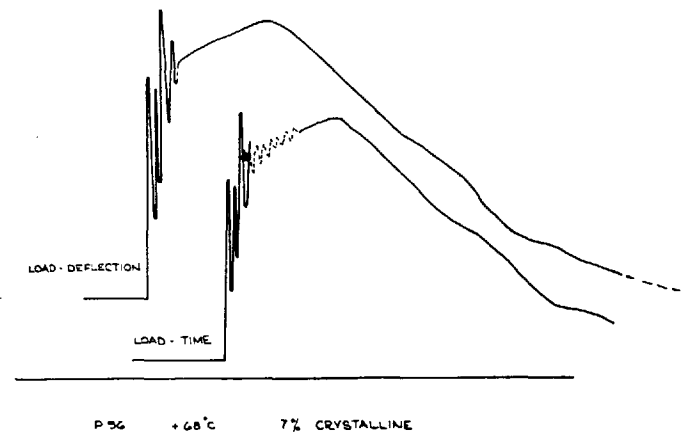
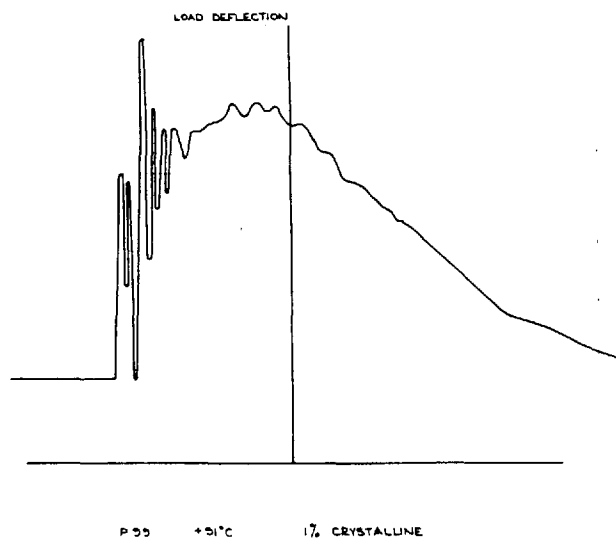


FIG. 27

FIG. 28 A



REPRODUCTIONS OF THE LOAD TIME
AND LOAD DEFLECTION DIAGRAMS
OBTAINED IN THE INSTRUMENTED
IMPACT TESTS

STEEL 'P'

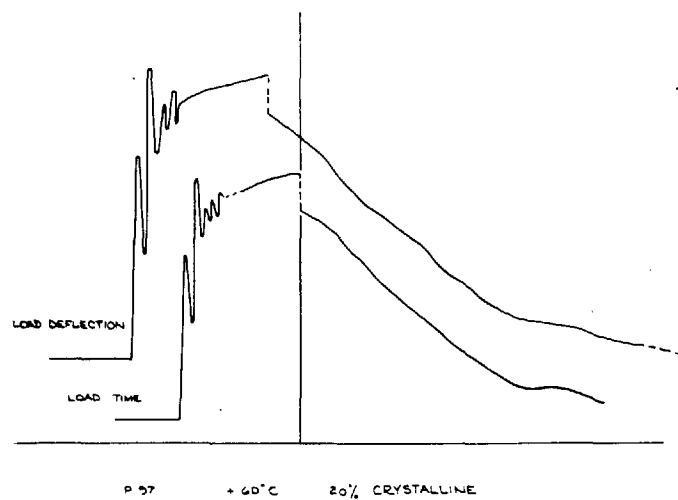
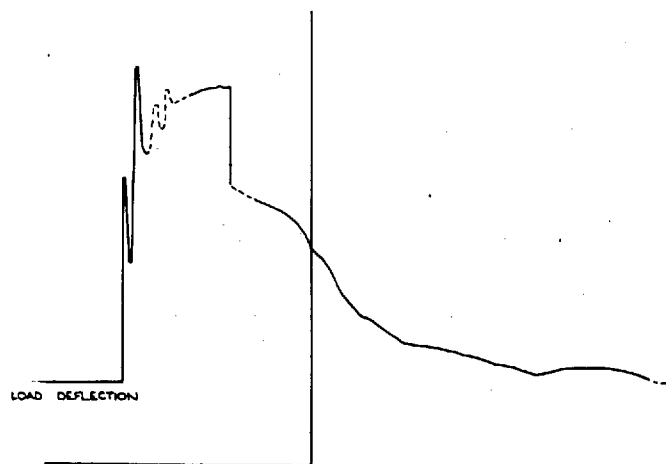
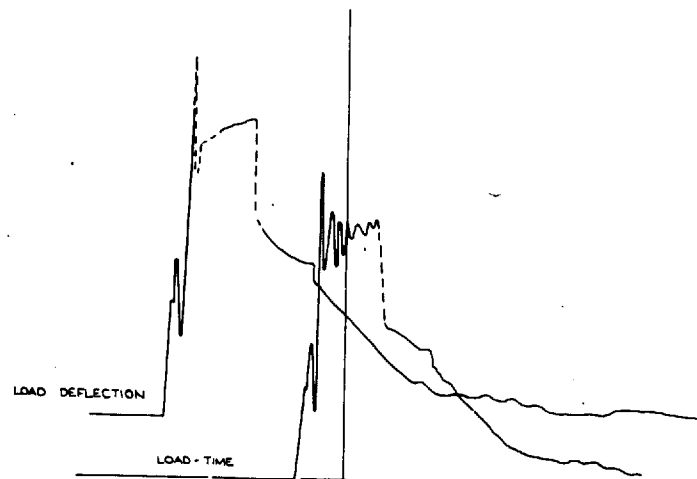


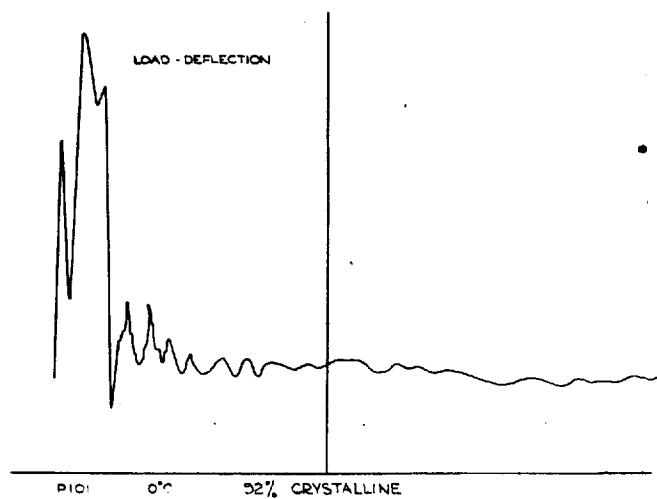
FIG. 28B



P 24 + 48°C 70% CRYSTALLINE

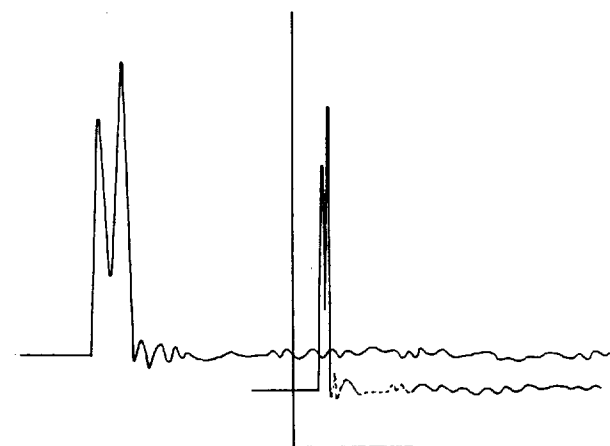


P 95 + 32°C 75% CRYSTALLINE

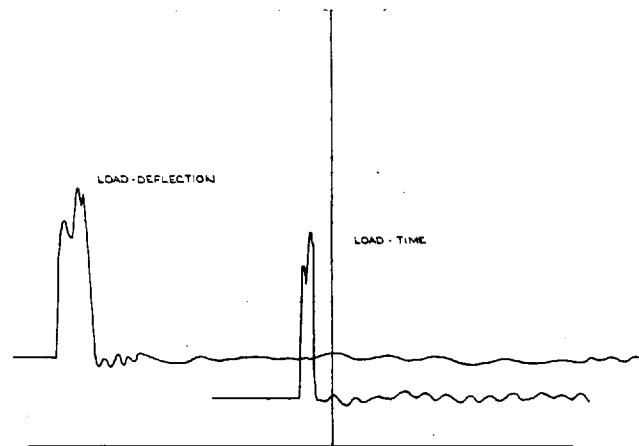


P101 0°C 52% CRYSTALLINE

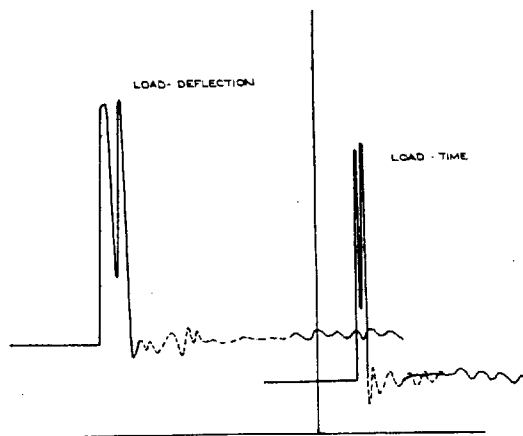
FIG. 28c



P103 -55°C 99% CRYSTALLINE

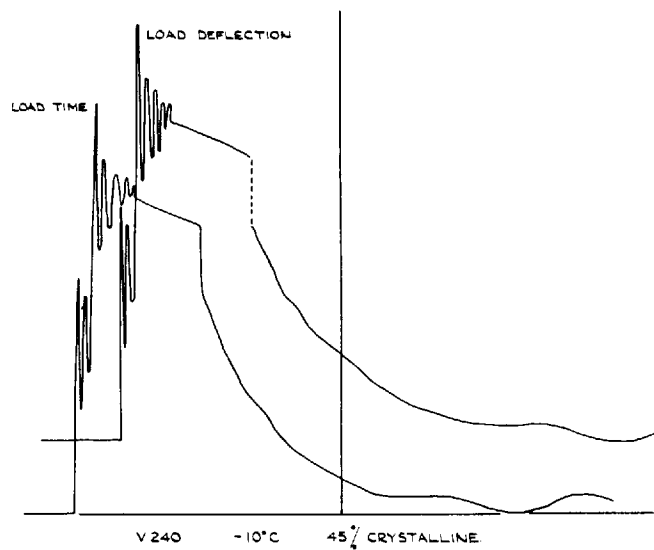
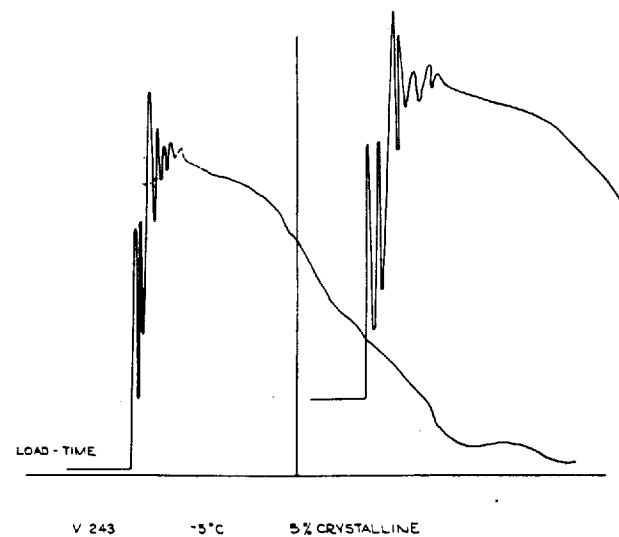
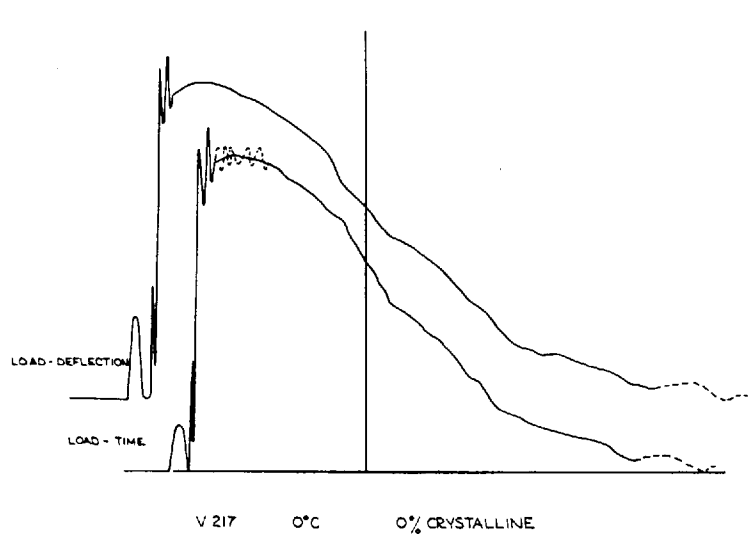


P104 -72°C 100% CRYSTALLINE



P106 -100°C 100% CRYSTALLINE

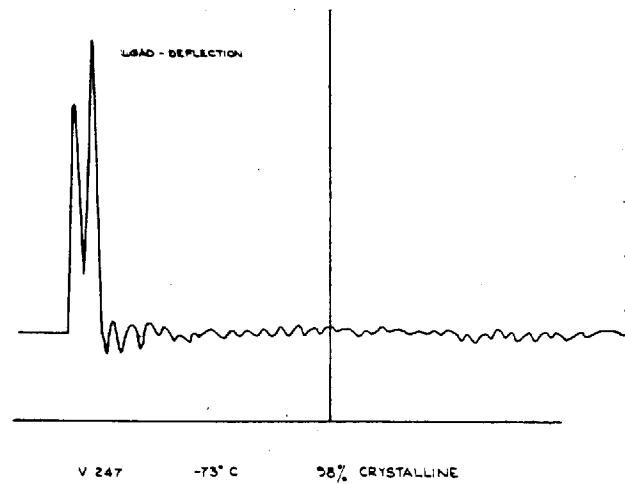
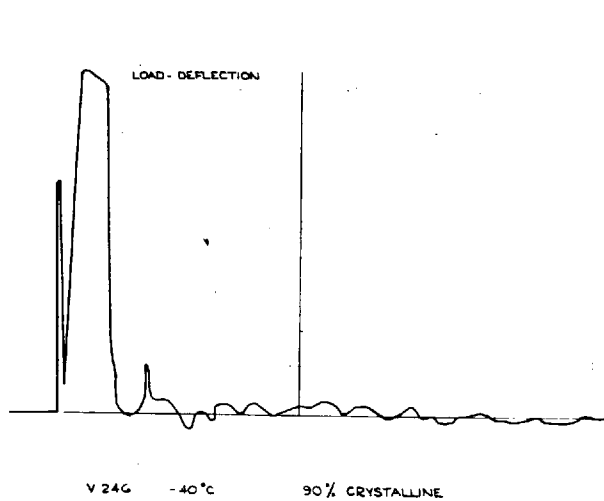
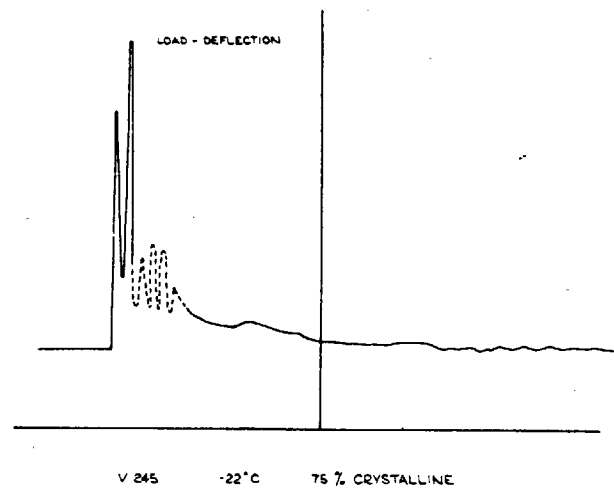
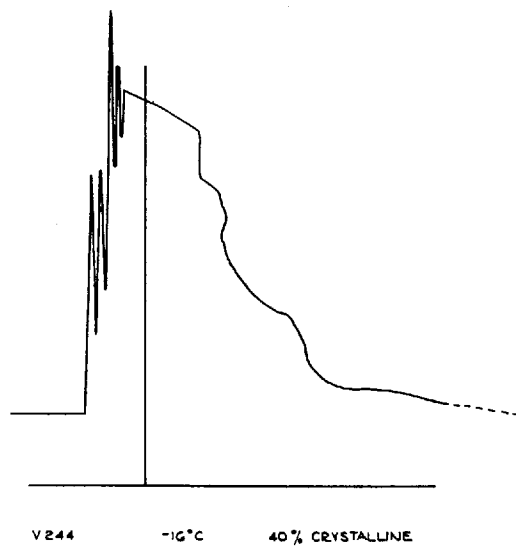
FIG. 29 A



REPRODUCTIONS OF THE LOAD TIME
AND LOAD DEFLECTION DIAGRAMS
OBTAINED IN THE INSTRUMENTED
IMPACT TESTS

STEEL 'V'

FIG. 29 B



LOAD- DEFLECTION DIAGRAM FOR THE CHARPY IMPACT TEST

LOAD -TIME DIAGRAM IS SHOWN REPLOTED TO
THE SAME SCALE; 1) ASSUMING THAT THE
VELOCITY DECREASES UNIFORMLY WITH TIME.
2) ASSUMING THAT THE
VELOCITY REMAINS CONSTANT AT THE FINAL
VALUE $[-\circ-\circ-]$.

STEEL V 0°C

LOAD [1 IN. = 0.18 TON]

ORIGIN OF REPLOT

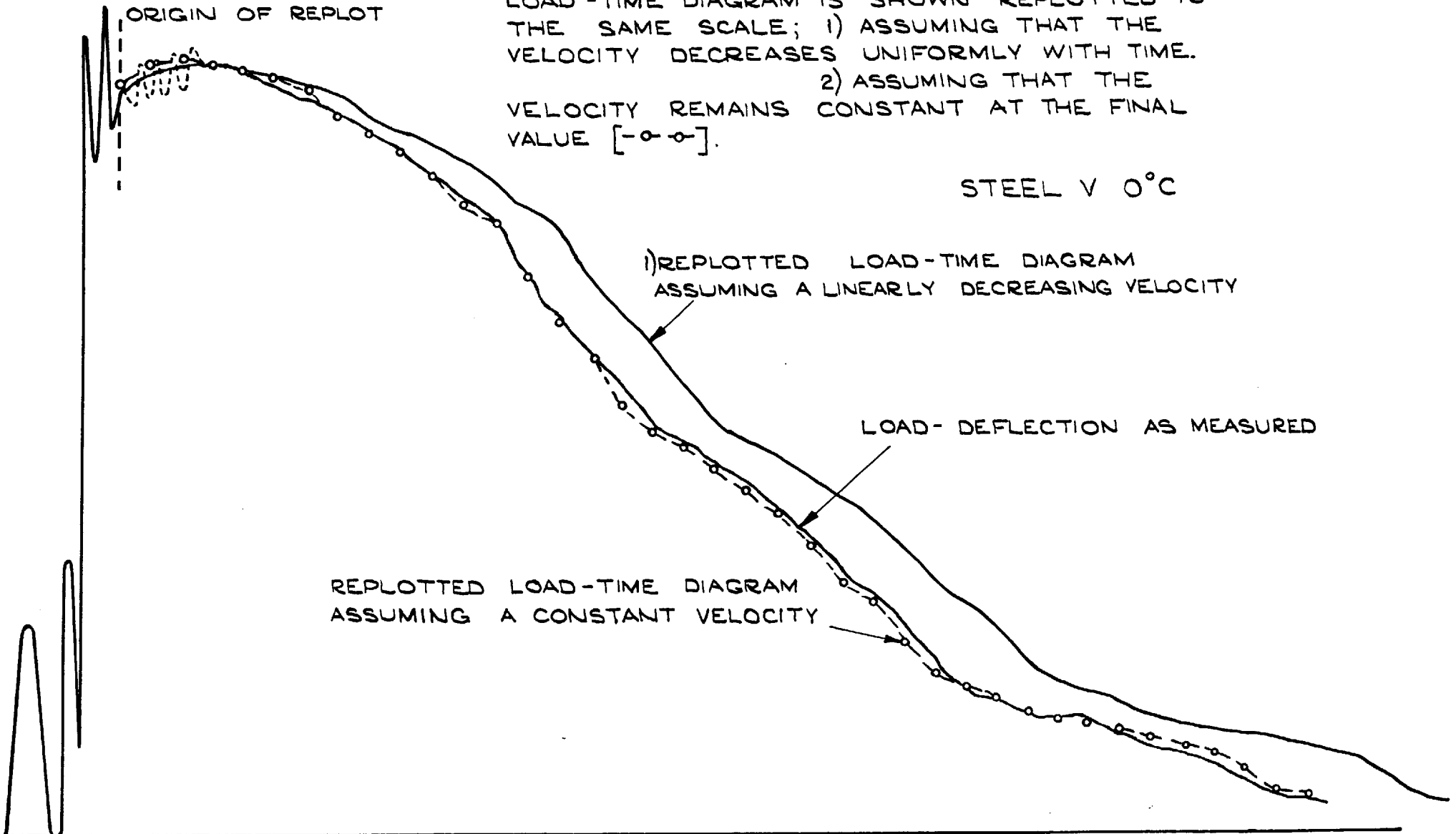
1) REPLOTED LOAD-TIME DIAGRAM
ASSUMING A LINEARLY DECREASING VELOCITY

LOAD- DEFLECTION AS MEASURED

REPLOTED LOAD-TIME DIAGRAM
ASSUMING A CONSTANT VELOCITY

DEFLECTION [1 IN. = 0.0516 IN. DEFLECTION]

FIG. 30



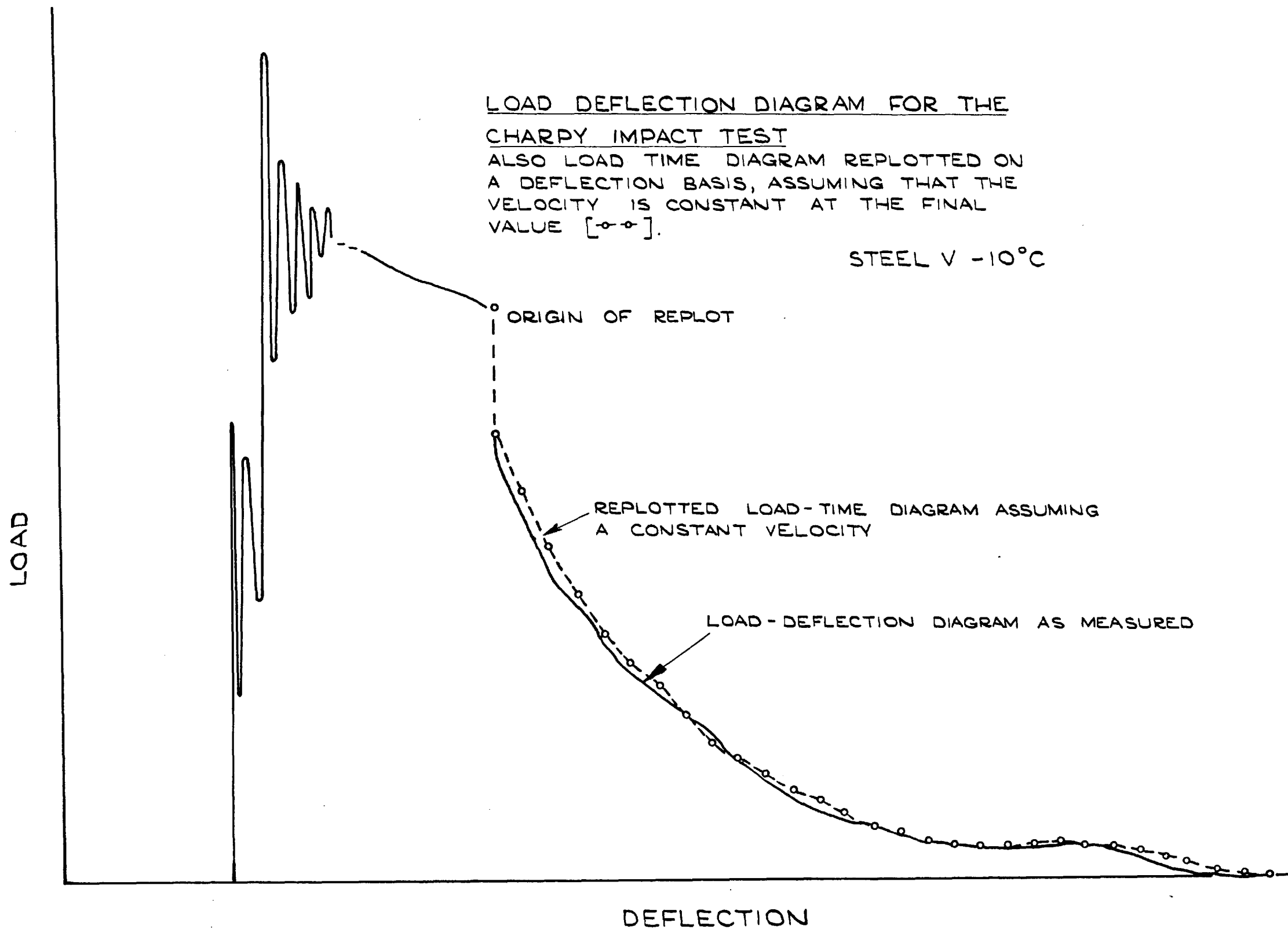


FIG.3I

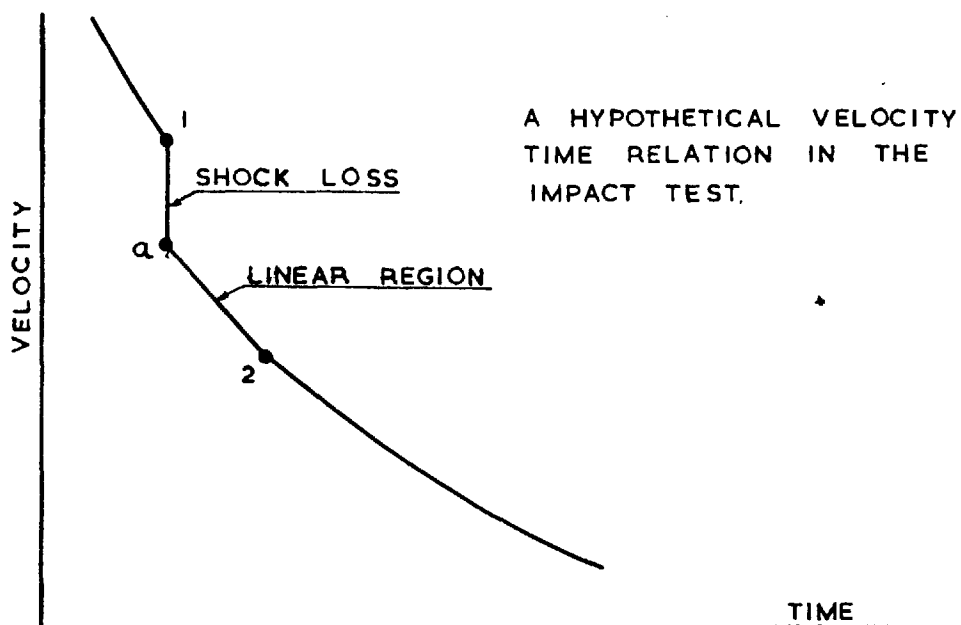
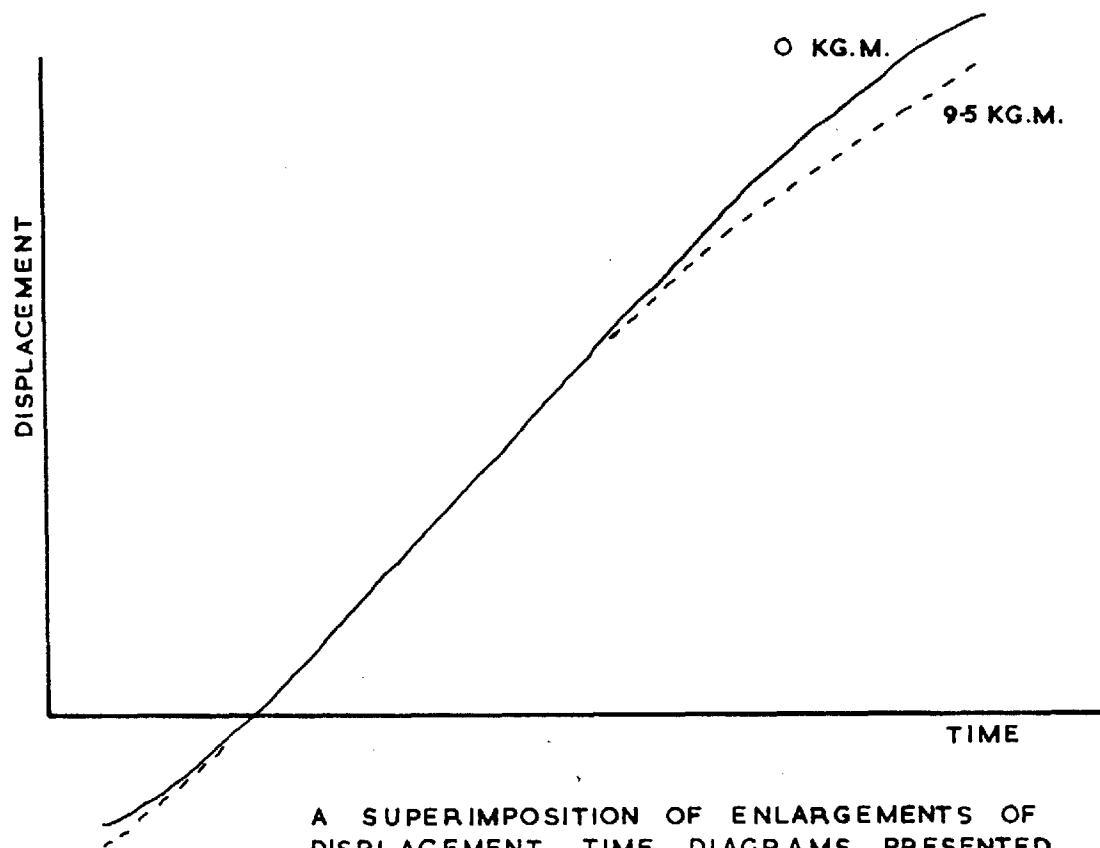


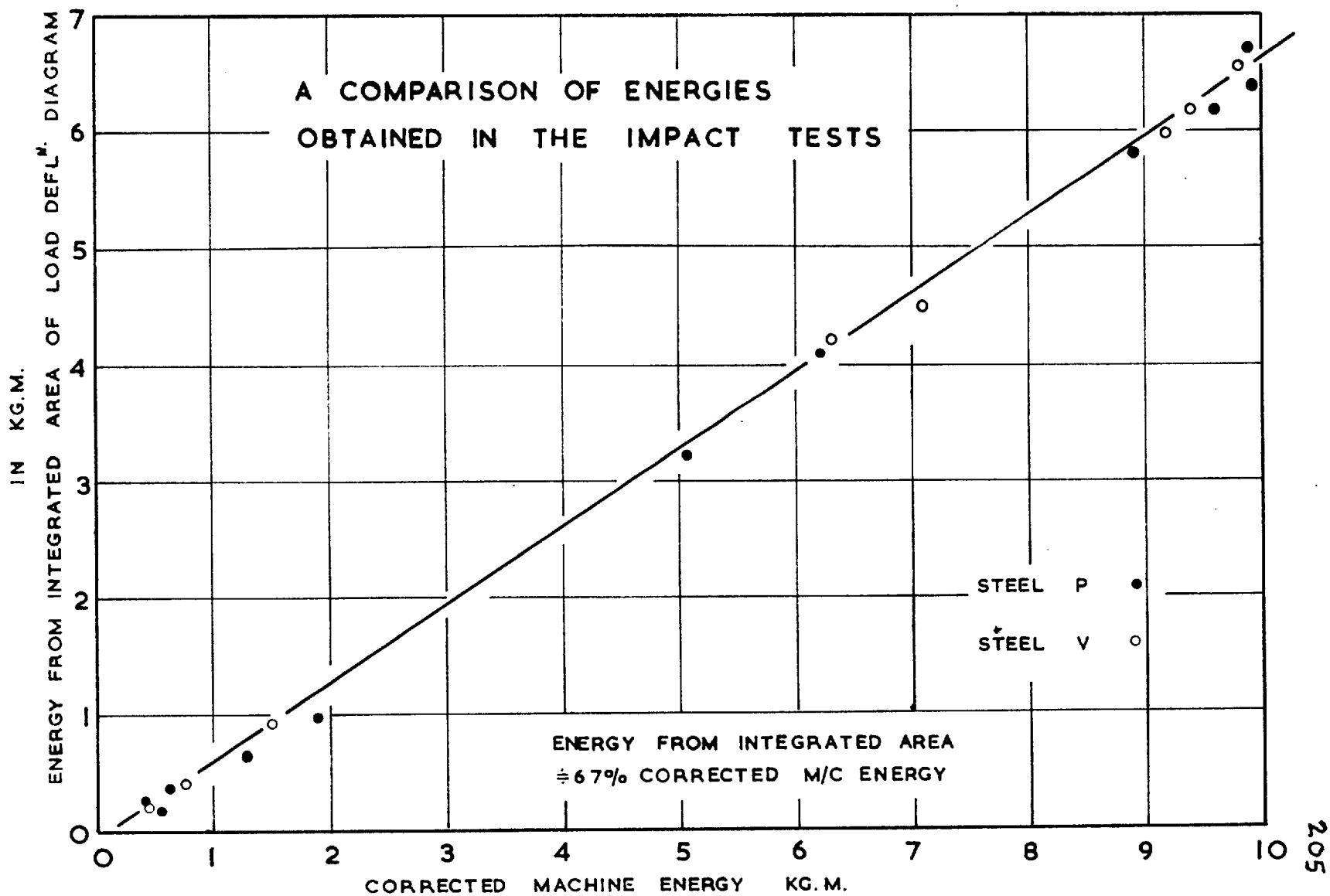
FIG. 33

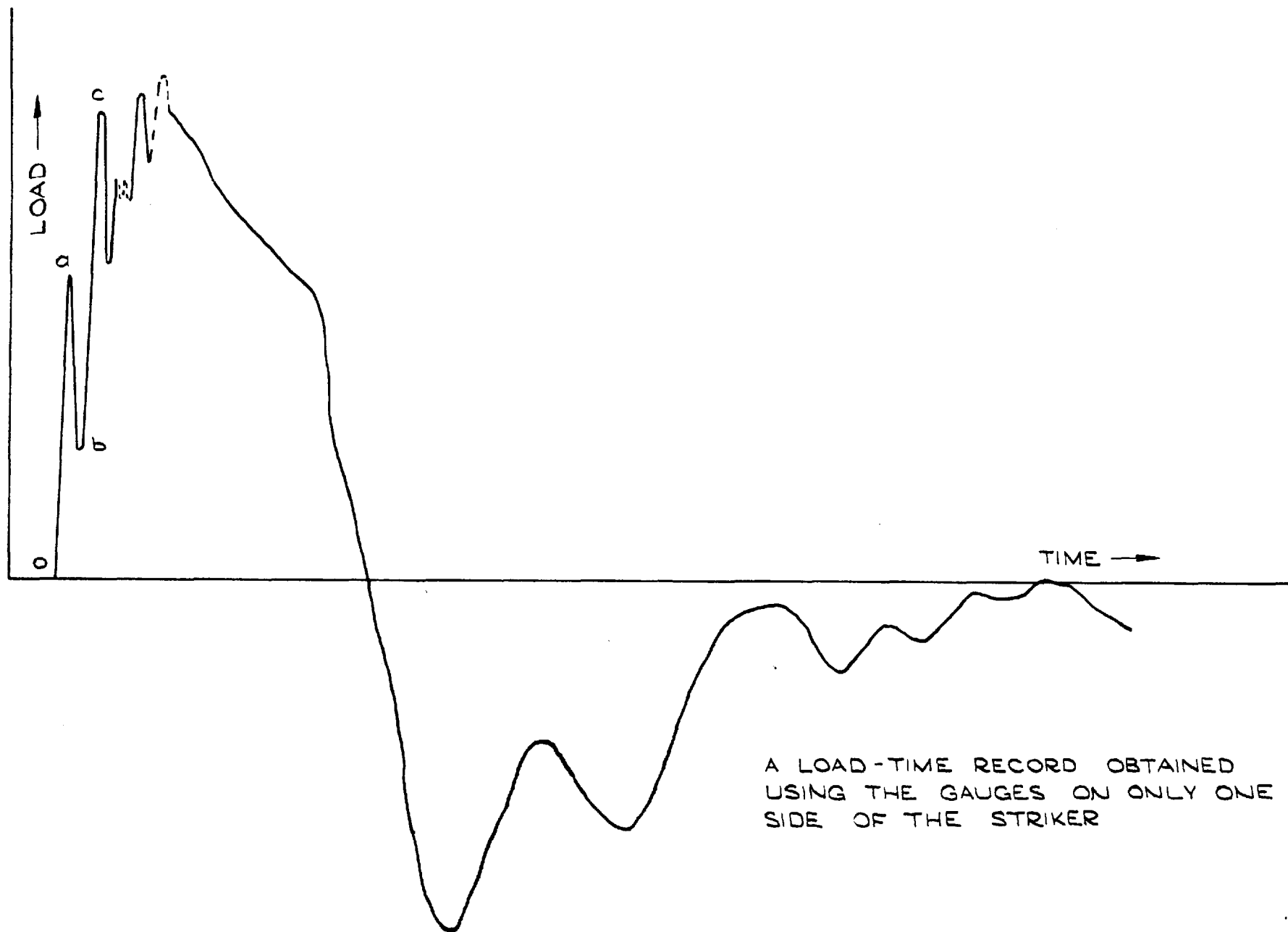


A SUPERIMPOSITION OF ENLARGEMENTS OF
DISPLACEMENT TIME DIAGRAM PRESENTED
BY TANAKA AND UMEKAWA.

FIG. 32

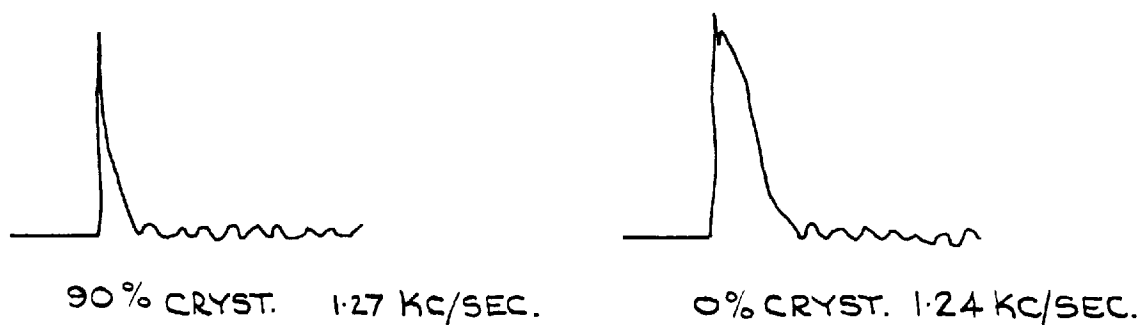
FIG. 34





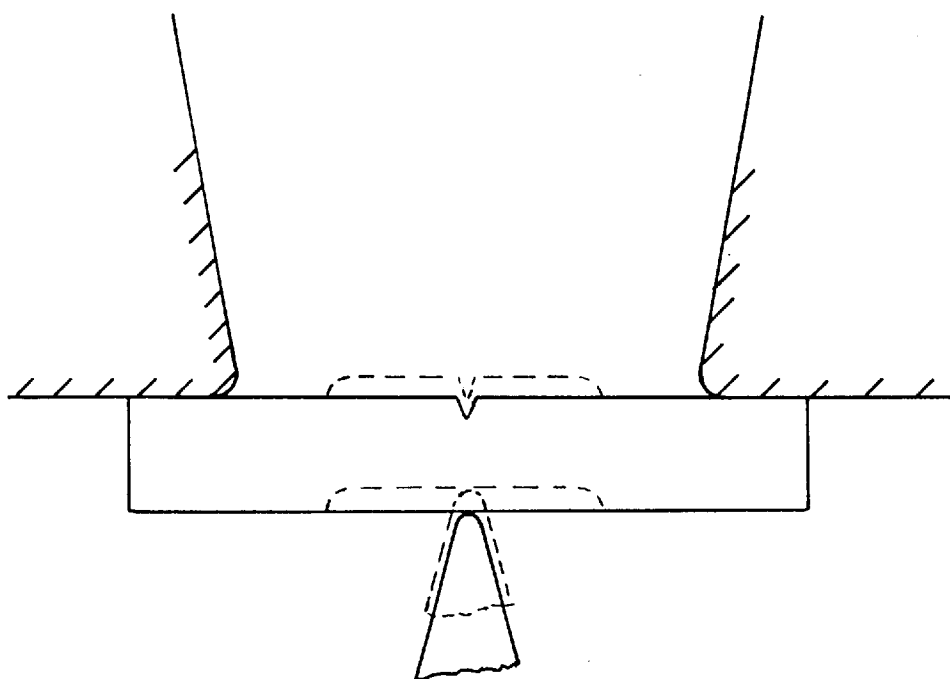
A LOAD-TIME RECORD OBTAINED
USING THE GAUGES ON ONLY ONE
SIDE OF THE STRIKER

FIG 35



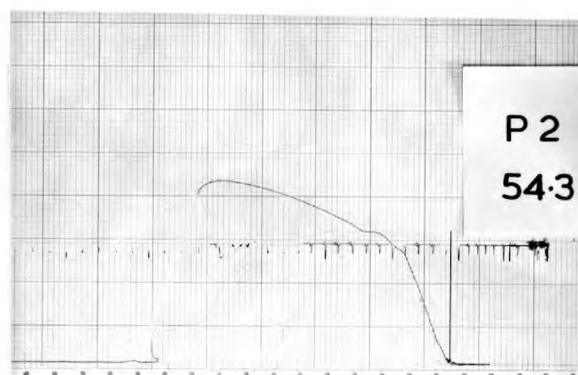
RESIDUAL OSCILLATIONS IN
THE LOAD CELL

FIG. 36

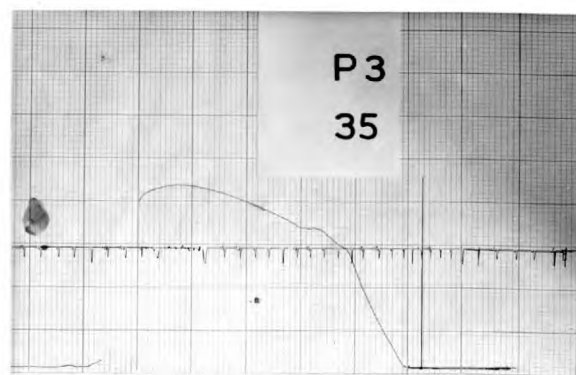


EFFECTIVE REDUCTION IN SPECIMEN LENGTH
BY THE EFFECT OF A SHEAR WAVE

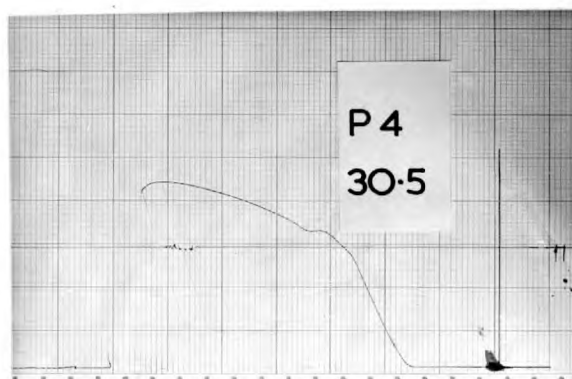
FIG. 37



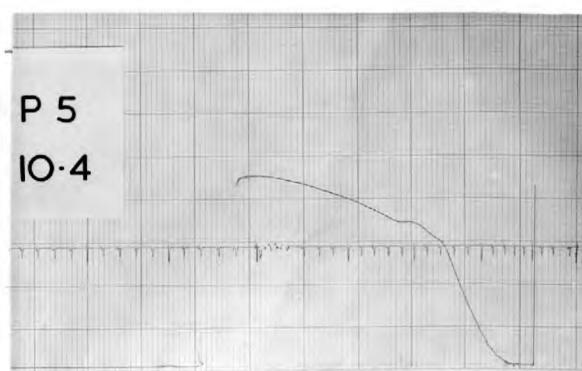
0 %



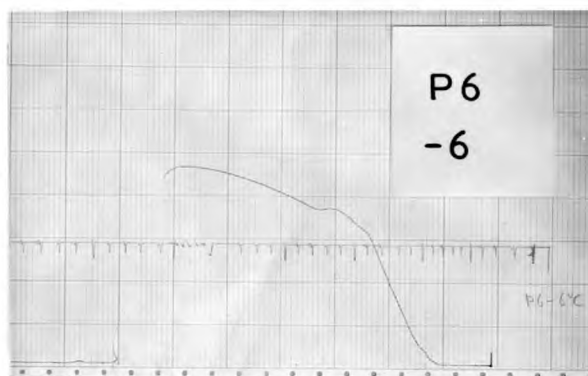
10 %



45 %



80 %



55 %

TESTING TEMPERATURES ARE
GIVEN ON THE DIAGRAMS IN °C.
% CRYSTALLINITIES ARE NOTED
BELOW.

SCALE:

↑ 1 GRAPH DIV. = 1.0 TON
← " " = 0.05 IN.

FIG. 38
LOAD EXTENSION DIAGRAMS FOR THE
NOTCH-TENSILE TESTS
(TO A REDUCED SCALE)

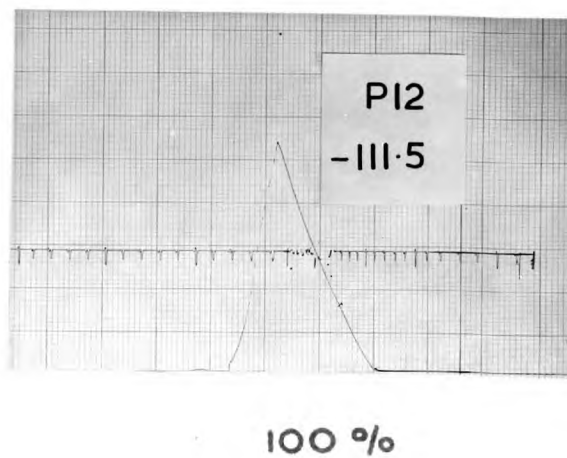
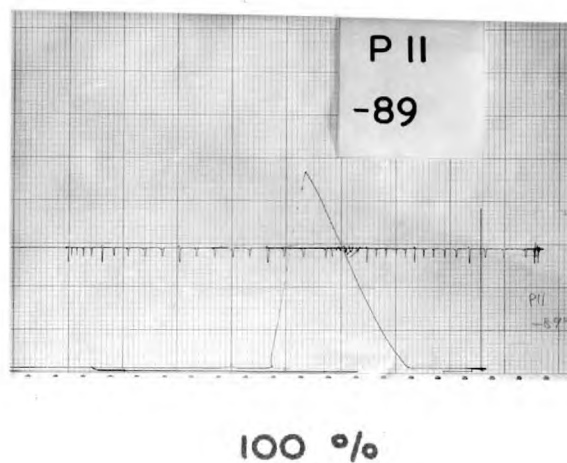
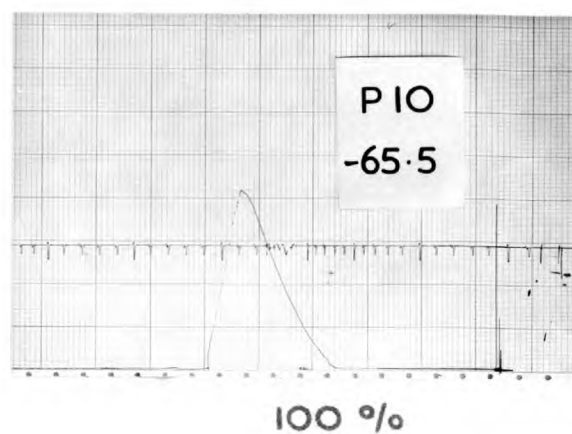
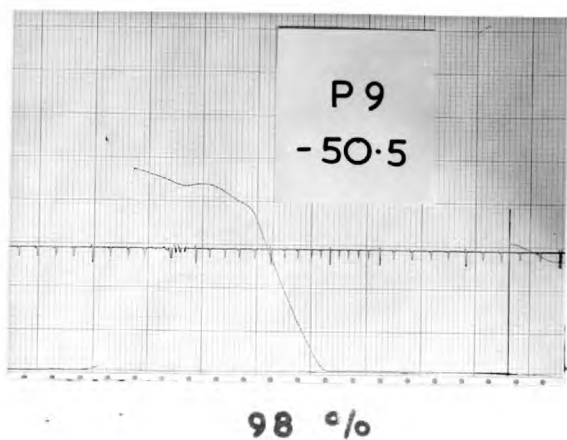
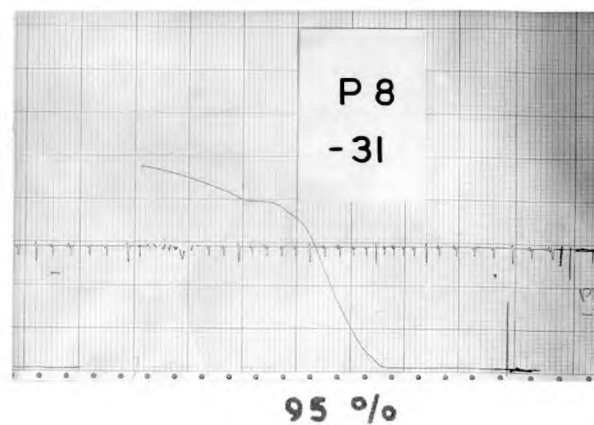
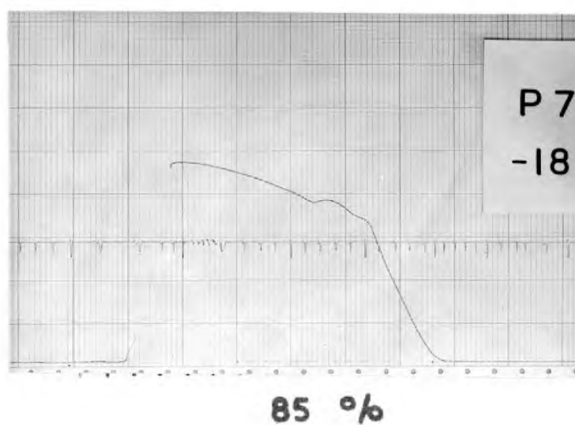
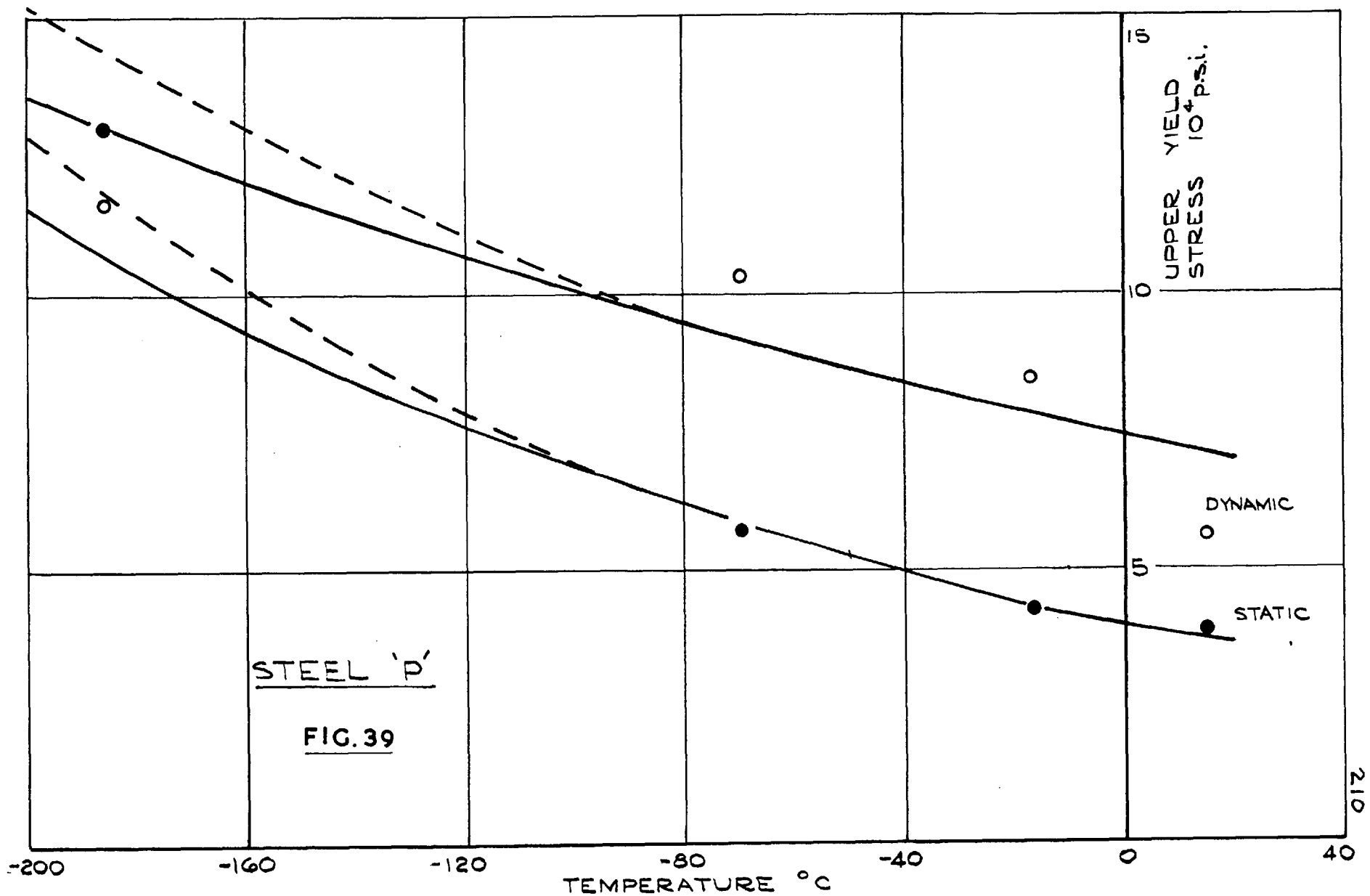


FIG 38 CONTINUED



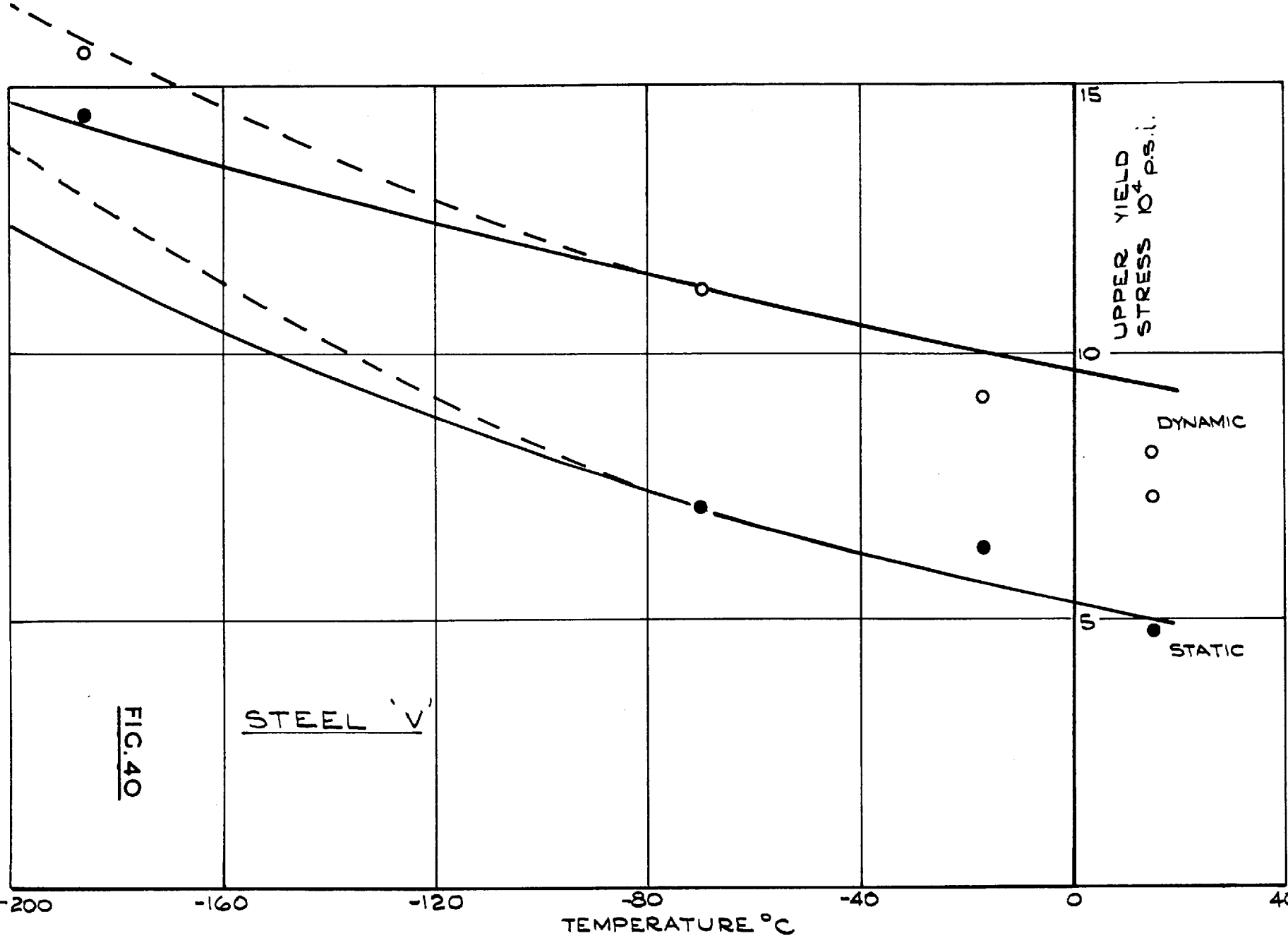


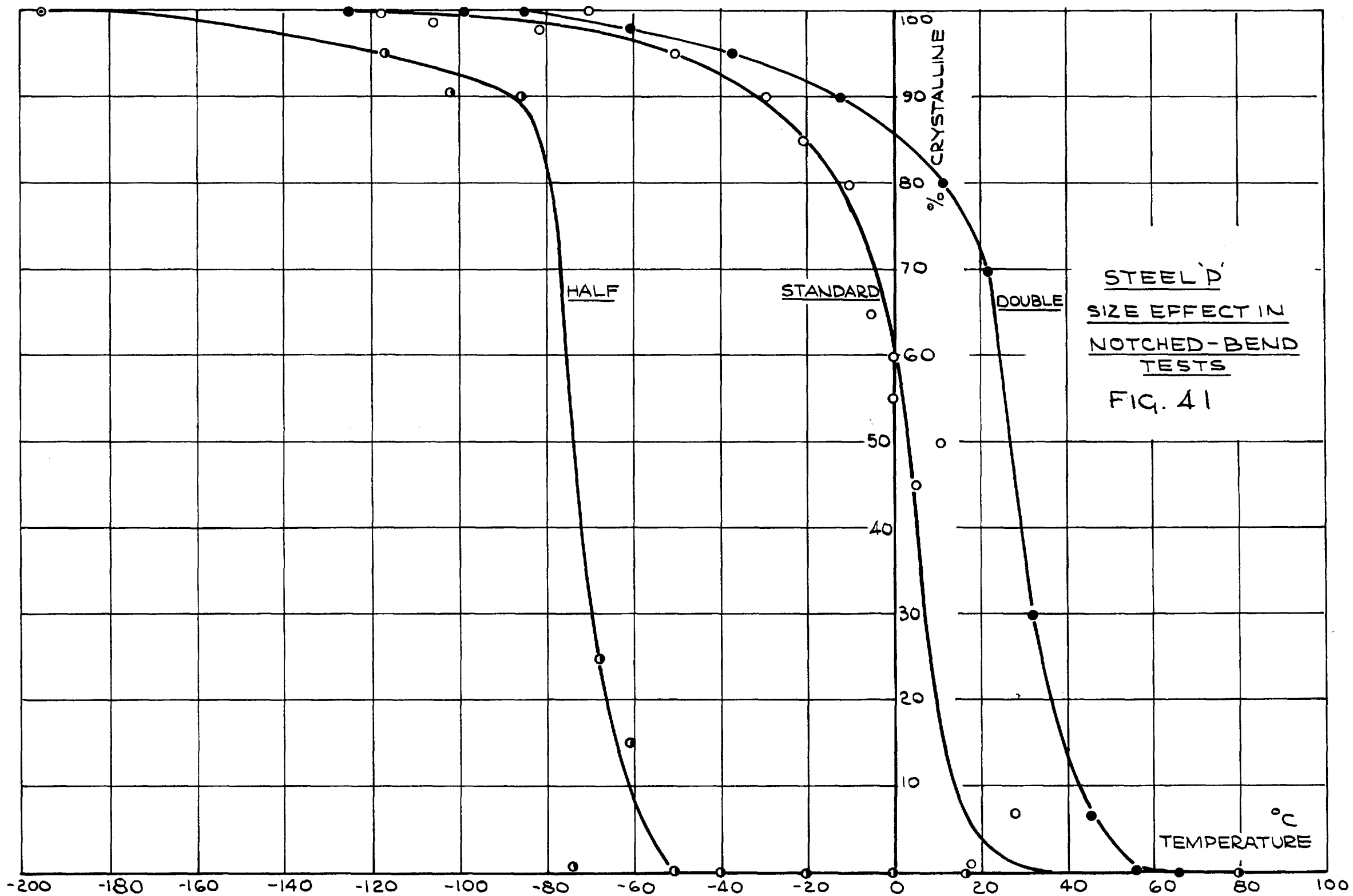
FIG. 40

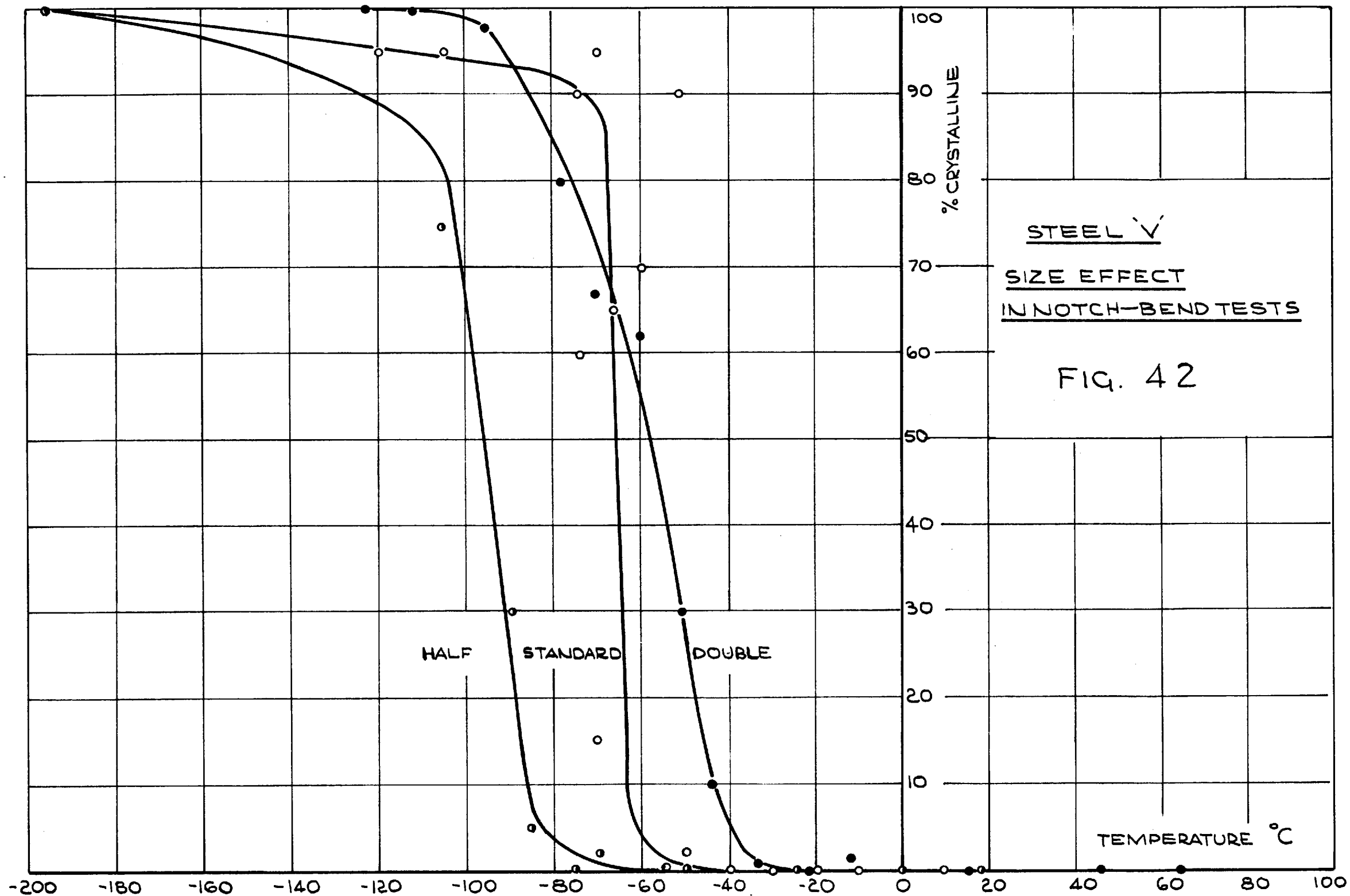
STEEL 'V'

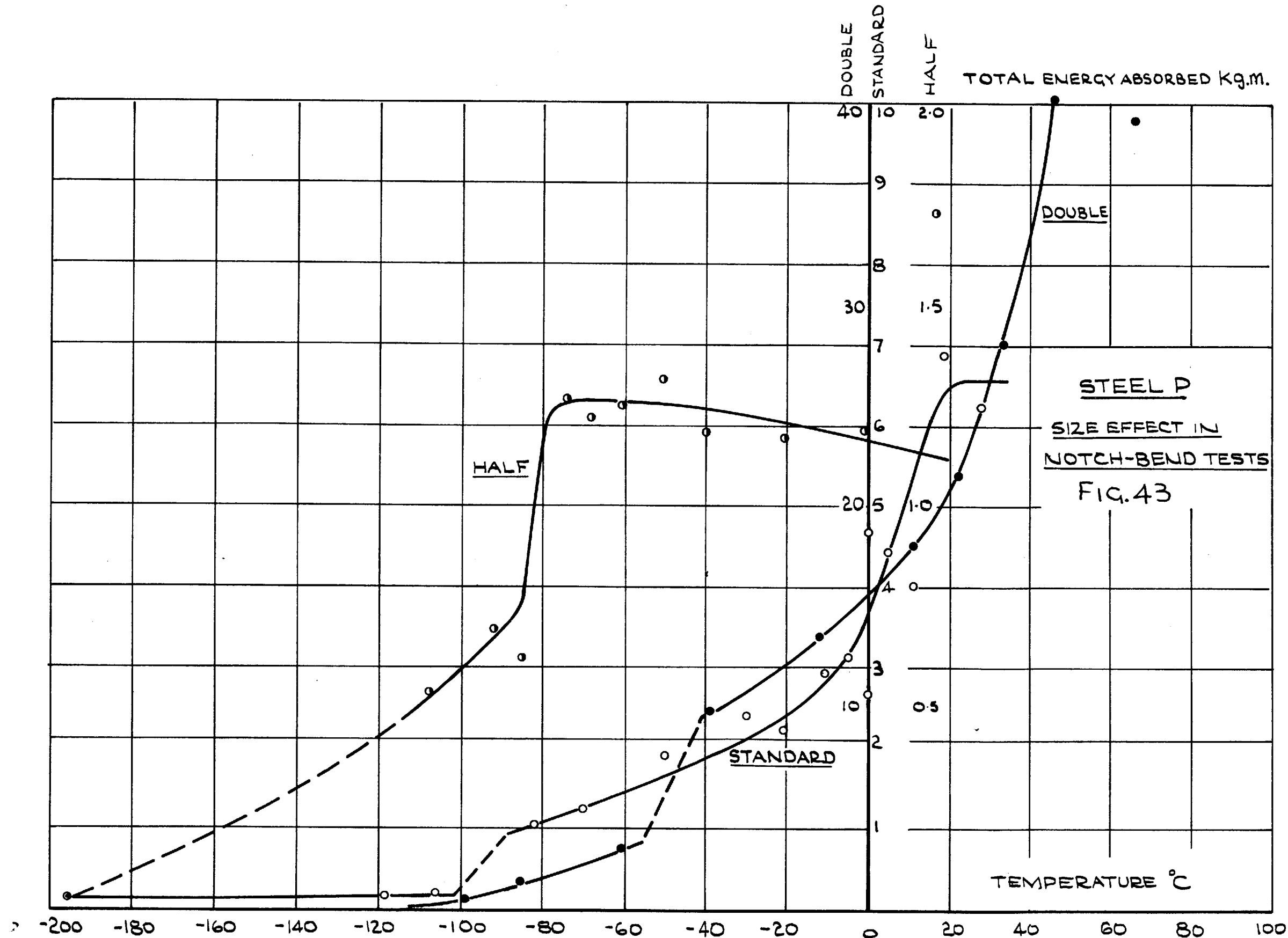
UPPER YIELD
STRESS 10⁴ ps.i.

DYNAMIC

STATIC

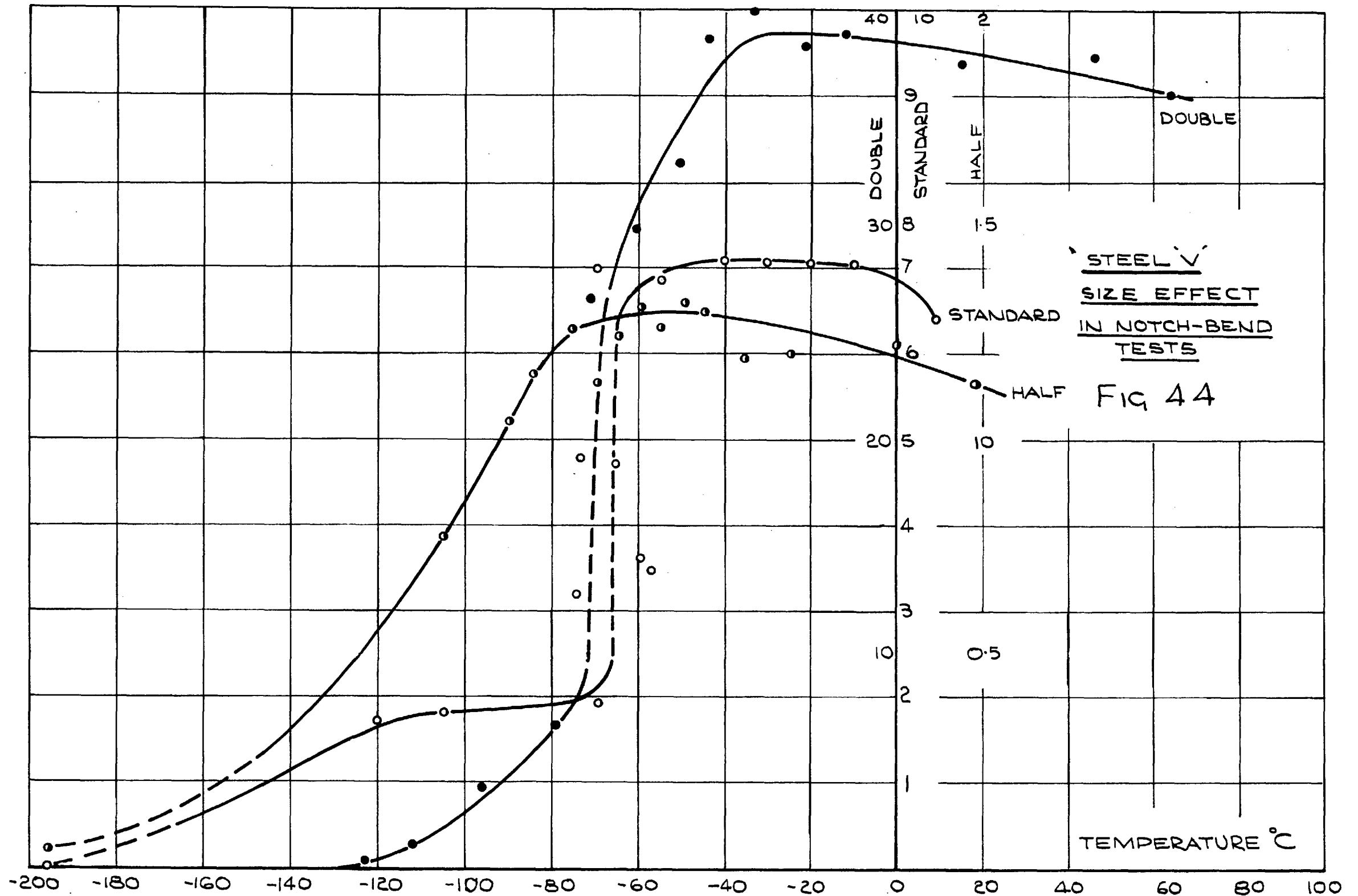


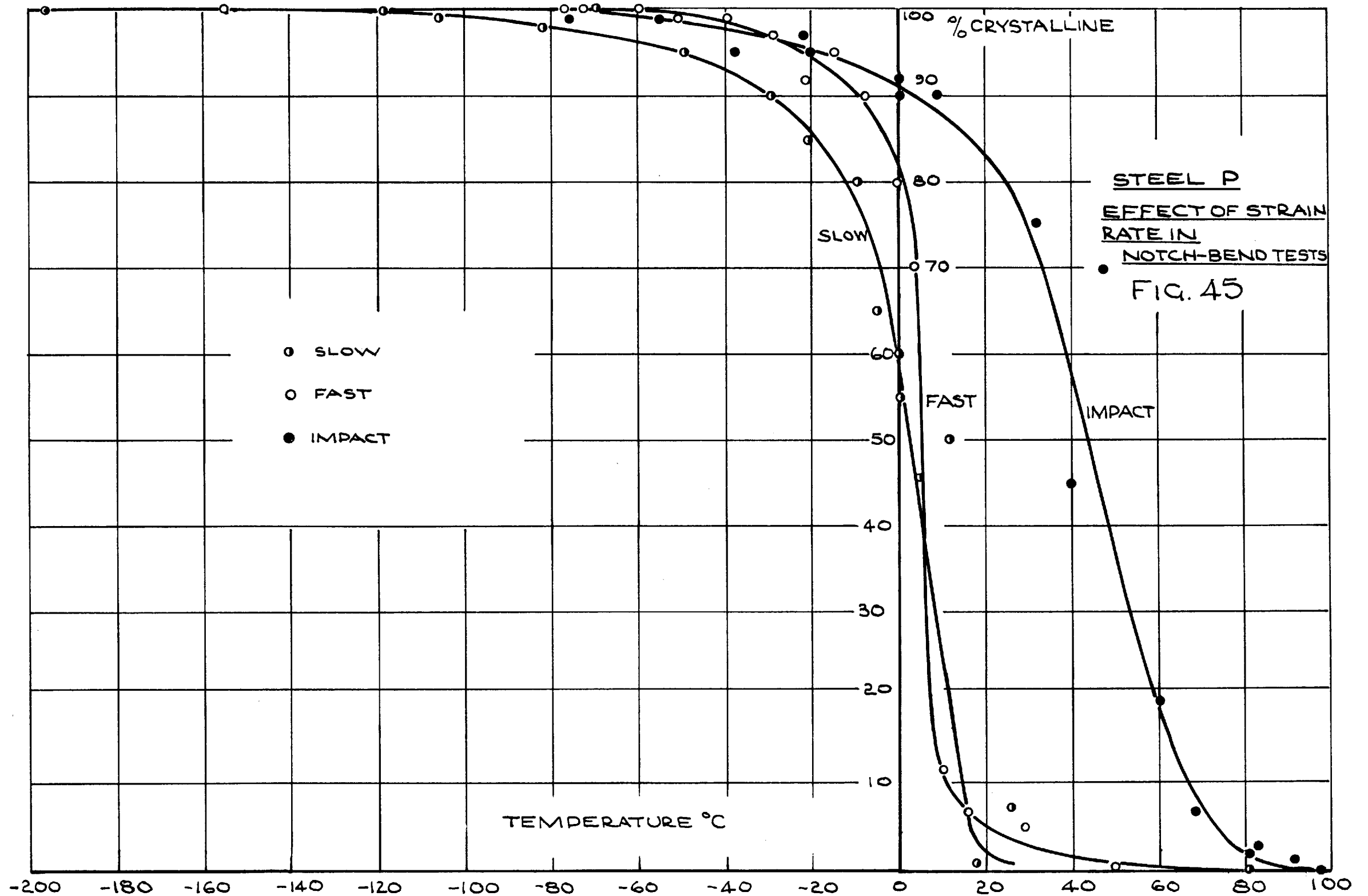


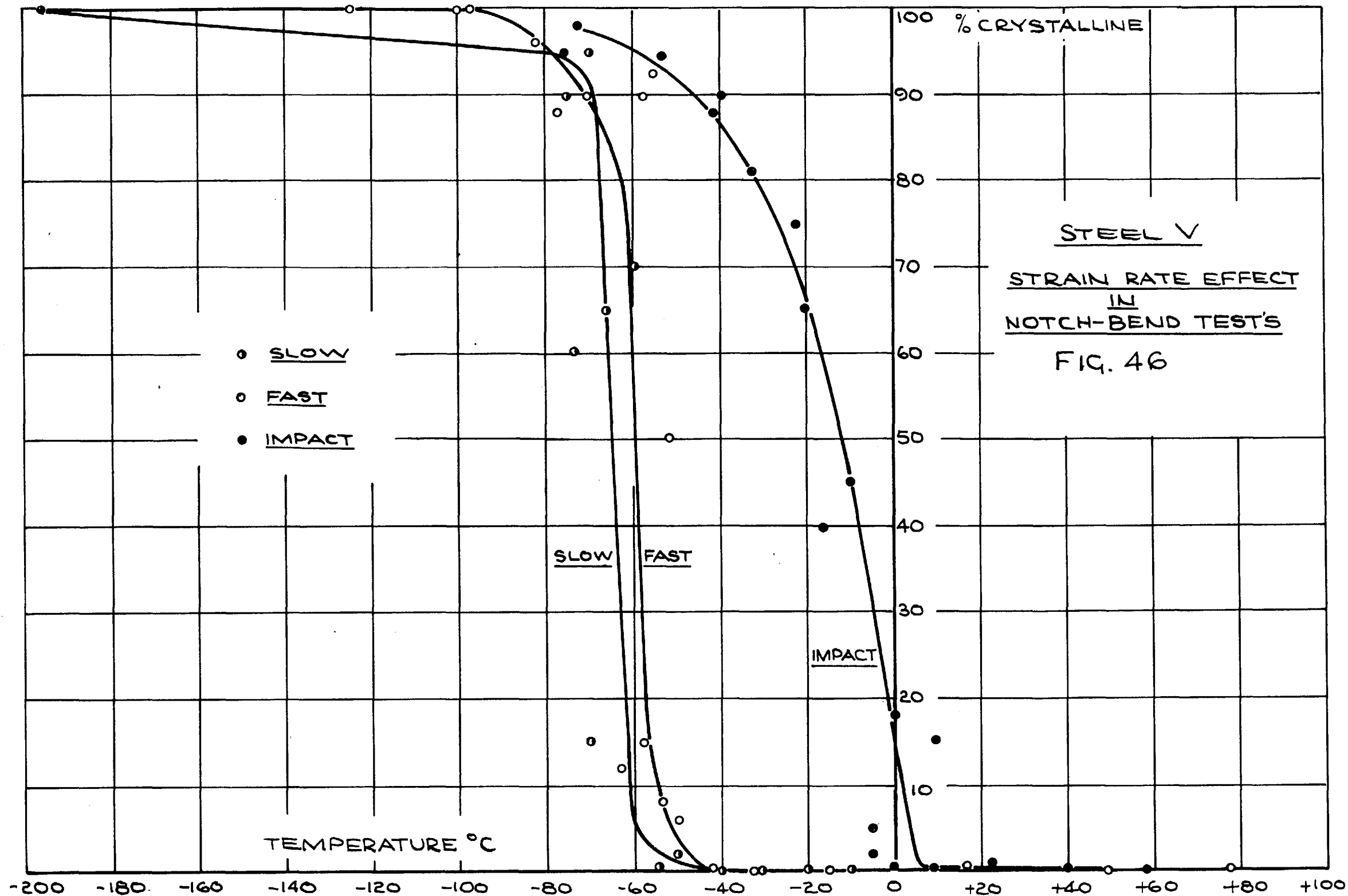


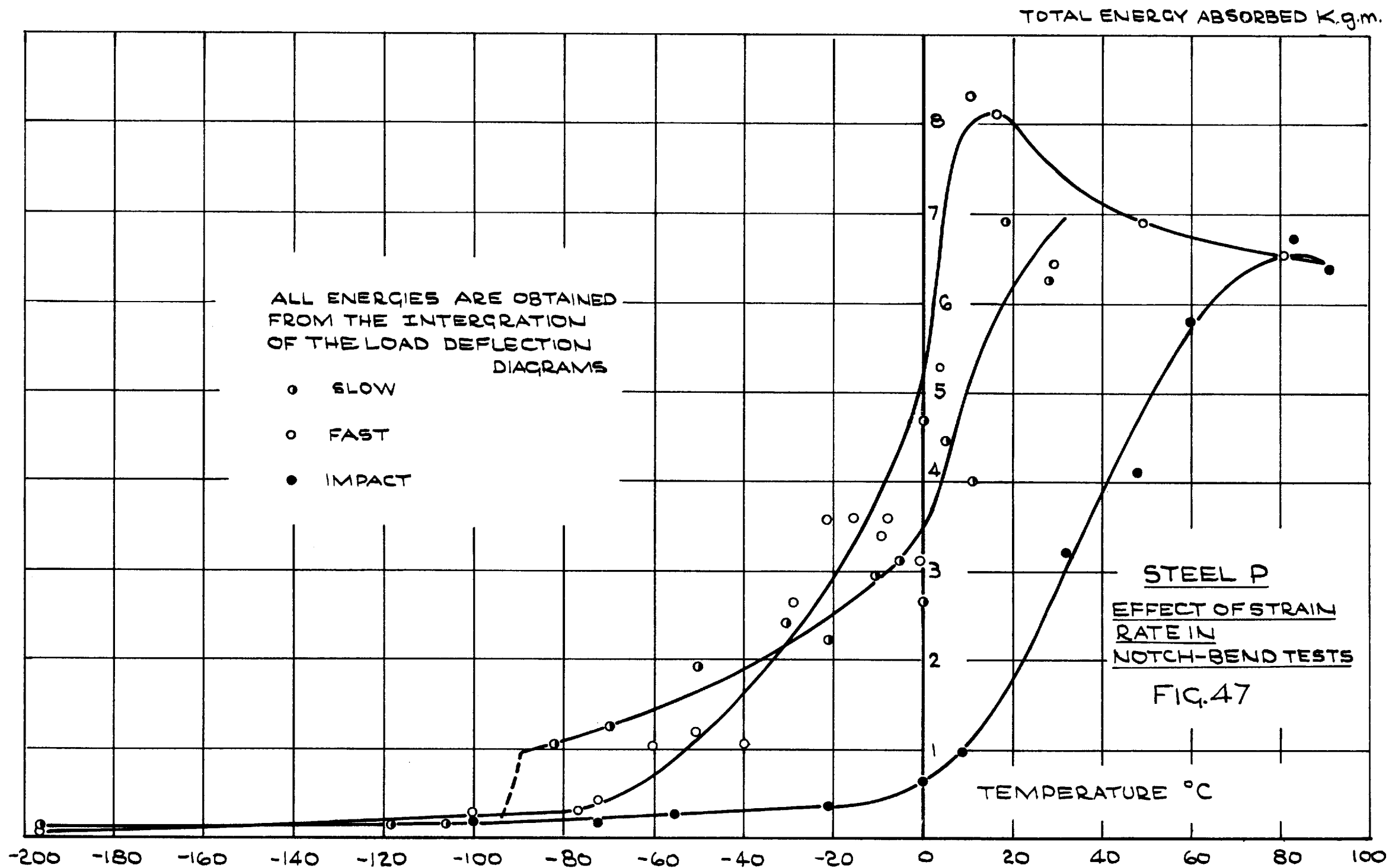
STEEL P
SIZE EFFECT IN
NOTCH-BEND TESTS
FIG. 43

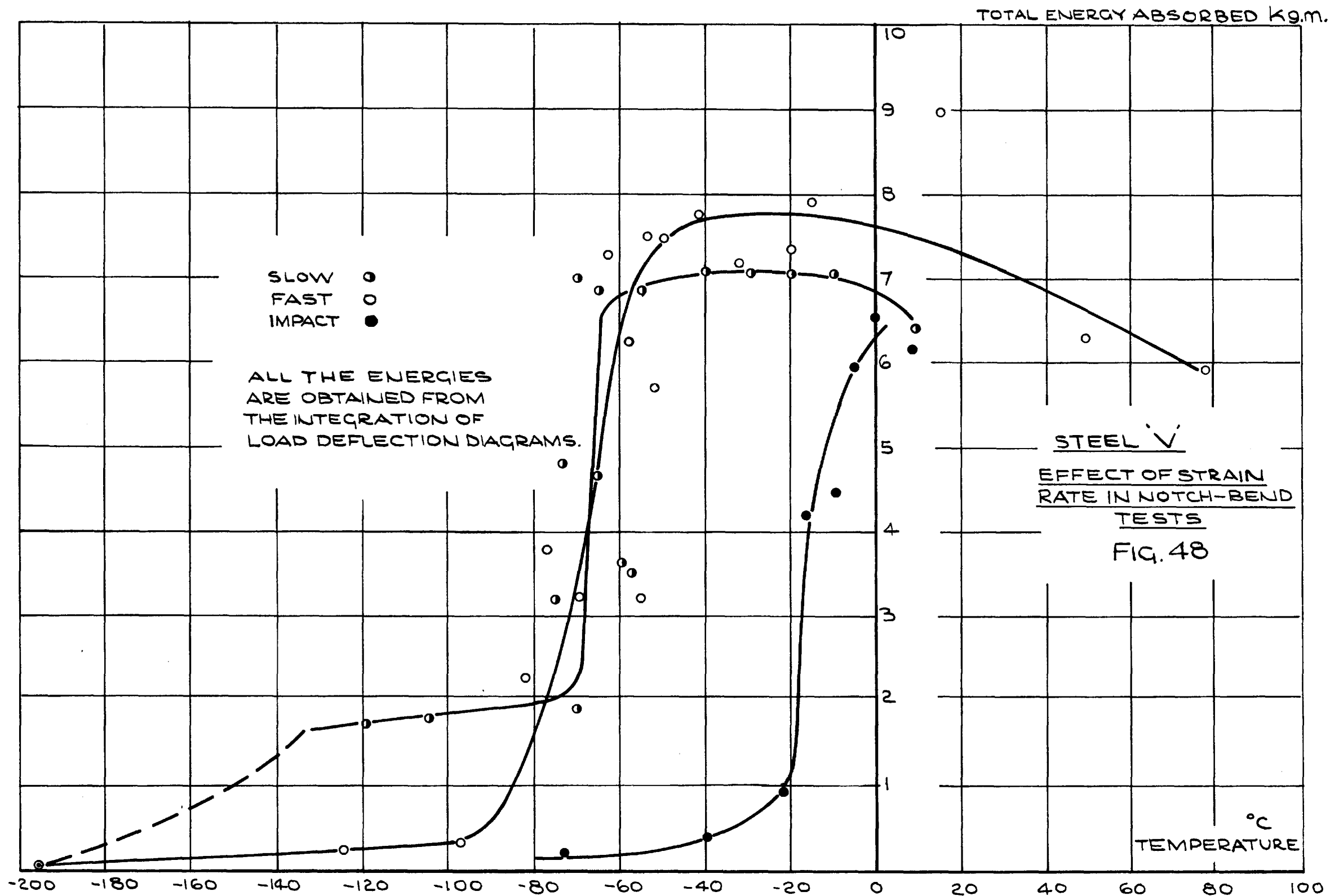
TOTAL ENERGY ABSORBED kg.m.

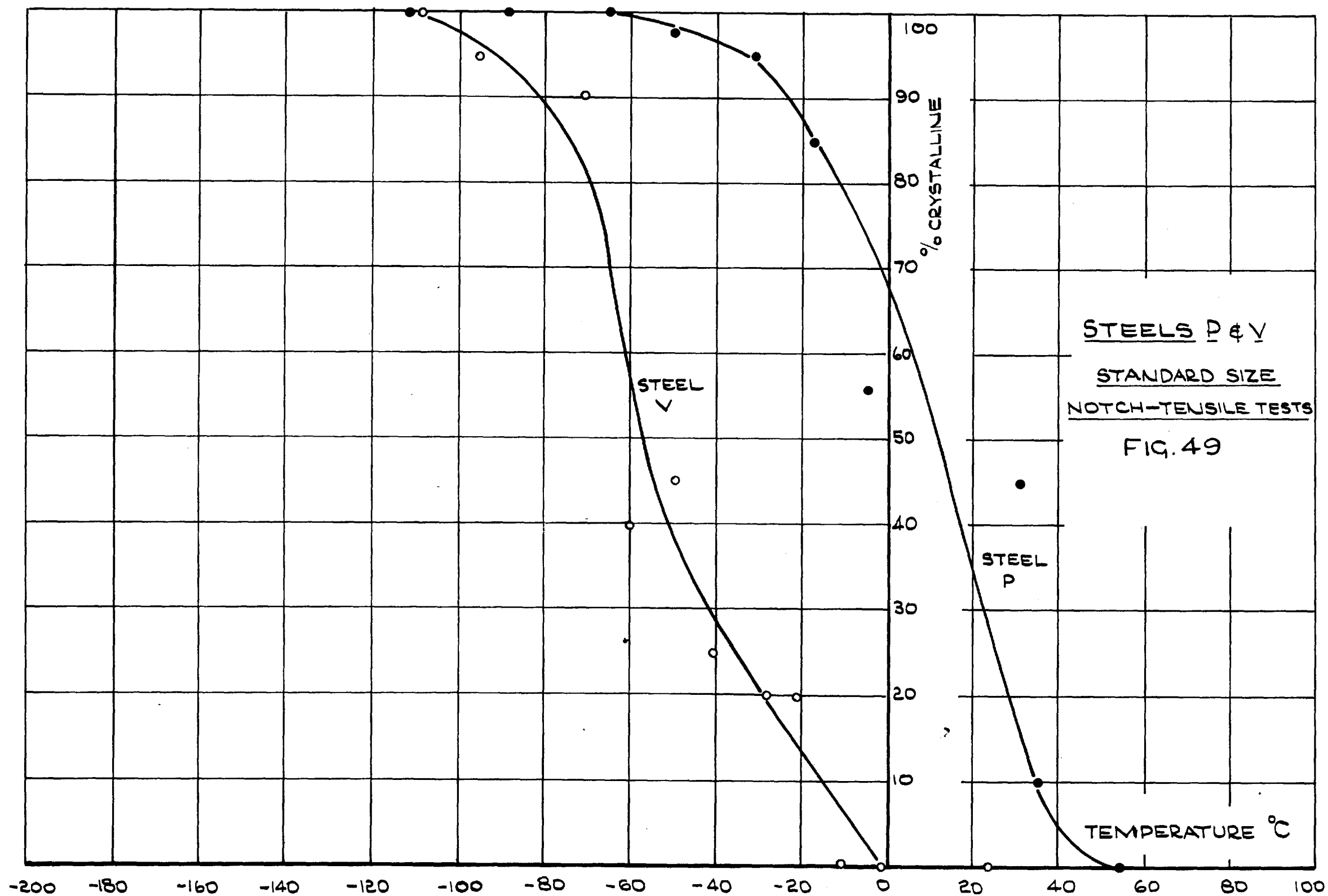


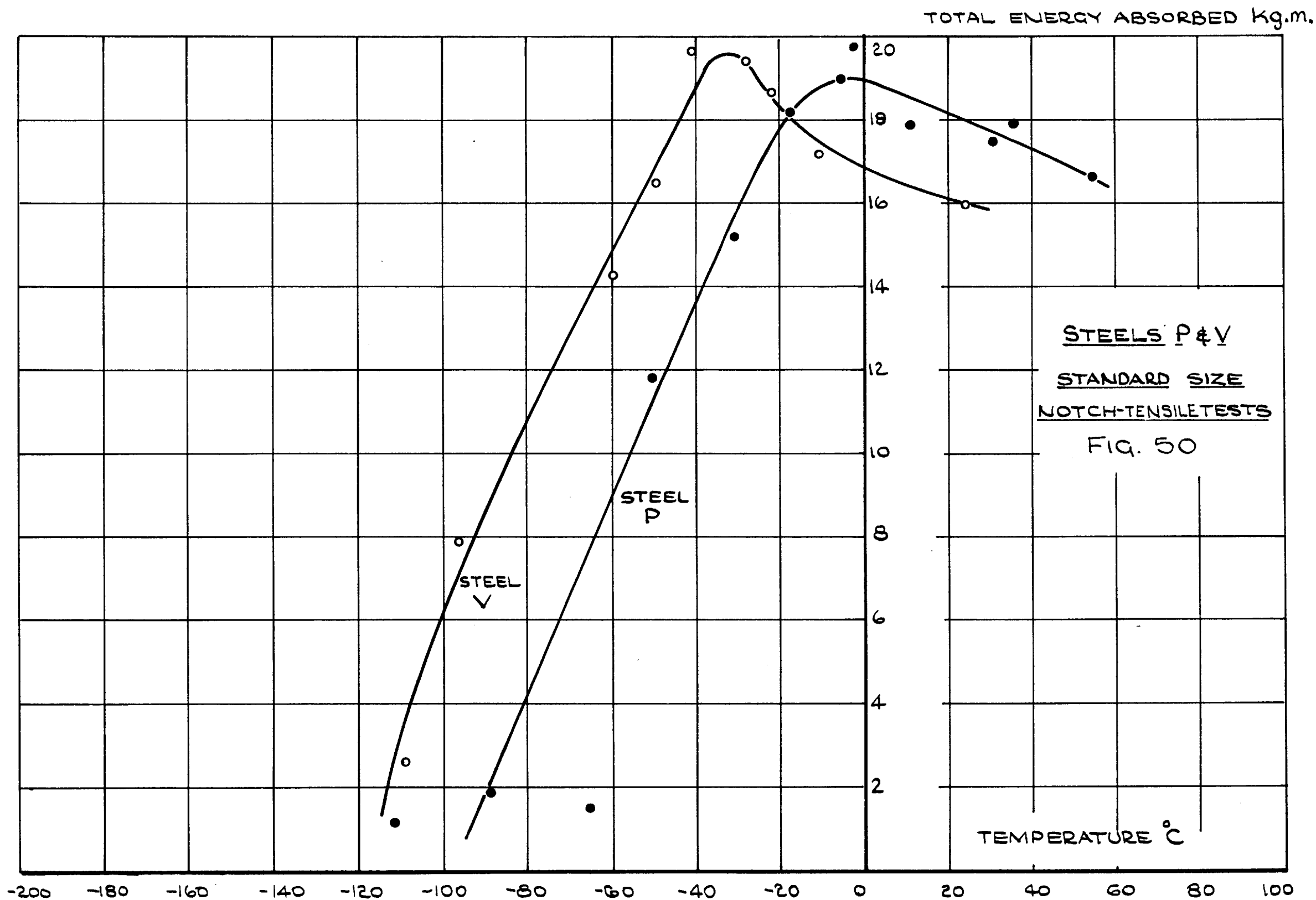












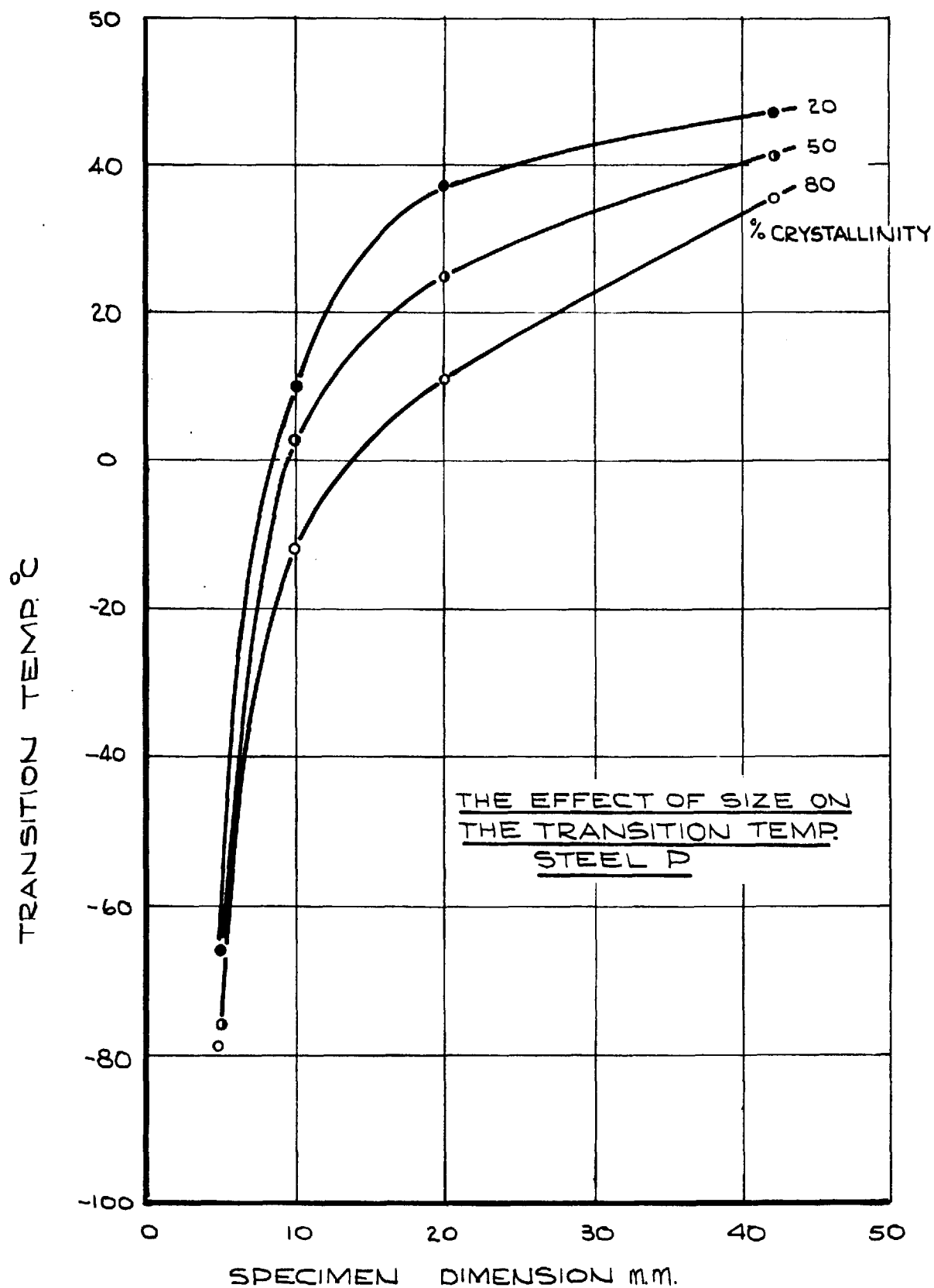


FIG. 51

THE EFFECT OF SIZE ON TRANSITION
TEMP. STEEL 'V'

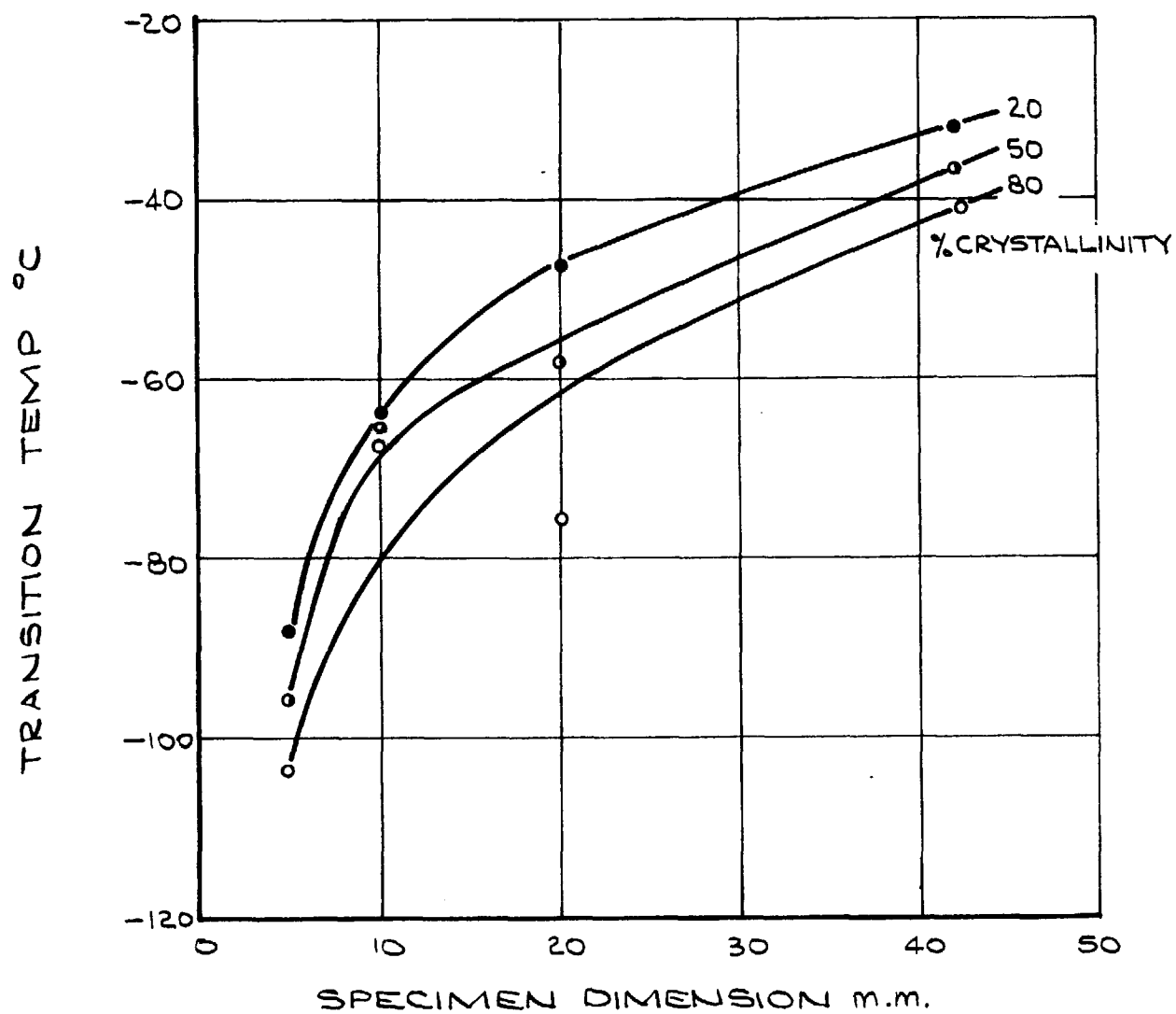


FIG. 52

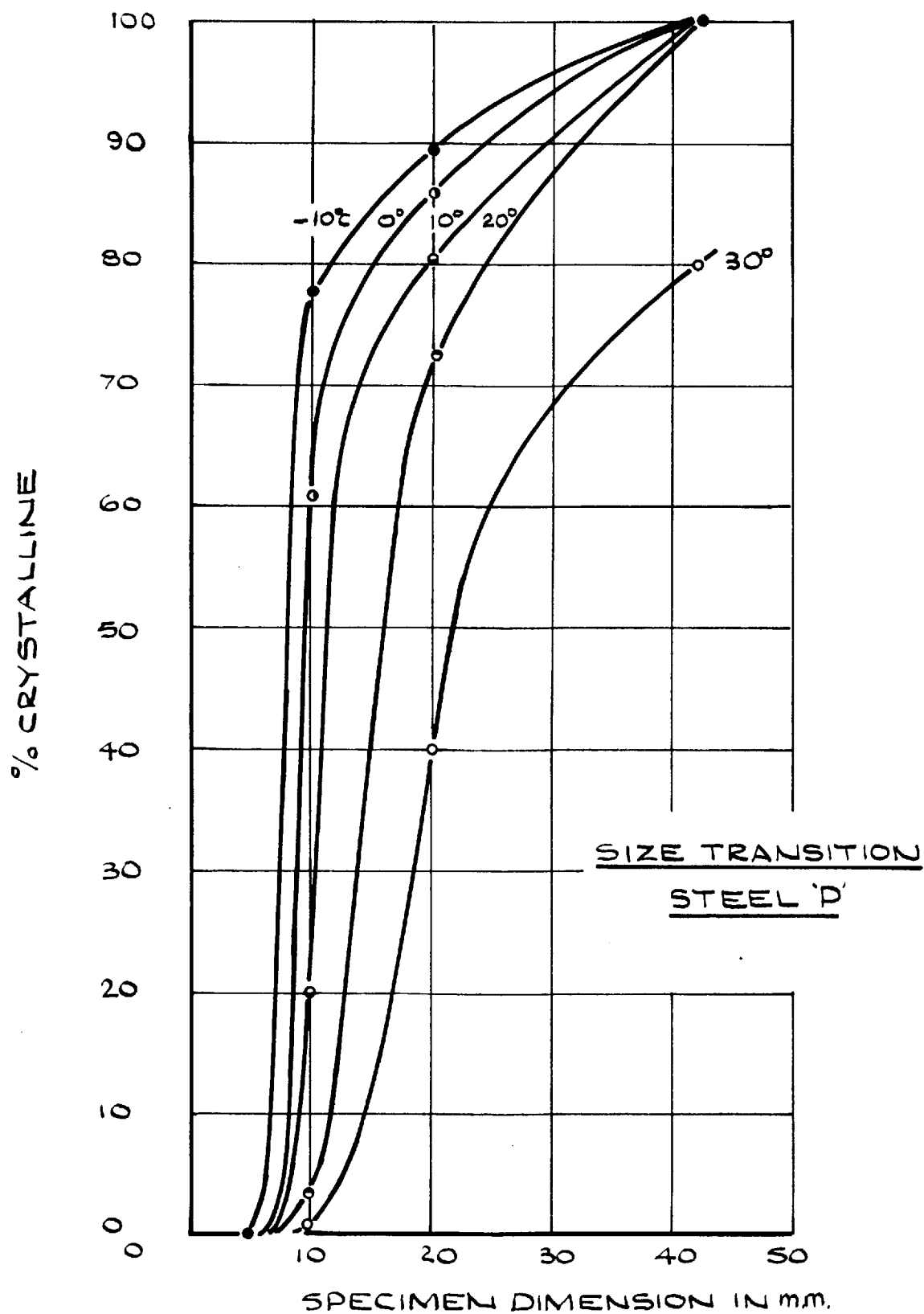


FIG. 53

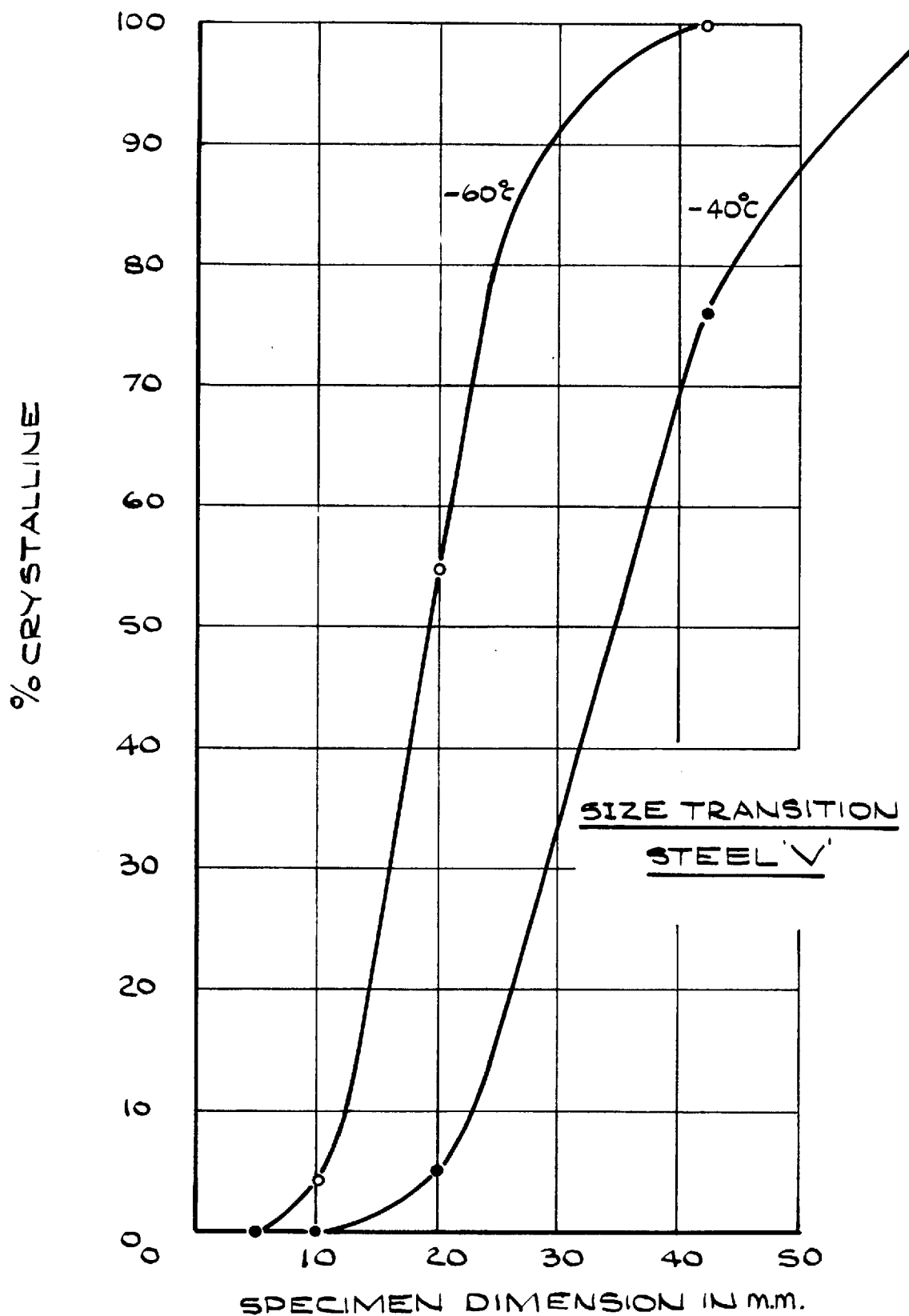
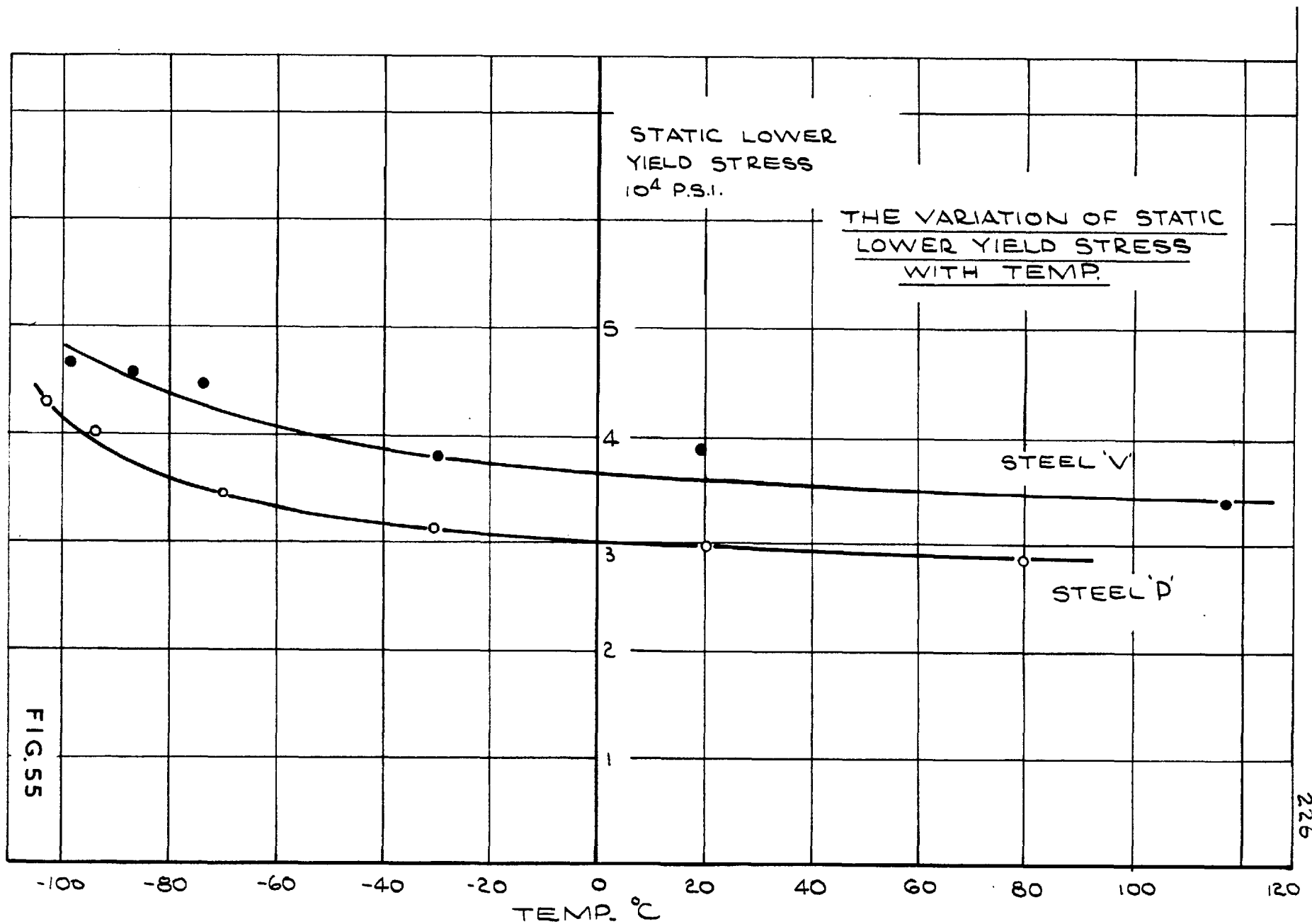


FIG. 54



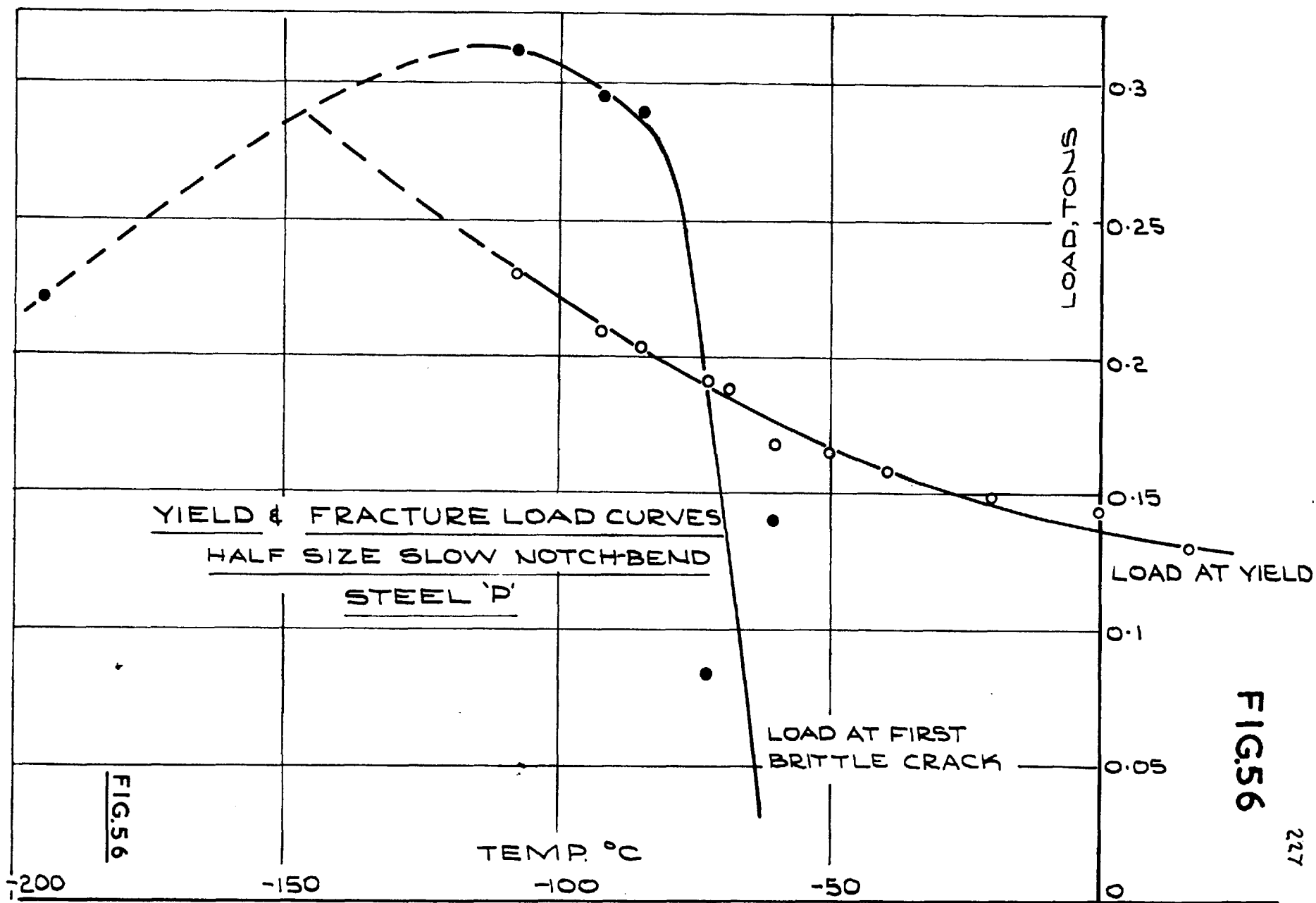


FIG.56

FIG.56

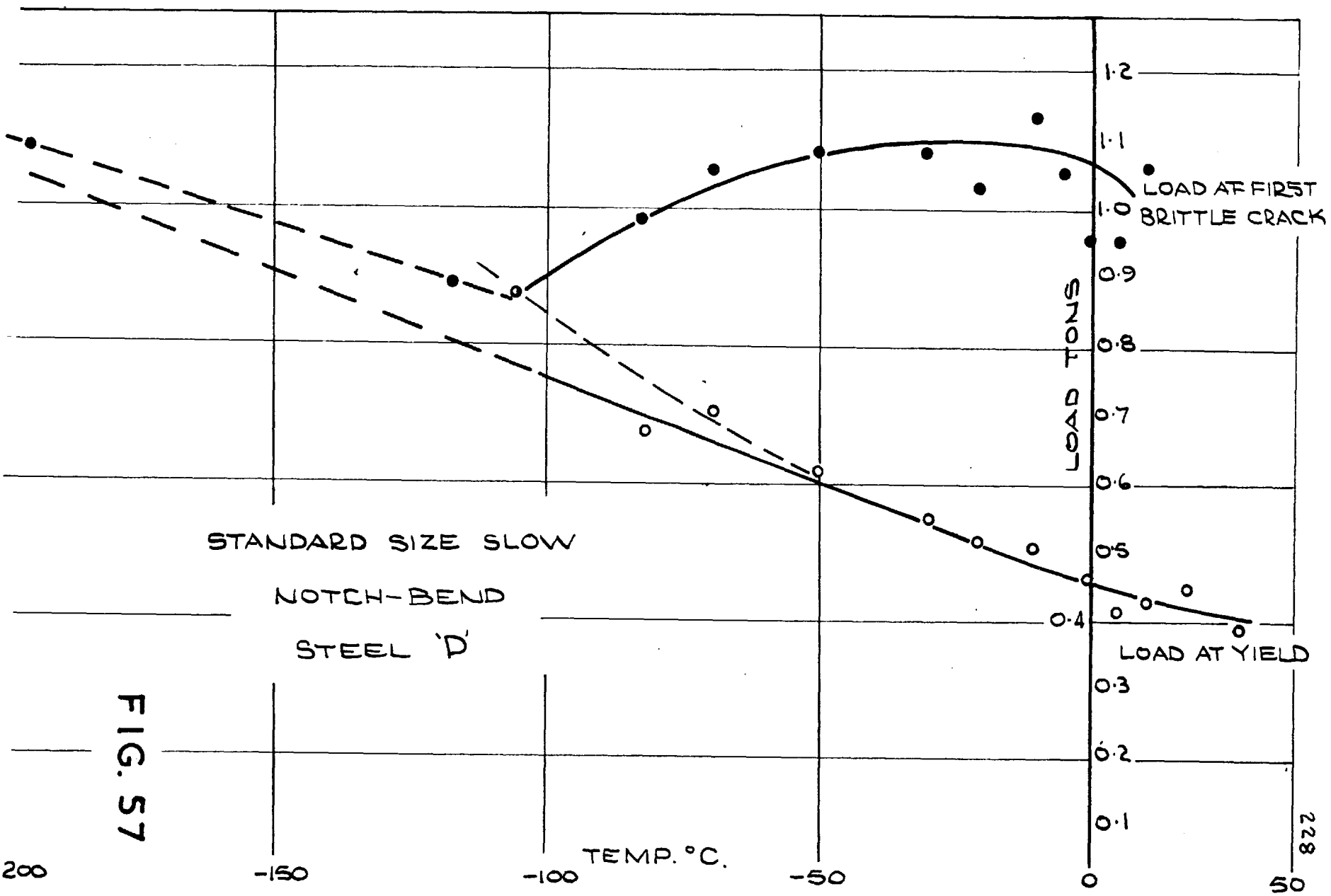


FIG. 57

FIG. 58

DOUBLE SIZE NOTCH-BEND

STEEL 'D'

STRESS
T.D.,
LBS./SQ. IN.

48
44
40
36
32
28
24
20
16
12
8
4

TEMP. °C

YIELD

FIRST BRITTLE CRACK

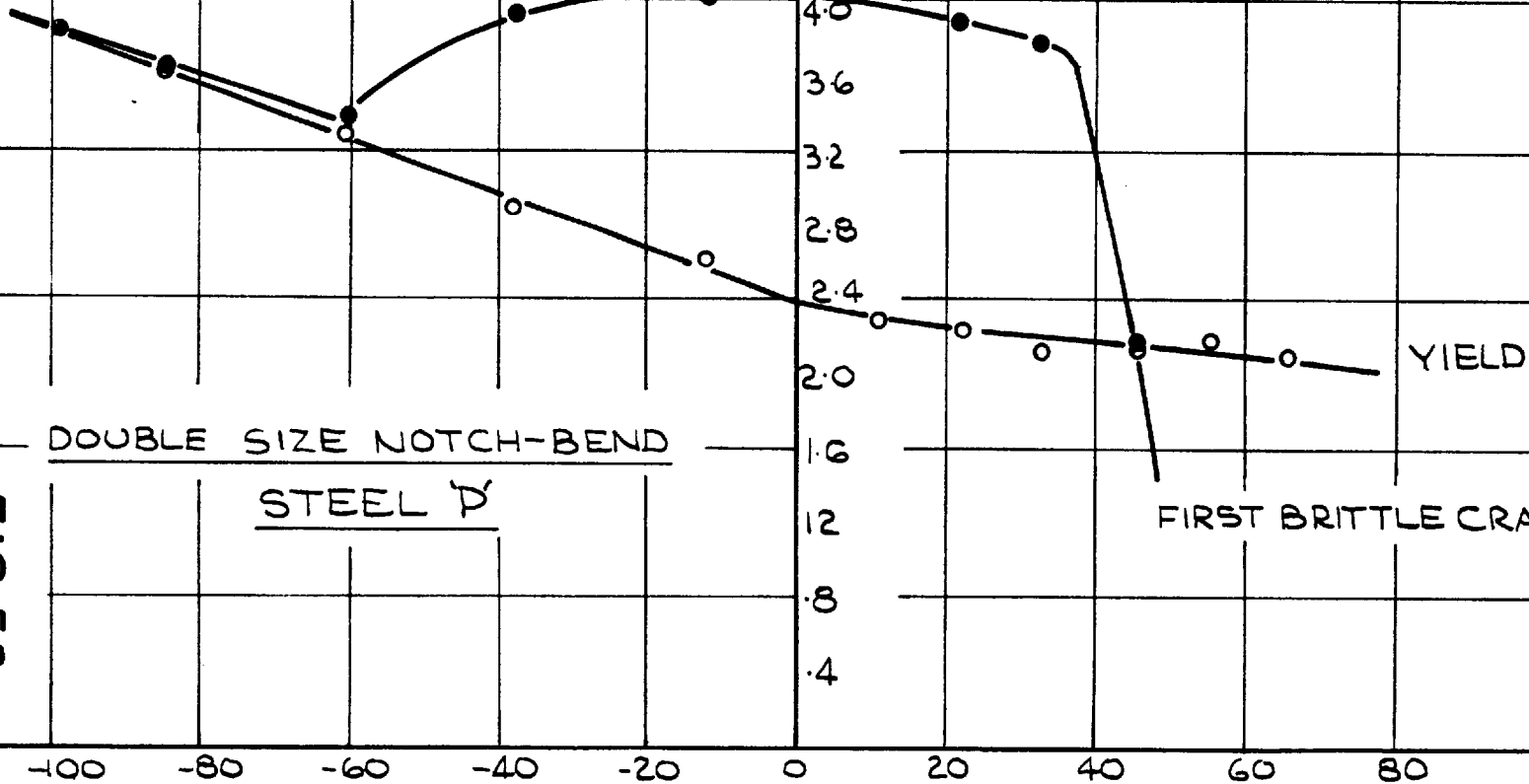


FIG. 59

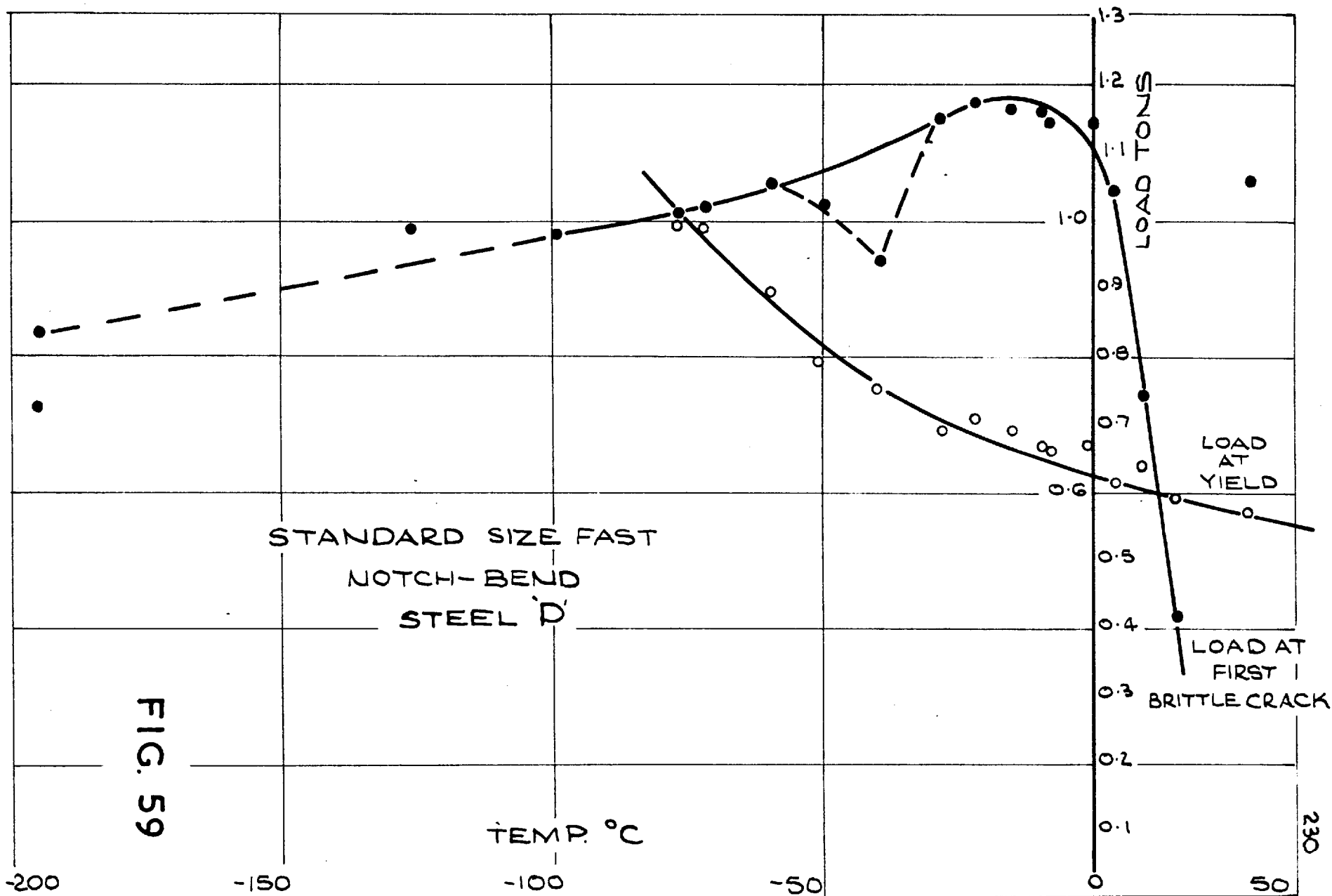
STANDARD SIZE FAST
NOTCH-BEND
STEEL 'D'

TEMP. °C

LOAD TONS

LOAD
AT
YIELD

LOAD AT
FIRST
BRITTLE CRACK



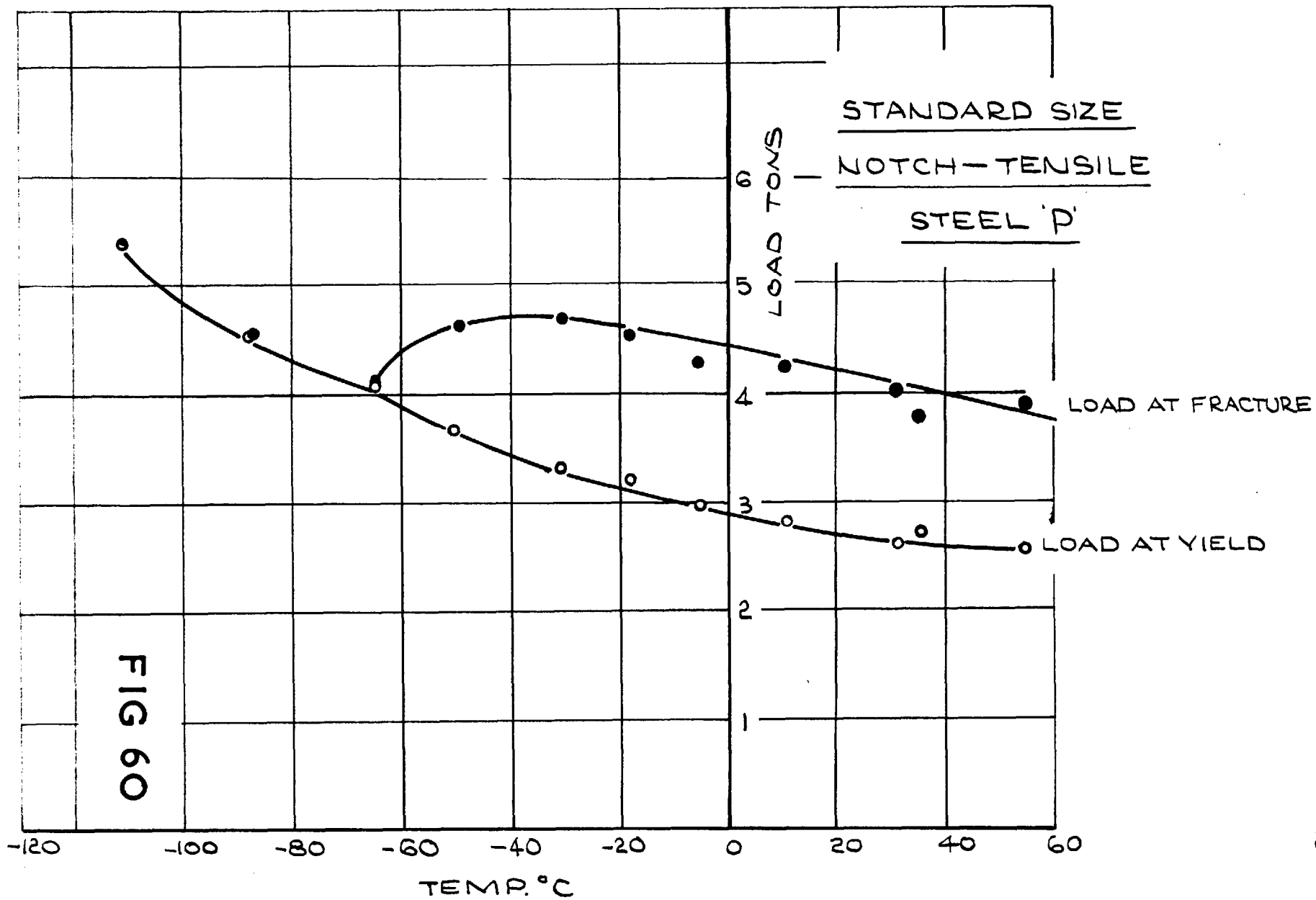
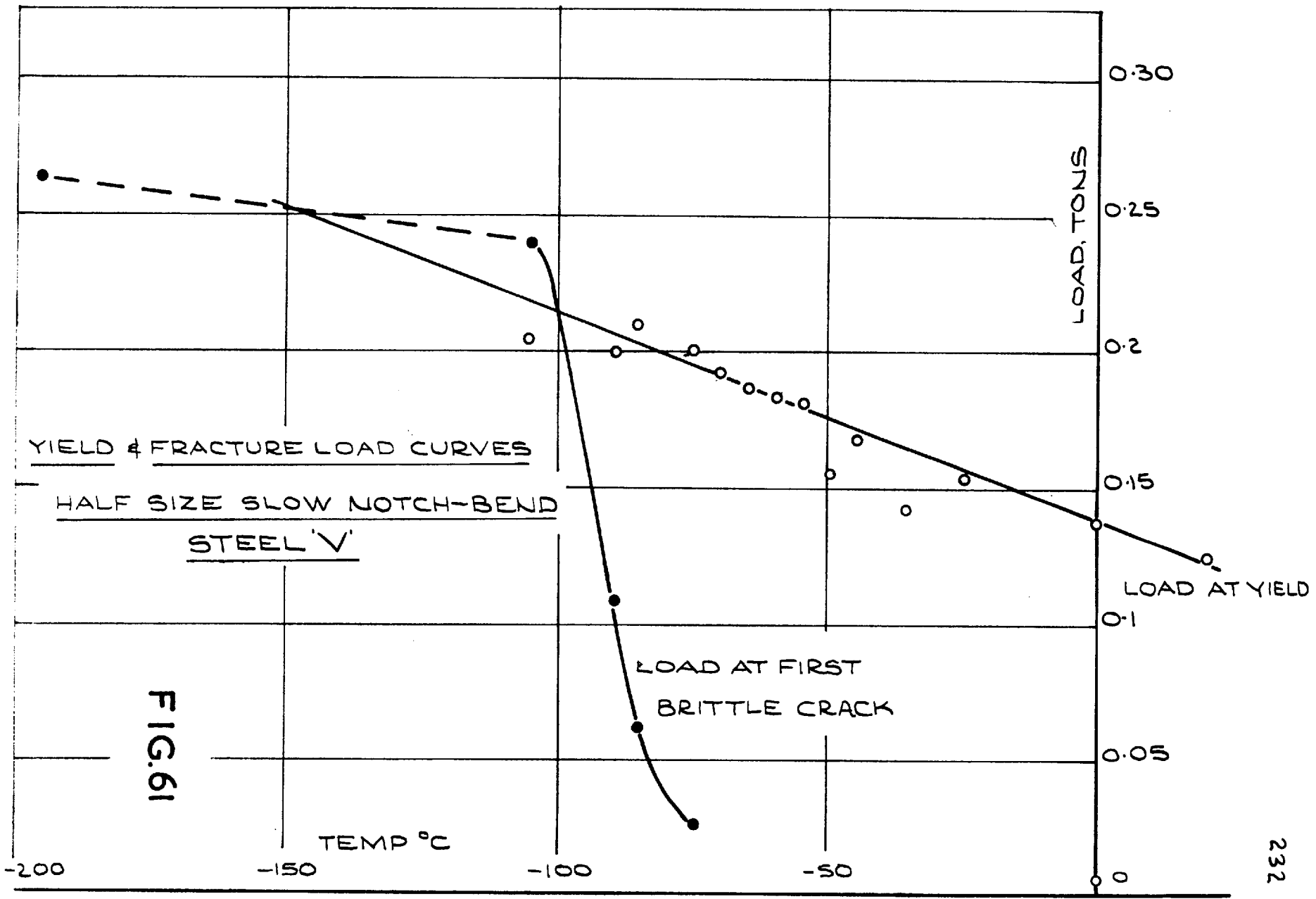


FIG. 61



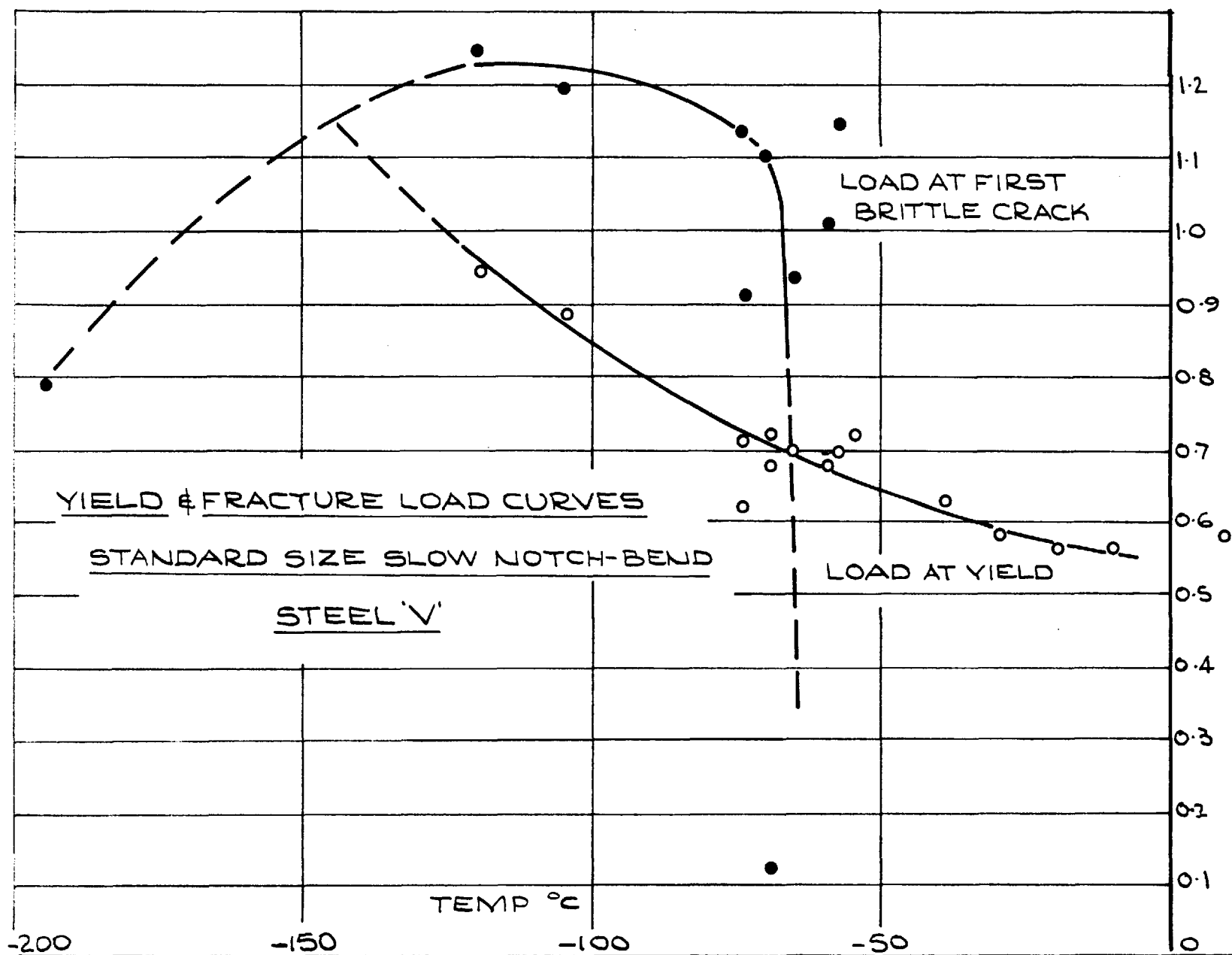


FIG 62

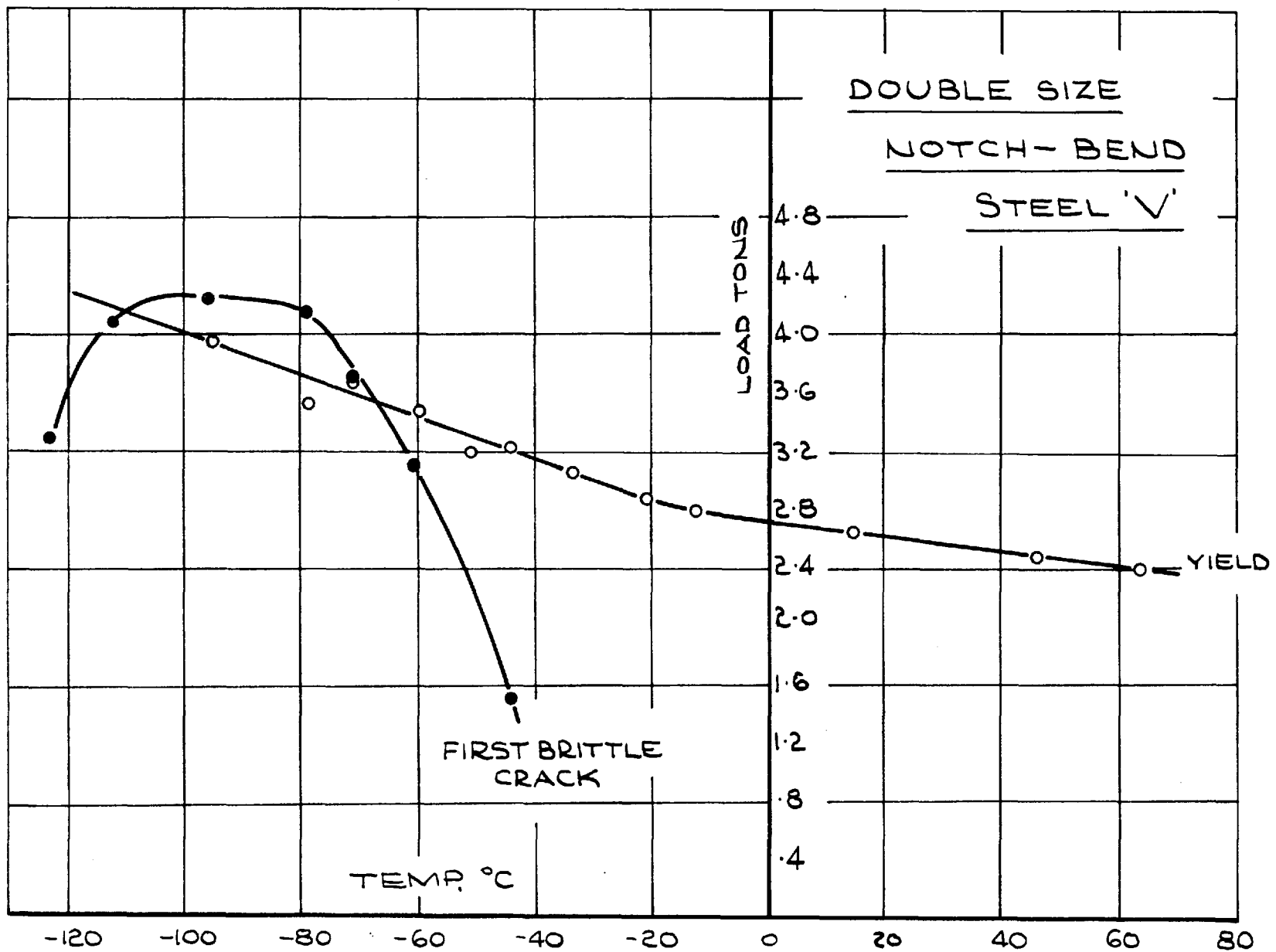


FIG. 63

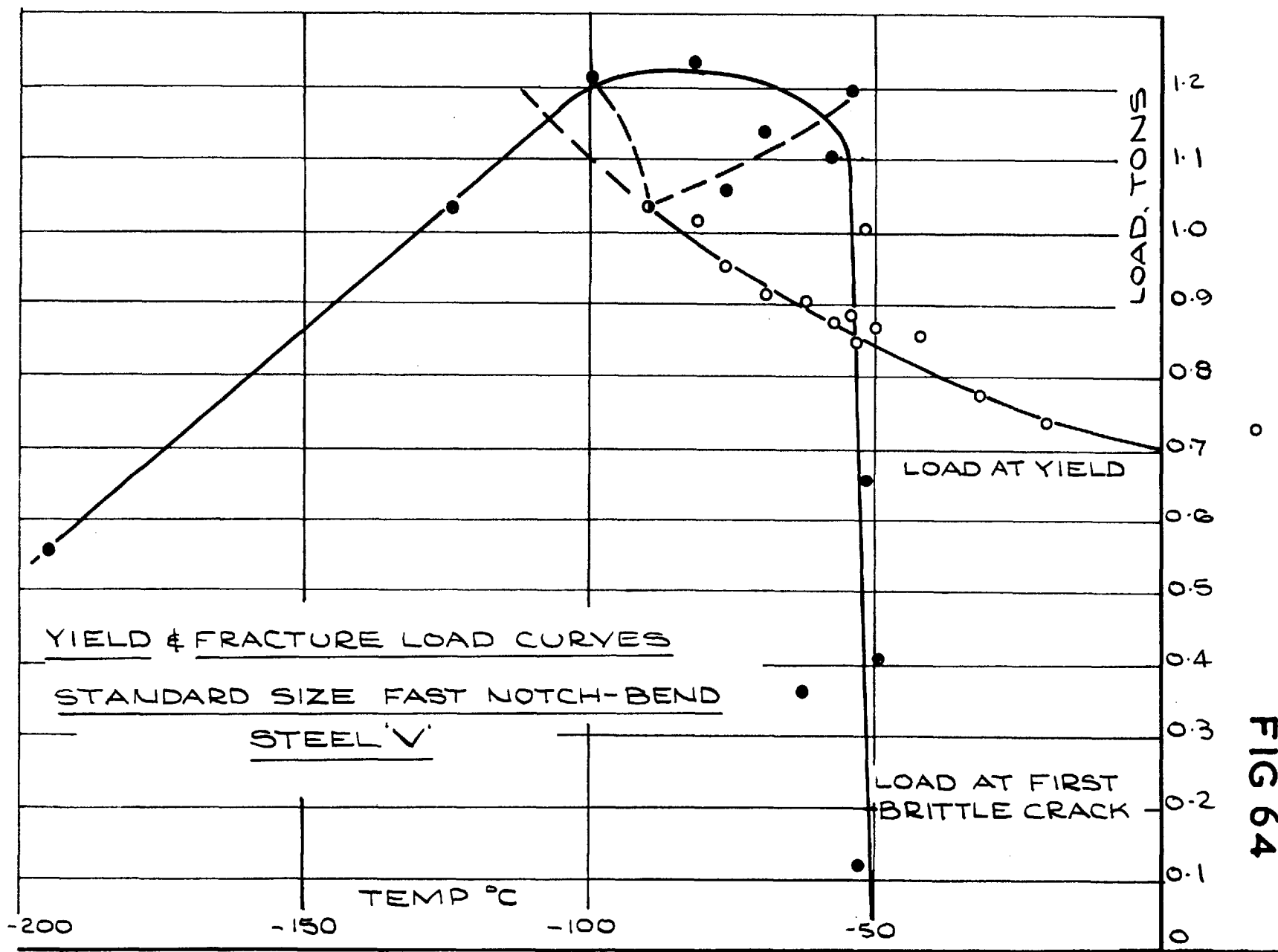


FIG 64

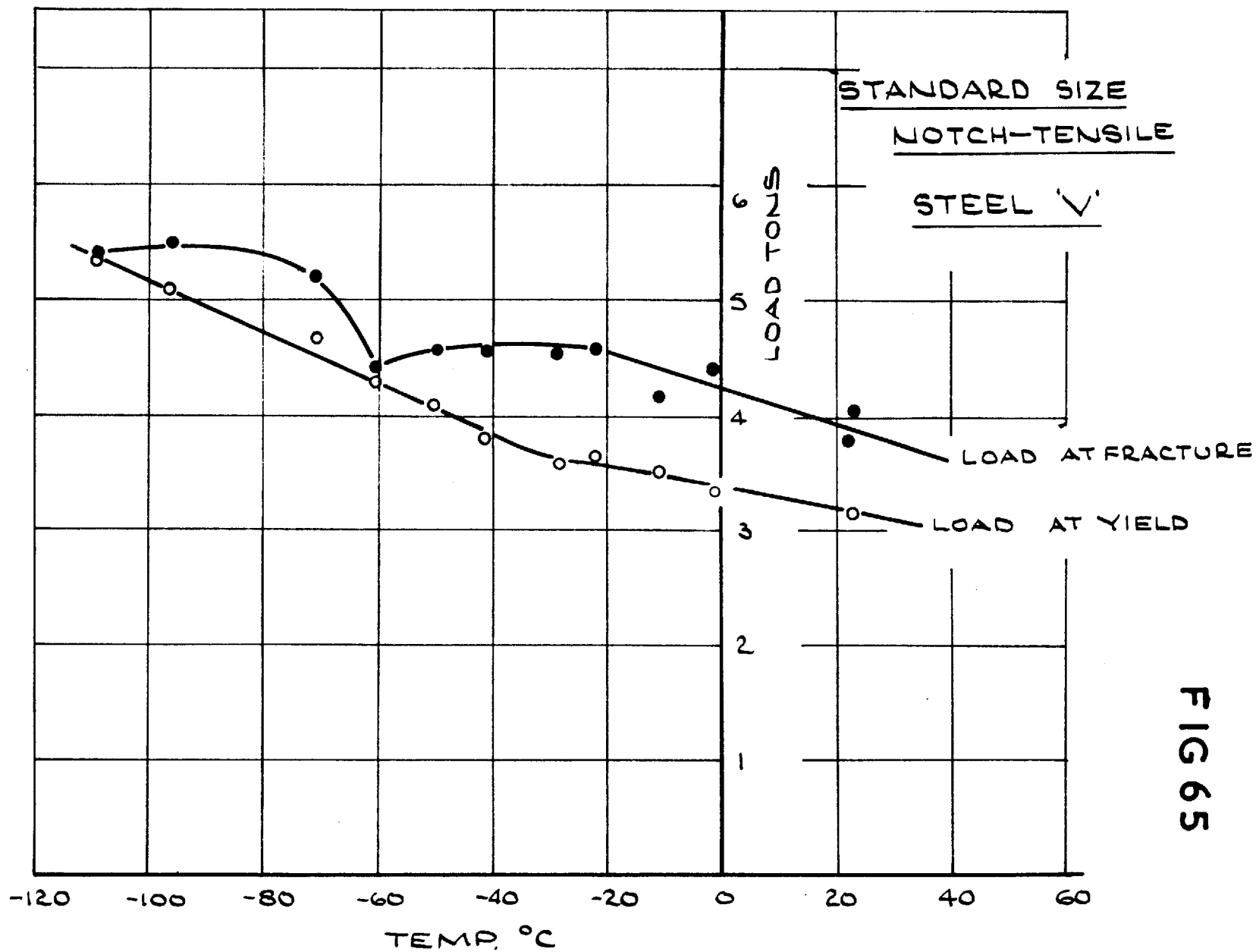
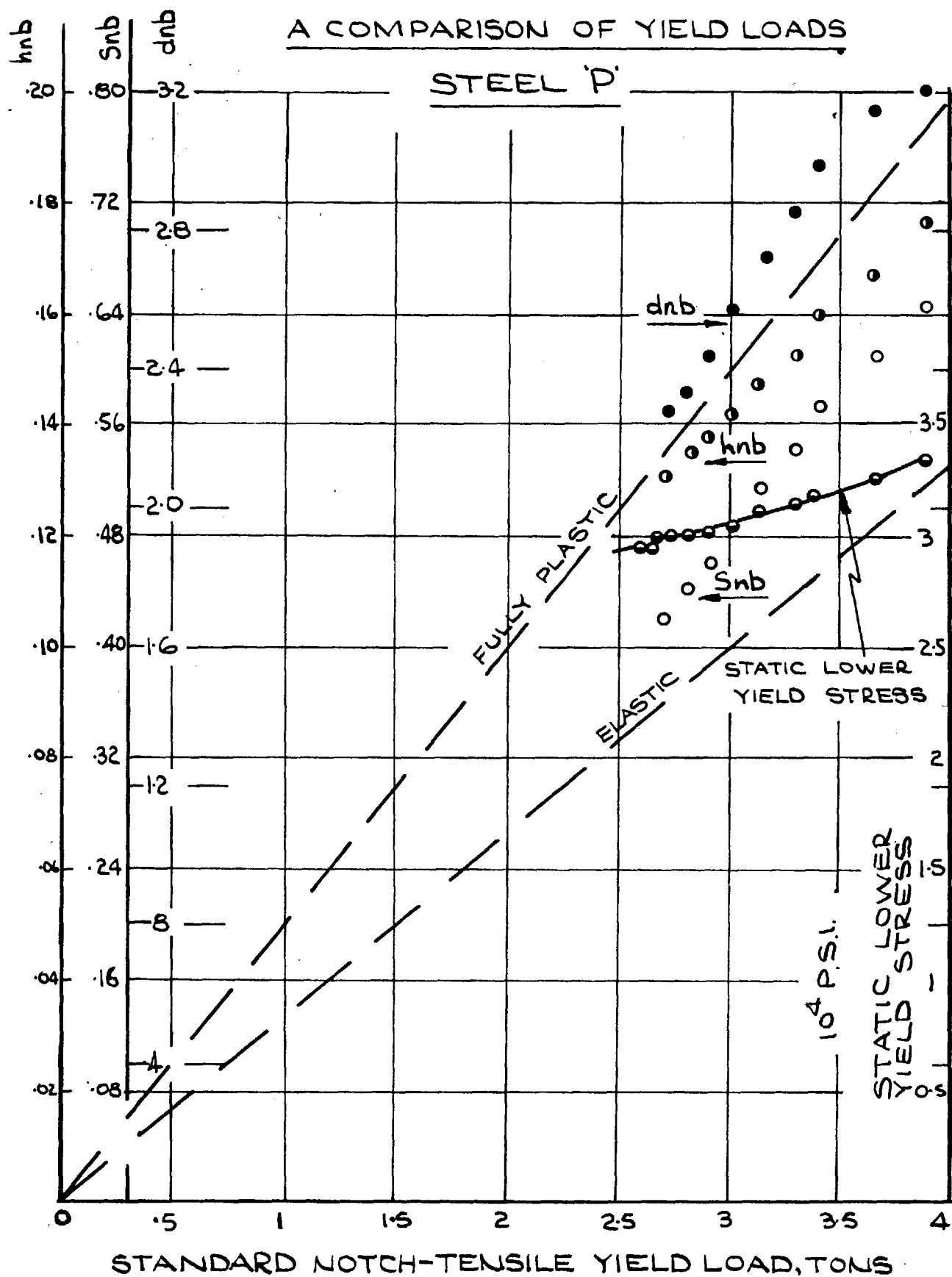


FIG 65

YIELD LOAD TONS

FIG 66



A COMPARISON OF YIELD LOADS

STEEL 'V'

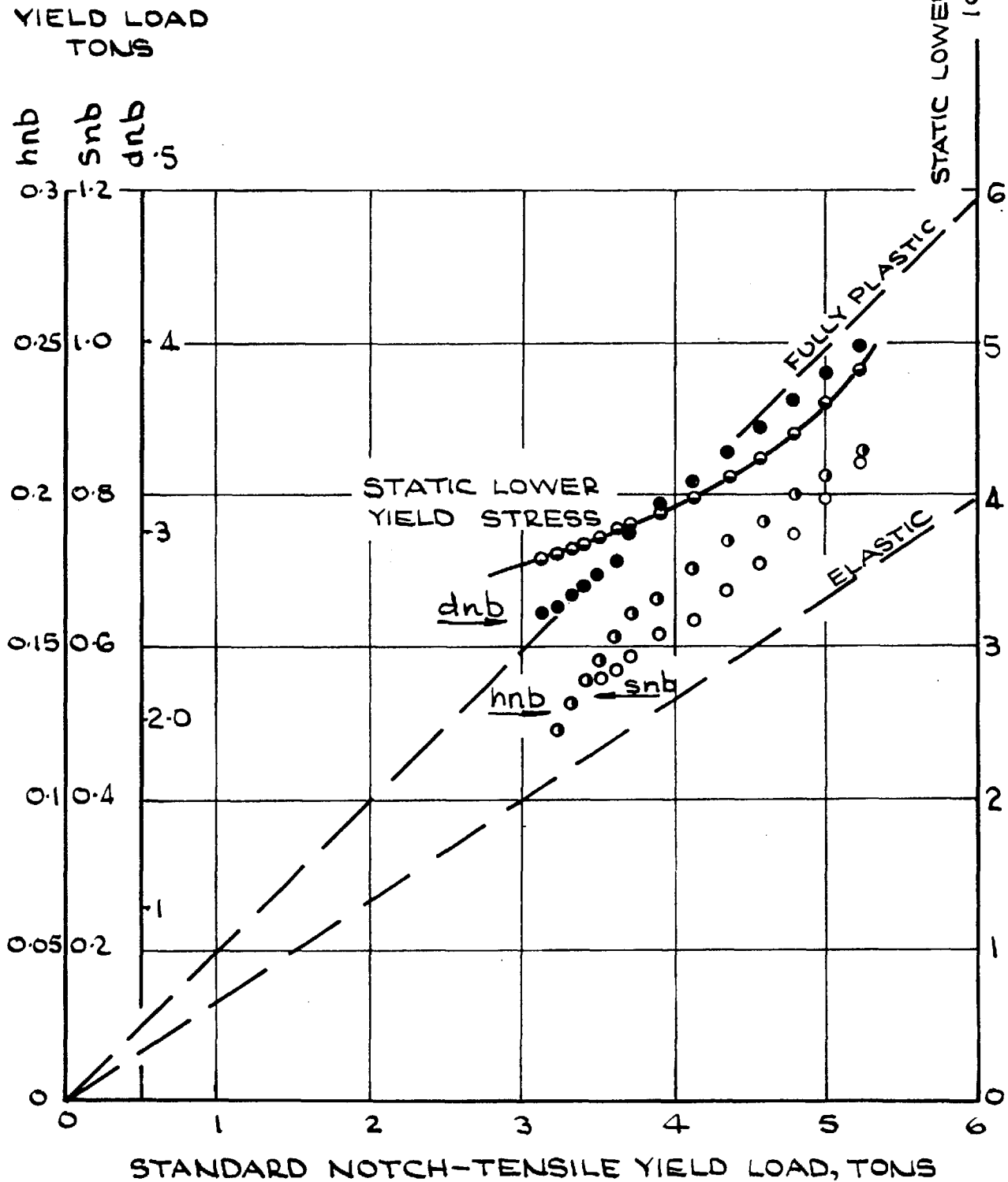
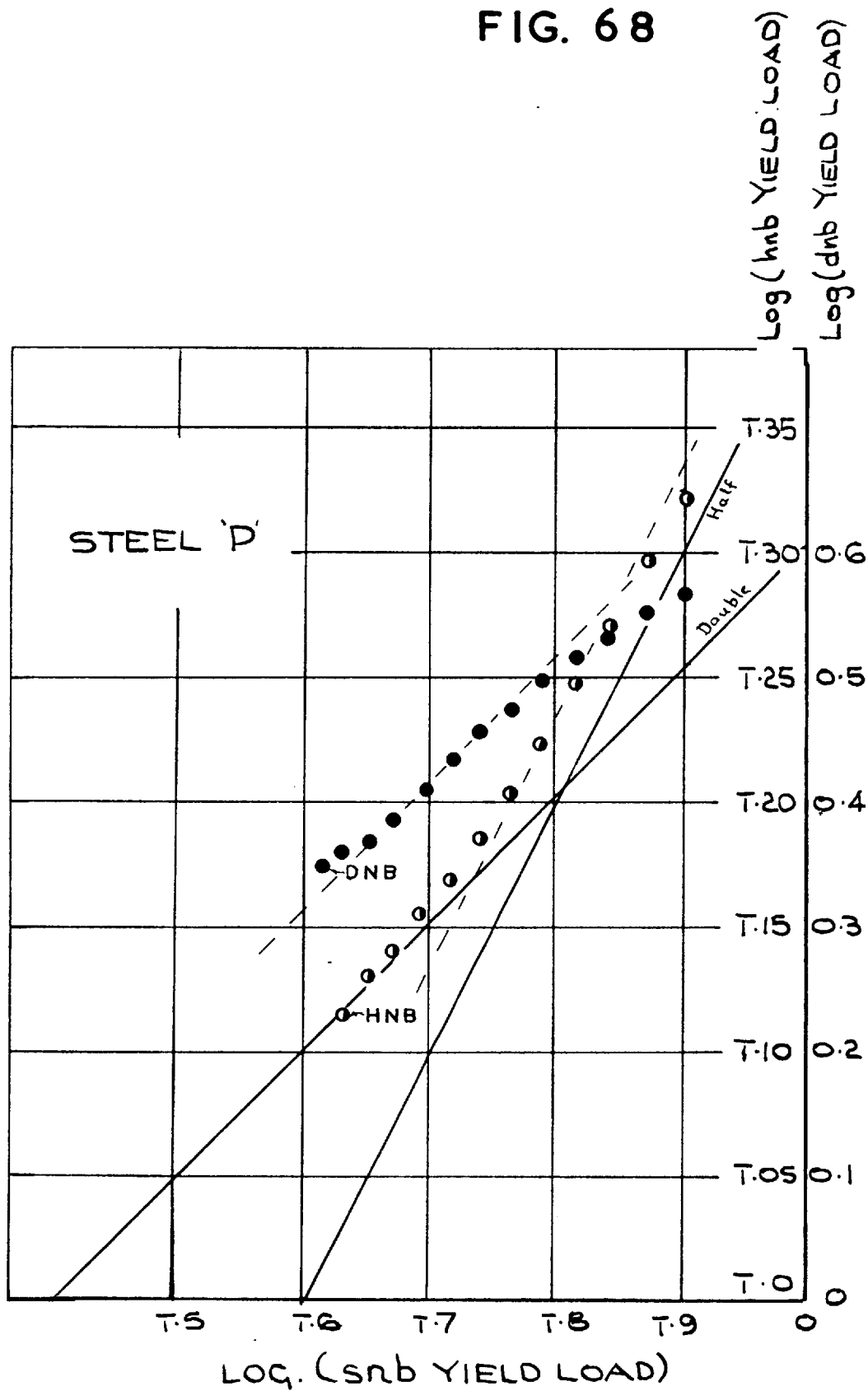
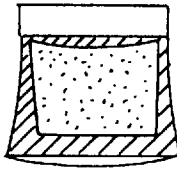


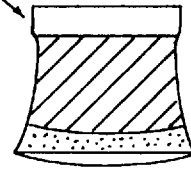
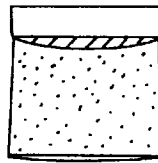
FIG. 68



STANDARD
AND
DOUBLE



HALF



FRACTURE APPEARANCE IN THE SLOW NOTCH BEND TESTS

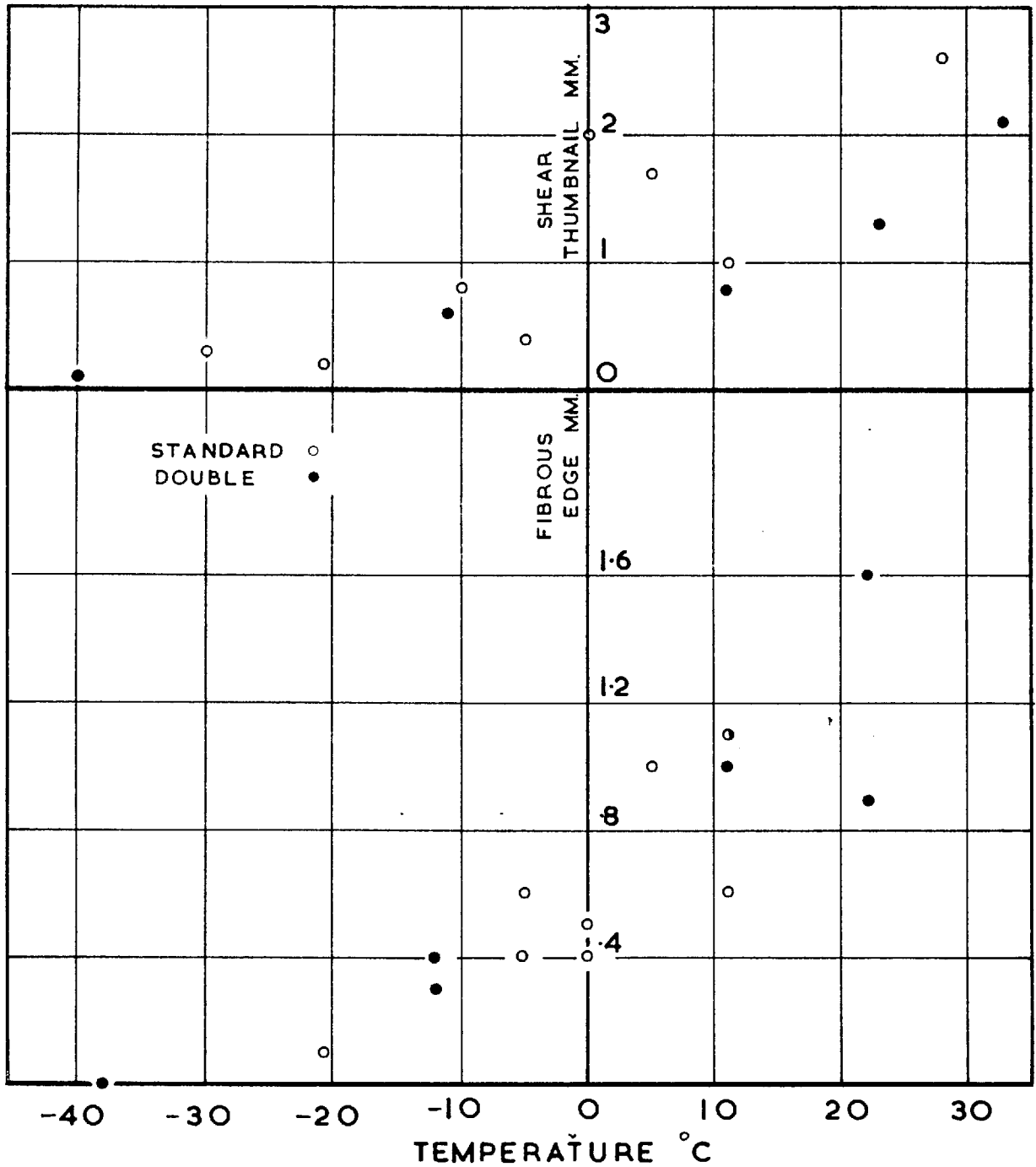


FIG. 70

APPENDIX 'A'TABLES OF RESULTS

<u>Table No.</u>		<u>Page</u>
3	Standard Size Slow Notch Bend Steel 'P'	242
4	Standard Size Slow Notch Bend Steel 'V'	243
5	Standard Size Fast Notch Bend Steel 'P'	244
6	Standard Size Fast Notch Bend Steel 'V'	245
7	Standard Size Impact Notch Bend Steel 'P'	246
8	Standard Size Impact Notch Bend Steel 'V'	247
9	Half Size Slow Notch Bend Steel 'P'	248
10	Half Size Slow Notch Bend Steel 'V'	249
11	Double Size Slow Notch Bend Steel 'P'	250
12	Double Size Slow Notch Bend Steel 'V'	251
13	Standard Size Notch Tensile Steel 'P'	252
14	Standard Size Notch Tensile Steel 'V'	253
15-16	Yield Stress Values Obtained by the Author	254
17-19	Yield Stress Values Obtained by Taylor	255

TABLE 3 STANDARD SIZE SLOW NOTCH BEND STEEL P

Specimen No.	Temp. °C.	% Expansion	Maximum Load		1st Brittle Crack		Yield Load Tons	mm. Fibrous	Energy kg.m.					% Crystalline	% Contraction
			Load Tons	Deflection in.	Load Tons	Deflection in.			Before Maximum Load	After Maximum Load	Before 1st. Brittle Crack	After 1st. Brittle Crack	Total		
20	+28	14.8	1.032	.127	-	-	.39	-	2.66	3.58	-	-	6.24	7	11.1
11	+18.2	15.1	1.064	.125	-	-	.455	8	2.72	4.17	-	-	6.89	1	12.5
22	+11	14.8	1.077	.080	1.070	.089	.435	-	2.84	1.16	3.095	0.91	4.00	50	12.6
23	+5.0	14.0	1.068	.133	0.962	.173	.415	-	2.93	1.51	3.92	0.52	4.44	45	12.9
12	0	12.0	1.07	.065	1.07	.065	.86	0.7-8	1.72	0.95	1.72	0.95	2.67	60	-
13	0	11.4	1.102	.131	.962	.177	.465	2.0	2.94	1.74	4.04	0.64	4.68	55	12.9
24	-5.0	11.9	1.064	.117	1.064	.117	.46	-	2.535	0.595	2.535	.595	3.13	65	6.75
14	-10	9.5	1.114	.124	1.114	.124	.51	1.0	2.78	0.17	2.78	0.17	2.95	80	10.1
15	-20.6	7.3	1.038	.092	1.038	.092	.52	0.4	1.92	0.31	1.92	0.31	2.23	85	8.6
16	-30	7.7	1.09	.097	1.09	.097	.55-6	0.2	2.155	.255	2.155	.255	2.410	90	6.6
17	-50.2	5.8	1.09	.085	1.09	.085	.62	0	1.85	.07	1.85	0.07	1.92	95	5.3
18	-70	4.1	1.06	.058	1.06	.058	.71	0	1.24	.015	1.24	.015	1.255	100	3.8
301	-82	3.6	0.988	0.052	0.988	.052	.68	0	1.05	0	1.05	0	1.05	98	3.5
302	-106	1.3	0.884	0.0105	0.884	.0105	.884	0	0.17	0	0.17	0	0.17	99	0.04
306	-118	-	0.900	0.0095	0.900	.0095	.900	0	0.15	0	0.15	0	0.15	100	-
307*	-196	0.25	1.094	0.0065	1.094	.0065	-	0	0.13	0	0.13	0	0.13	100	0

* Fracture occurred on Linear Section of Load-Time Curve

TABLE 4 STANDARD SIZE SLOW NOTCH BEND STEEL V

Specimen No.	Temp. °C.	% Expansion	Maximum Load		1st Brittle Crack		Load at Yield	mm. Fibrous	Energy kg.m.					% Crystalline
			Load Tons	Deflection in.	Load Tons	Deflection in.			Before Maximum Load	After Maximum Load	Before 1st. Brittle Crack	After 1st. Brittle Crack	Total	
138	+9	14.73	1.01	.089	-	-	0.58	8	1.96	4.45	-	-	6.41	0
139	-10	15.33	1.098	.099	-	-	0.565	8	2.28	4.78	-	-	7.06	0
140	-20	15.4	1.093	.102	-	-	0.565	8	2.34	4.71	-	-	7.05	0
141	-30	15.87	1.109	.116	-	-	0.580	8	2.70	4.37	-	-	7.07	0
142	-40	15.8	1.130	.107	-	-	0.63	8	2.58	4.52	-	-	7.10	0
143	-50	15.94	-	-	-	-	-	8	-	-	-	-	-	2
144	-55	15.5	1.137	.095	-	-	0.72	8	2.34	4.52	-	-	6.86	$\frac{1}{2}$
146	-57.5	9.64	1.197	.112	1.145	.131	0.70	1.3	2.80	0.68	3.42	0.06	3.48	90
145	-60	10.22	1.136	.097	1.01	.138	0.68	2.2	2.39	1.24	3.55	0.08	3.63	70
147	-66	12.2	1.158	.097	0.94	.176	0.70	2.9	2.41	2.28	4.64	0.05	4.69	65
148	-70	16.5	1.171	.105	0.12	.407	0.68	6.8	2.67	4.33	6.97	0.03	7.00	15
151	-70	5.39	1.10	.068	1.10	.068	0.725	0.5	1.61	0.32	1.61	0.32	1.93	95
149	-74	12.58	1.184	.101	0.912	.177	0.714	3.5	2.64	2.14	4.75	0.03	4.78	60
150	-74.8	8.88	1.149	.1095	1.136	.118	0.625	1.5	2.84	0.35	3.10	0.09	3.19	90
306	-105	8.1	1.198	.070	1.198	.070	.887	0.4	1.83	0	1.83	0	1.83	95
307	-120	8.1	1.248	.066	1.248	.066	.946	0.4	1.74	0	1.74	0	1.74	95
309*	-196	0.5	0.794	.0065	0.794	.0065	-	0.	0.06	0	0.06	0	0.06	100

* Fracture occurred on linear section of load-time curve.

TABLE 5 STANDARD SIZE FAST NOTCH BEND STEEL P

Specimen Number	Temp. °C.	% Expansion	Maximum Load		1st Brittle Crack		Load at Yield	mm. Fibrous	Energy kg.m.					% cryst-alline
			Load Tons	Deflection in.	Load Tons	Deflection in.			Before Maximum Load	After Maximum Load	Before 1st Brittle Crack	After 1st Brittle Crack	Total	
187	80.5	14.68	1.099	.1215	-	-	.500	8	2.44	4.10	-	-	6.54	0
186	49.4	16.1	1.065	.121	-	-	.536	8	2.62	4.30	-	-	6.92	< 1
185	29.4	15.1	1.09	.128	1.065	.155	.575	8	2.88	3.56	3.68	2.76	6.44	5
157	16	14.8	1.11	.127	0.478	.3115	.598	8	2.80	5.31	6.75	1.36	8.11	7
158	10	14.45	1.14	.143	0.749	.271	.641	8	3.32	4.99	6.60	1.71	8.31	12
159	3.8	10.97	1.13	.139	1.053	.194	.616	1.3	3.16	2.10	4.67	0.59	5.26	70
160	-0.5	10.3	1.148	.138	1.148	.138	.675	.6	3.22	0.50	3.22	0.50	3.72	80
161	-8	10.24	1.159	.132	1.15	.140	.661	.7	3.10	0.49	3.34	0.25	3.59	90
174	-9.5	8.30	1.165	.136	1.165	.136	.672	.4	3.26	0.14	3.26	0.14	3.40	95
162	-15	8.73	1.185	.129	1.172	.140	.698	.4	3.14	0.47	3.51	0.10	3.61	95
163	-21.5	8.94	1.19	.129	1.179	.1406	.715	.9	3.15	0.43	3.51	0.07	3.58	92
175	-28.5	7.22	1.155	.1105	1.155	.1105	.696	.3	2.59	0.06	2.59	0.06	2.65	97
176	-40	4.22	0.965	.051	0.95	.0536	.760	trace	0.99	0.07	1.04	0.02	1.06	99
178	-50.5	4.27	1.033	.056	1.033	.056	.798	Trace	1.15	0.05	1.15	0.05	1.20	99
179	-60	3.08	1.06	.043	1.06	.043	.901	0	0.97	0.05	0.97	0.05	1.02	100
181	-72.5	1.65	1.025	.023	1.025	.023	.995	0	0.42	0.01	0.42	0.01	0.43	100
182	-77	.254	1.017	.018	1.017	.018	1.0	0	0.30	0.01	0.30	0.01	0.31	100
184	-155	.178	.840	Not known	-	Not known	-	0	-----	-----	Not Known	-----	-----	100
183	-196	0	.694	Not known	-	Not known	-	0	-----	-----	Not Known	-----	-----	100
193	-100	.97	.990	.017	.990	.017	-	0	0.29	0	0.29	0	0.29	100
194	-127	.79	.994	Not known	.994	Not known	-	0	-----	-----	Not Known	-----	-----	100
190	-196	.18	.84	Not known	.84	Not known	-	0	-----	-----	Not known	-----	-----	100
192	-196	.15	.719	.0061	.719	-	-	0	0.056	0	0.056	0	0.056	100

TABLE 6 STANDARD SIZE FAST NOTCH BEND STEEL V

Specimen Number	Temp. °C.	% Expansion	Maximum Load		1st Brittle Crack		Load at Yield	mm. Fibrous	Energy kg.m.					% Crystalline
			Load Tons	Deflection in.	Load Tons	Deflection in.			Before Maximum Load	After Maximum Load	Before 1st Brittle Crack	After 1st Brittle Crack	Total	
190	77.4	14.0	0.986	.0869	-	-	.547	8	1.80	4.14	-	-	5.94	0
188	49.4	15.01	1.035	.0951	-	-	.604	8	2.01	4.31	-	-	6.32	0
152	16.4	14.79	1.13	.115	-	-	.734	8	2.60	6.39	-	-	8.99	<1
153	-15	14.86	1.122	.117	-	-	.713	8	2.72	5.19	-	-	7.91	<1
158	-20	14.9	1.126	.0967	-	-	.745	8	2.25	5.10	-	-	7.35	0
168	-32	15.1	1.17	.108	-	-	.78	8	2.06	5.10	-	-	7.16	0
169	-42	15.0	1.20	.0992	-	-	.865	8	2.44	5.35	-	-	7.79	$\frac{1}{4}$
171	-50	16.3	1.22	.0973	.41	.300	.875	8	2.53	6.94	6.77	0.70	7.47	6
174	-52	14.3	1.21	.095	.66	.232	1.01	4.1	2.43	3.29	5.65	0.08	5.72	50
175	-53.5	15.61	1.20	.0956	.12	.385	.854	7.1	2.51	5.00	7.39	0.12	7.51	8
172	-55	8.45	1.20	.09-.11	1.20	.113	.89	1.0	2.97	0.26	2.97	0.26	3.23	92
178	-58	14.05	1.17	.093	1.11	.114	.88	8	2.46	3.80	2.99	3.27	6.26	15
181	-63	14.84	1.23	.104	.365	.304	.91	8	2.36	4.94	6.87	0.43	7.30	12
179	-70	8.54	1.195	.087	1.145	.0953	.92	1.75	2.26	0.97	3.17	0.06	3.23	90
182	-76.8	9.6	1.215	.0765	1.07	.1355	.96	2.2	1.93	1.83	3.70	0.06	3.76	88
183	-82	5.96	1.27	.0762	1.24	.082	1.02	0.7	2.01	0.26	2.22	0.05	2.27	96
170	-196	0.18	0.56	$\div$.0065	0.56	$\div$.0065	-	-	$\div$ 0.04	0	$\div$ 0.04	0	$\div$ 0.04	100
191	-97	1.32	1.045	.013	1.045	.013	1.045	-	0.31	0	0.31	0	0.31	100
187	-100	2.16	1.220	Not known	1.22	Not known	-	-	Not known					100
192	-125	1.22	1.04	.015	1.04	-	-	-	0.24	0	0.24	0	0.24	100

TABLE 7 STANDARD SIZE IMPACT NOTCH BEND STEEL P

Specimen No.	Temp. °C.	% Expansion	Maximum Load		1st Brittle Crack		Energy kg.m.				Integrated area m/c energy %	Load at 1st. Reversal	% Crystalline
			Load Tons	Deflection in.	Load Tons	Deflection in	Before 1st. Brittle Crack	After 1st. Brittle Crack	TOTAL	Corrected m/c Value			
99	+91	18.5	1.16	.028	-	-	-	-	6.38	9.89	64.6	.707	1
98	+83	18.8	0.895	.020	-	-	-	-	6.71	9.90	67.8	.552	3
96	+68	17.7	1.002	.0138	-	-	-	-	6.16	9.60	64.3	.766	7
97	+60	17.1	1.010	.0167	0.984	.1225	2.72	3.08	5.80	8.91	65.0	.702	20
94	+48	13.72	1.043	.0128	0.980	.0930	2.13	1.97	4.10	6.21	66.0	.678	70
95	+32	11.2	1.195	.025	0.981	.076	1.56	1.66	3.22	5.06	63.6	.515?	75
100	+9	4.04	1.108	.013	1.025	.023	0.52	0.46	0.98	1.89	51.9	.778	90
101	0	3.42	1.164	.013	0.999	.022	0.43	0.22	0.65	1.29	50.4	.740	92
102	-21	2.04	1.188	.012	1.188	.012	0.28	0.09	0.37	0.63	58.8	.750	97
103	-55	1.05	1.039	.014	1.039	.014	0.27	0	0.27	0.41	65.8	.761	99
104	-72	2.34	0.603	.010	0.600	.013	0.19	0	0.19	0.55	34.6	.442	100
106	÷ -100	0.82	0.868	.011	0.868	.011	0.21	0	0.21	-	-	.778	100

STEEL V
TABLE 8 STANDARD SIZE IMPACT NOTCH BEND

Specimen Number	Temp. °C.	Expansion	Maximum Load	Reflection at Maximum Load	Load at 1st. Brittle Crack	Reflection at 1st. Brittle Crack	Energy kg.m.				Crystalline	Integrated Area m/c energy %	Load at 1st. Reversal
							Total	Before 1st. Brittle Crack	After 1st. Brittle Crack	Corrected m/c Value			
184	+25	-	1.52	-	-	-	7.96	-	-	-	0	-	-
216	+9	17.72	1.929	.917	-	-	2.10	-	-	9.42	0	65.4	.814
217	0	18.7	1.197	.925	-	-	6.55	-	-	9.82	trace	69.7	.266
243	-5	17.8	1.255	.926	-	-	5.96	-	-	9.20	5	64.9	.814
246	-10	15.5	1.541	.917	0.916	.120	4.40	2.94	1.55	7.09	45	63.4	.769
244	-16	15.0	1.545	.925	0.944	.110	4.22	2.21	1.61	6.72	60	60.9	.306
245	-22	7.72	1.933	.919	1.953	.919	6.94	0.28	0.36	1.49	75	63.1	.793
246	-40	1.91	1.27	.911	1.21	.921	9.41	0.30	0.11	9.77	90	53.3	.785
247	-75	1.32	1.091	.911	1.091	.911	9.22	0.22	0	9.45	98	42.9	.775

TABLE 9 HALF SIZE SLOW NOTCH BEND STEEL P

Specimen Number	Temp. °C.	% Expansion	Maximum Load		1st. Brittle Crack		Load at Yield	mm. Fibrous	Energy kg.m.					% crystalline
			Load Tons	Deflection in.	Load Tons	Deflection in.			Before Maximum Load	After Maximum Load	Before 1st Brittle Crack	After 1st Brittle Crack	Total	
I	+80	19.5	-	-	-	-	-	4	-	-	-	-	-	0
E	+16.5	20.4	.254	.085	-	-	.132	4	.451	1.182	-	-	1.733	0
F	- 0.5	13.7	.281	.088	-	-	.144	4	.482	.708	-	-	1.190	0
G	-20.5	14.1	.276	.085	-	-	.149	4	.455	.721	-	-	1.176	0
H	-40	13.7	.284	.086	-	-	.159	4	.485	.700	-	-	1.185	0
J	-51	13.9	.294	.098	-	-	.166	4	.564	.753	-	-	1.317	0
K	-61	13.9	.310	.093	.142	.184	.170	4	.523	.725	.914	.134	1.248	15
L	-68	14.0	.300	.087	.085	.222	.189	3.0	.519	.701	1.220	0	1.220	25
M	-74	14.1	.302	.077	-	-	.192	4	.427	.843	-	-	1.270	trace
A	-85.4	12.7	.300	.0605	.291	.091	.205	0.5	.386	.239	.625	0	.625	90
B	-92.4	13.2	.308	.077	.297	.102	.212	0.7	.506	.187	.693	0	.693	90
C	-108.3	12.2	.316	.071	.313	.078	.232	0.15	.480	.059	.539	0	.539	95
D	-196	1.3	.223	.0076	.223	.0076	-	0	.022	0	.022	0	.022	100

TABLE 10 HALF SIZE SLOW NOTCH BEND STEEL V

Specimen No.	Temp. °C.	% Expansion	Maximum Load		1st Brittle Crack		Load at Yield	Energy kg.m.					% Crystalline
			Load Tons	Deflection in.	Load Tons	Deflection in.		Before Maximum Load	After Maximum Load	Before 1st. Brittle Crack	After 1st. Brittle Crack	Total	
C	18	18.5	.242	.075	-	-	.124	.372	.760	-	-	1.132	0
D	0	18.2	.260	.082	-	-	.137	.445	.774	-	-	1.219	0
E	-24.5	18.1	.270	.069	-	-	.153	.390	.810	-	-	1.200	0
F	-35	17.3	.260	.069	-	-	.142	.372	.818	-	-	1.190	0
G	-44.5	18.9	.285	.070	-	-	.1685	.422	.877	-	-	1.299	0
H	-49.6	18.6	.284	.078	-	-	.156	.469	.847	-	-	1.316	0
J	-55	18.8	.284	.0695	-	-	.182	.422	.838	-	-	1.260	0
K	-59	17.5	.289	.067	-	-	.183	.392	.906	-	-	1.308	0
L	-65	16.6	.298	.068	-	-	.187	.425	.816	-	-	1.241	0
M	-70	17.7	.294	.068	-	-	.192	.425	.710	-	-	1.135	2
N	-75	17.5	.290	.057	.024	.328	.201	.350	.912	1.258	.004	1.262	Faint trace
O	-85.4	17.6	.315	.068	.0615	.211	.2095	.433	.716	1.087	.062	1.149	5
P	-90	18.6	.288	.063	.1075	.184	.200	.388	.646	1.034	0	1.034	30
R	-105.3	17.4	.294	.071	.240	.118	.204	.450	.323	0.773	0	.773	75
S	-196	0.9	.263	.001	.263	.001		.041	0	.041	0	.041	100

TABLE II DOUBLE SIZE SLOW NOTCH BEND STEEL P

Specimen No.	Temp. °C.	% Expansion	Maximum Load Tons	Maximum Deflection in.	Load at 1st Brittle Crack Tons	Deflection at 1st Brittle Crack in.	Load at Yield	mm. Fibrous	Energy kg.m.					% Crystalline
									Before Maximum Load	After Maximum Load	Before 1st Brit. Crack	After 1st Brit. Crack	Total	
8	+66	13.28	3.72	.203	-	, -	2.10	16	15.51	23.67	-	-	39.18	0
9	+56	12.58	3.91	.199	-	-	2.21	16	15.91	27.50	-	-	43.41	trace
7	+46	13.04	3.74	.191	2.18	.394	2.16	7.5	14.61	26.01	31.02	9.60	40.62	7
6	+32.8	11.10	3.36	.193	3.82	.2245	2.24	2.1	15.32	12.77	18.50	9.59	28.09	30
1	+22	8.95	3.95	.202	3.93	.214	2.28	1.3	17.24	÷4.3	18.55	÷3.0	÷21.5	70
2	+11	8.96	4.02	.200	4.02	.200	2.325	0.8	16.55	÷1.5	16.55	÷1.5	÷18.0	80
3	-12	7.29	4.08	.162	4.03	.162	2.655	0.6	13.50	0	13.50	0	13.50	90
4	-38	5.6	3.96	.122	3.96	.122	2.92	0.1	9.90	0	9.90	0	9.90	95
5	-60.5	2.92	3.42	.049	3.42	.049	3.32	0	2.87	0	2.87	0	2.87	98
10	-85	2.53	3.67	.019	3.67	.019	3.65	0	1.42	0	1.42	0	1.42	100
11	-99	2.04	3.89	.008	3.89	.008	-	0	0.41	0	0.41	0	0.41	100
12	-125	1.40	-	-	-	-	-	0	-	-	-	-	-	100

TABLE 12 DOUBLE SIZE SLOW NOTCH BEND STEEL V

Specimen No.	Temp. °C.	% Expansion	Maximum Load		1st Brittle Crack		Load at Yield	mm. Fibrous	Energy kg.m.					% Crystalline
			Load Tons	Deflection in.	Load Tons	Deflection in.			Before Maximum Load	After Maximum Load	Before 1st Brittle Crack	After 1st Brittle Crack	Total	
10	+64	12.54	3.70	.172	-	-	2.40	16	13.24	22.84	-	-	36.08	0
9	+46	12.54	3.72	.155	-	-	2.50	16	11.90	26.00	-	-	37.90	0
1	+15	12.9	3.79	.158	-	-	2.64	16	13.10	24.48	-	-	37.58	0
2	-12	13.12	3.94	.142	-	-	2.785	16	11.78	27.20	-	-	38.98	2
3	-21	13.24	3.94	.173	-	-	2.87	16	15.01	≐23.4	-	-	≐38.4	0
4	-33.5	14.1	4.14	.175	-	-	3.04	16	15.97	≐24.2	-	-	≐40.2	1
5	-44	13.4	4.21	.162	1.50	.445	3.22	16	14.19	≐24.5	36.00	≐2.7	≐38.7	10
6	-51	11.34	4.15	.169	4.00	.224	3.20	No full lip	14.29	18.63	20.23	12.69	32.92	30
7	-60.5	10.6	4.36	.1495	3.10	.307	3.48	6.5	13.92	15.88	29.80	0	29.80	62
8	-71	9.82	4.44	.147	3.69	.268	3.66	5.0	13.69	12.95	26.64	0	26.64	67
11	-79	8.59	4.15	.075	4.15	.075	≐3.52	3.0	6.69	0	6.69	0	6.69	80
12	-96	5.40	4.24	.044	4.24	.044	≐3.94	0.2	3.84	0	3.84	0	3.84	98
13	-112	3.75	4.11	.026	4.11	.026	-	0	1.07	0	1.07	0	1.07	100
14	-123	0.94	3.28	.007	3.28	.007	-	0	0.40	0	0.40	0	0.40	100

TABLE 13
STEEL 'P' STANDARD SIZE NOTCH TENSILE

Specimen Number	Temp. °C.	Maximum Load	Deflection at Maximum Load	Load at Fracture	Deflection at Fracture	Yield Load		Energy in kg.m.			% Crystalline
						1	2	Before Maximum	After Maximum	Total	
2	+54.3	4.21	.151	3.97	.177	2.60	3.05	13.81	2.84	16.65	0
3	+35	4.22	.152	3.85	.193	2.74	3.22	13.45	4.48	17.93	10
4	+30.5	4.28	.168	4.09	.1905	2.65	3.15	15.25	2.24	17.49	45
5	+10.4	4.43	.1725	4.30	.190	2.86	3.34	15.90	1.97	17.87	80
6	-6	4.57	.1675	4.35	.187	3.02	3.63	16.70	2.30	19.00	55
7	-18	4.63	.172	4.59	.182	3.25	3.75	17.05	1.13	18.18	85
8	-31	4.75	.155	4.75	.155	3.34	3.95	15.20	0	15.20	95
9	-50.5	4.65	.115	4.65	.115	3.71	4.31	11.78	0	11.78	98
10	-65.5	4.13	.030	4.13	.030	4.08	-	1.53	0	1.53	100
11	-89	4.54	.031	4.54	.031	4.54	-	1.89	0	1.89	100
12	-111.5	5.41	.0195	5.41	.0195	-	-	1.194	0	1.194	100

TABLE 14
STEEL 'V' STANDARD SIZE NOTCH TENSILE

Specimen Number	Temp. °C.	Maximum Load	Deflection at Maximum Load	Load at Fracture	Deflection at Fracture	Yield Load		Energy in kg.m.			% Crystalline
						1	2	Before Maximum	After Maximum	Total	
2	23.5	4.34	0.140	4.10	0.164	3.19	3.53	13.50	2.50	16.00	0
1	22	4.30	—	3.80	—	—	3.28	—	—	—	0
3	-1.2	4.66	0.172	4.46	0.191	3.37	3.81	17.62	2.22	19.84	0
5	-10.0	4.57	0.137	4.23	0.105	3.53	4.12	13.88	3.31	17.19	0
4	-22	4.34	0.144	4.03	0.1035	3.68	4.21	16.19	2.44	18.63	20
6	-20.4	4.70	0.100	4.50	0.1805	3.61	4.13	17.05	2.34	19.39	20
7	-41	4.73	0.161	4.61	0.183	3.83	4.27	17.11	2.58	19.69	25
9	-50	4.85	0.0735	4.62	0.009	4.13	4.17	13.42	3.08	16.50	45
8	-60	4.34	0.072	4.48	0.125	4.34	4.37	7.75	6.50	14.25	40
10	-71	5.58	—	5.20	—	4.68	4.74	—	—	49.6	90
11	-80	5.53	0.006	5.53	0.066	5.11	5.18	7.96	0	7.96	95
12	-100	5.46	0.030	5.46	0.030	5.40	—	2.63	0	2.63	100

TABLE 15STATIC YIELD STRESS VALUES OBTAINED BY THE AUTHOR

Steel	Temperature °C.	Yield Stress (10^4 p.s.i.)	
		Upper	Lower
P	+30	3.15	2.90
P	+20.7	-	3.00
P	-30	-	3.16
P	-70	-	3.46
P	-93	-	4.025
P	(-103)	-	(4.325)
V	+118	4.27	3.46
V	+19.7	-	3.92
V	-30	-	3.82
V	-74	-	4.48
V	-87	-	4.58
V	(-99)	-	(4.69)

TABLE 16YIELD STRESS VALUES OBTAINED IN THE INSTRON MACHINE

Steel	Rate of Cross-Head Movement in./min.	Yield Stress (10^4 p.s.i.)		Stress Rate (where applicable) p.s.i./sec.
		Upper	Lower	
P	0.002	-	3.16	-
P	0.01	-	3.33	-
P	0.10	-	3.44	-
V	0.002	-	3.78	-
V	0.01	4.01	3.99	5.30
V	0.10	4.11	4.10	4.9×10^3

TABLES 16 TO 19 ARE YIELD STRESS RESULTS OBTAINED BY TAYLOR (110)

(All stresses are in 10^4 p.s.i.)

TABLE 17 - STEEL 'P'

Temperature °C.	Dynamic Yield Stress		Static Yield Stress	
	Upper	Lower	Upper	Lower
+15	6.61	5.15	3.85	3.74
-17	8.46	6.58	4.255	4.075
-70	10.30	8.29	5.69	5.42
-186	11.65	10.77	13.0	12.66

TABLE 18 - STEEL 'V'

Temperature °C.	Dynamic Yield Stress		Static Yield Stress	
	Upper	Lower	Upper	Lower
+15	(8.14 7.30)	(5.33 5.46)	4.79	4.64
-17	9.19	6.65	6.34	5.15
-70	11.20	8.29	7.10	5.69
-186	15.58	13.52	14.44	12.56

TABLE 19

DYNAMIC UPPER YIELD STRESSES OF STEELS 'P' AND 'V'
AT DIFFERENT RATES OF STRESSING (+15°C.)

Steel	Rate of Stressing in. (10^4 p.s.i. per sec.)		
	4.93	8.58	13.6
P	6.45	6.92	7.03
V	7.39	7.89	8.14

APPENDIX 'B'THE YIELD POINT PHENOMENA IN MILD STEEL AND THEIR
VARIATION WITH TEMPERATURE AND STRAIN RATE

It has been noted that in general the body centred cubic metals exhibit a sharp yield point, discontinuous yielding, and a delay time for yielding under a constant rapidly applied stress (22, 5b). Although some of the hexagonal close packed metals may also occasionally exhibit this behaviour, it is rarely found in face centred cubic metals. An explanation of this may be found in the concept of dislocations being locked by "atmospheres" which are caused, for example, by large solute atoms in the lattice.

Cottrell and Bilby (85) have used this idea to explain the upper and lower yield points and also strain ageing, and more recently Cottrell (44) has explained radiation damage and its effect on brittle fracture from a similar viewpoint. Various methods have been employed to use the above concepts in a quantitative sense and two examples are shown below, differing mainly in the simplifying assumptions which are made in their derivation.

1. The Theory of Campbell and Doby (86)

Cottrell and Bilby's theory was used for the activation energy, U , and it was proposed that yield occurred when the density of released dislocations reached a certain critical value. Using Yokobori's (87) approximation for the activation energy, Campbell (84) suggested that yield would occur at a

time t_1 , when:-

$$\int_0^{t_1} \left[\frac{f(t)}{\sigma_0} \right]^\lambda dt = \text{constant}, C \quad \text{---} \quad \text{(B.1)}$$

where the stress followed a relation $\sigma = f(t)$

and σ_0 = the yield stress at 0°K .

λ = a dimensionless constant.

At a constant stress rate $\dot{\sigma}$ this gives

$$t_1 = \left[(1 + \lambda) C \left(\frac{\sigma_0}{\dot{\sigma}} \right)^\lambda \right]^{\frac{1}{1+\lambda}} \quad \text{---} \quad \text{(B.2)}$$

or

$$\sigma_{yd} = \sigma_0 \left[(1 + \lambda) C \frac{\dot{\sigma}}{\sigma_0} \right]^{\frac{1}{(1+\lambda)}} \quad \text{---} \quad \text{(B.3)}$$

and the delay time to yield at constant stress

$$t_d = C \left(\frac{\sigma_0}{\sigma} \right)^{-\lambda} \quad \text{---} \quad \text{(B.4)}$$

Campbell compared the above theory with the results of tests at constant strain rate, constant stress, and also impact tests. These results gave good agreement with the theory at low loading times but were not satisfied at loading times greater than one second. These results together with the fact that there appears to be a static yield stress, which is independent of rate of strain, led Campbell to suggest that when the time of loading becomes sufficiently great some other mechanism must come into operation. Campbell and Doby (86)

therefore modified the hypothesis by introducing a strain ageing process caused by the diffusion of carbon atoms and suggested that equation (B.1) be replaced by:-

$$\exp. (-bt_1) \int_0^{t_1} \sigma^{-d} \exp.(bt)dt = c \text{ ————— (B.5)}$$

Where

$1/b$ = Mean time to relock a dislocation

$$c = C\sigma_0^{-d}$$

By means of this modification quite good agreement was obtained with the experimental results.

Campbell and Marsh (88) continued the investigation by considering the delay time, t_d , which they defined as the time elapsing before the onset of gross yield measured from the time when the applied stress reached the static yield stress. Both the earlier work of Campbell and also that of Krafft and Sullivan (89) indicated that for rapid loading the logarithm of the delay time varies linearly with the logarithm of the applied stress. The experimental results of Campbell and Marsh confirmed the validity of this relationship for delay times less than 200 milliseconds. They also confirmed (in the range t_d 200 milliseec.) the relationship between the applied stress and $t_d d^3$, proposed by Russell et al. (90) and derived the relationship

$$\sigma \propto (t_d d^3)^{-\frac{1}{d}} \text{ ————— (B.6)}$$

Where d = mean grain diameter.

2. Hendrickson and Wood's Theory (21)

This theory employs the model of a yield nucleus presented by Vreeland, Wood and Clark (91) and also Fisher's approximation (92) to the activation energy. From this basis Hendrickson and Wood derived a rather complex expression for the upper yield stress as a function of the constant applied stress rate, and compared their experimental results with this theory and also with that of Campbell. In general the experimental points lay between the two theoretical curves. They also developed an expression for the delay time,

$$t_d = 2aG/\nu\sigma \exp. (B/\sigma T) \text{ ————— (B.7)}$$

Where

G = Shear Modulus

a, B, ν are constants

T = Absolute temperature

It was found that both this equation and Campbell's expression for the delay time gave good agreement with experimental results at higher temperatures (+250°F. to -75°F.) but at very low temperatures there was some divergence.

Experimental Yield Stress Values

The marked effect of temperature upon the yield stress has been studied for some time (93, 94) as has also the effect of strain rate (79b, 94, 95 - 98). Many of the more recent tests have combined both effects and investigated the variation of the upper and lower yield stress values over a range of temperatures (21, 81, 99-101) and also the variation with

grain size (102). Increased emphasis has also been placed on the yield delay time (20, 81) and its variation with grain size (88, 103).

APPENDIX 'C'MATERIAL SPECIFICATION AND THE PREPARATION OF SPECIMENS1. Material

The two steels employed in these investigations were chosen from a number of pedigreed steels selected for testing by a research panel of the Admiralty Advisory Committee on Structural Steel and detailed properties are available in the reports to this Committee (67). Both steels were obtained in the form of 1 in. thick plates. The general descriptions, conditions and properties of the two steels were as follows:-

Steel 'P'

An ordinary mild steel, semi-killed in the as rolled condition.

Analysis

0.18%C. 0.54%Mn. 0.05%Si. 0.039%S. 0.028%P. 0.16%Ni.

0.024%Sn. 0.10%Cu. 0.070%Cr. 0.005% soluble Al.

Ferrite Grain Size 39 Grains/in.² at x 100.

Plate Specification:

MS/1/ii/NC (See Ref. (67)).

Steel 'V'

A semi-killed, normalised steel.

Analysis

0.15%C. 1.19%Mn. 0.08%Si. 0.039%S. 0.020%P. < 0.005%Al.

Ferrite Grain Size 80 Grains/in.² at x 100.

Location of Plates: 2997/1/ii/IC, 2997/2/ii/IC, and

2997/2/iii/IC (See Ref. (67)).

Mechanical Properties (Stresses in Ton/sq.in.)TABLE

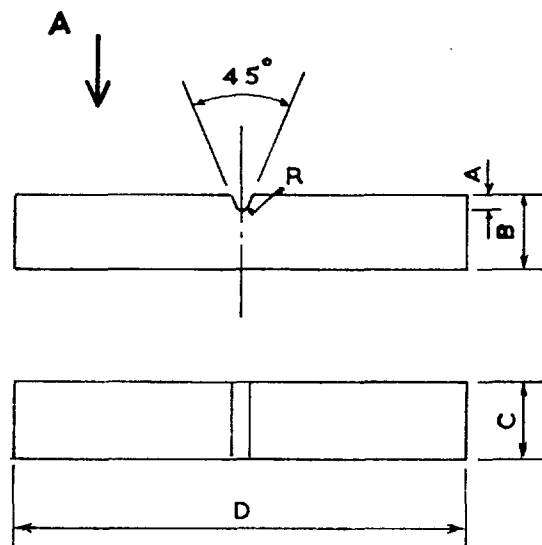
<u>Steel</u>	<u>Direction of Test</u>	<u>Yield Stress</u>	<u>0.3% Proof Stress</u>	<u>U.T.S.</u>	<u>Reduction of Area %</u>
P	Longitudinal	14.7	17.5	27.4	54
	Transverse	14.9	14.3	26.5	53
V	Longitudinal	16.1	16.4	26.7	60
	Transverse	16.7	16.3	26.0	48

Microstructure

The microstructures of the two steels are shown in Plates 9 and 10. Plates 9(a) and 10(a) show the microstructure as viewed from on top of the plate (Direction A in Fig. C.1.), whilst Figs. 9(b) and 10(b) show the microstructure of the ends of the specimens (Direction B in Fig. C.1.).

2. Types of Specimen Employed and their Preparation.

The behaviour of these steels in the various standard tests revealed a quite strong dependence of the properties upon the orientation of the specimen in relation to the final rolling direction of the steel. Because of this, each of the pedigreed plates was marked in many places to show the rolling direction, and great care was taken at all stages of the various machining operations to remember the exact orientation of the specimen blanks. Also flame-cutting operations were reduced to a minimum to avoid any possibility of heat treating the fracture zone. Flame-cutting was only used to reduce the



	HALF	STD.	DOUBLE
A	1	2	4
B=C	5	10	20
D	30	60	120
R	0.125	0.25	0.50

MILLIMETRES

NOTCH-BEND SPECIMEN
FULL SIZE

FIG.C.1

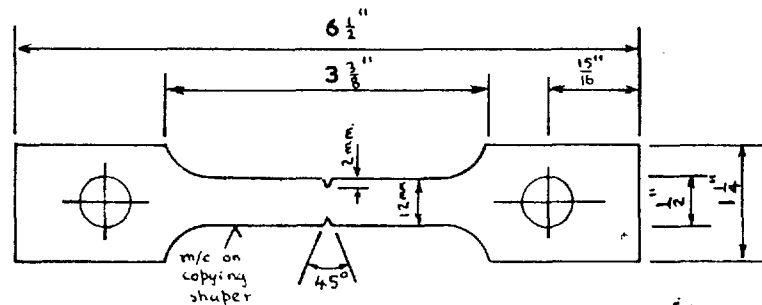
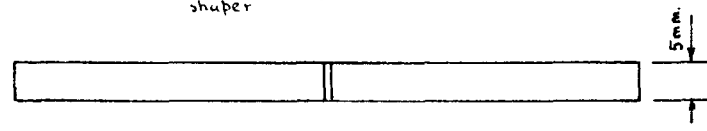
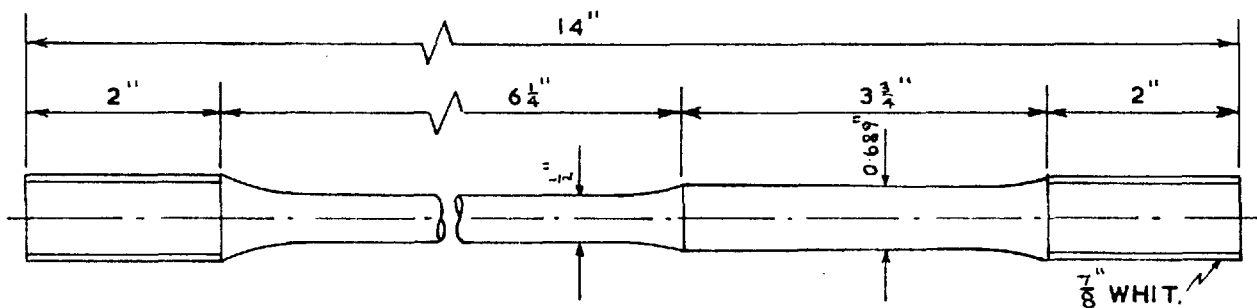


FIG.C.2



NOTCH-TENSILE
SPECIMEN

SCALE 1/2



STATIC YIELD STRESS SPECIMEN

SCALE 1/2

FIG. C.3

DIMENSIONS OF SPECIMENS EMPLOYED

plates to manageable proportions and the heat affected zones were either removed by machining, as in the notch-bend specimens, or located in unimportant places, such as the threaded ends of the yield stress specimens.

(a) Notch-Bend Specimens

The dimensions of the three sizes of notch-bend specimen are given in Fig. C.1. The method of preparation was much the same in all cases except that three half size specimens could be machined from the plate thickness, two standard size specimens and only one double size specimen. The portion of plate was cut into rough blanks with a bandsaw and a batch of blanks was then machined to slightly oversize on a shaper. The blanks were then ground to the final dimension on all four sides, a tolerance ± 0.001 in. being specified. A special jig was set up in a mill which both machined the ends of the specimen and cut the notch in one operation, thus ensuring both the accurate location of the notch and also that there was no variation of dimensions between specimens. All burrs were carefully removed by hand so that there should be the minimum of bevelling. This was important in ensuring the exact location of the specimen on the bending jig and also the accurate measurement of specimen deformation. Special profile milling cutters were obtained to cut the various notches and the resultant notch depth and notch profile were checked on a Hilger Measuring Projector.

(b) Notch-Tensile Specimens

In the case of the notch-tensile specimens it was considered satisfactory, provided that sufficient care was taken, if the profiled edges (see Fig. C.2) were machined on a copying shaper. The other two faces were ground to size and the two holes for the loading pins reamed to ensure a good fit. The dimensions of the specimen were intended to follow as closely as possible those of the standard Charpy V-notch specimen. Thus the parallel portion was the same length as that of the notch-bend specimen as was also the cross-sectional area beneath the notch. The notch was cut with the same milling cutter that was used for the standard size notch-bend specimens.

(c) Yield Stress Tests

The type of specimen employed in the static yield stress tests is shown in Fig. C.3. Despite their length the blanks for these specimens were still sawn from the portion of plate so that there should be no suspicion that flame-cutting had altered the material properties. A copying lathe employing a steady was used to turn the specimens down to about 0.005 in. oversize and a profiled grinding wheel was then used to obtain the final dimensions, two steadies being required for this operation. Care was taken to ensure that as good as possible a surface finish was obtained. The axiality of the threads with the gauge length was checked on the measuring projector to ascertain whether any eccentricity of load were possible from this cause. The check showed that an eccentricity of loading of 0.003 in. was possible.

APPENDIX 'D'INVESTIGATION OF THE CHARACTERISTICS OF THE
DENISON TESTING MACHINE

Using the methods described on p. 71, true load time records were obtained and compared with the values indicated by the mechanical recorder. The record obtained at the fastest rate of cross-head movement ($\dot{=}$ 20 in./min.), which corresponded in the standard test to a rate of loading of 12,000 lb./sec., is shown in Fig. D.0. A similar record was taken at the lowest rate of cross-head movement ($\dot{=}$ 1.5 in./min. but, since this used the same rate of paper travel as the time scale, it is far too long for convenient reproduction here. The error in the value of the maximum load was, however, only about 2% and it was possible to predict a percentage error at a given stage of the test. It did seem that the damping applied to the spring system was somewhat excessive and the following investigation was performed.

(a) Theory

The final lever and weighing system of this machine is shown diagrammatically in Fig. (D.1), and for the purposes of analysis may be represented by the system shown in Fig. (D.2). It is convenient to consider this system to be concentrated on the axis of the main weighing spring.

The mathematical analysis of such a system is as follows (112).

Let x = Displacement from the equilibrium position at time t .

S = Restoring force per unit displacement from the equilibrium position.

C = Damping force per unit velocity (Damping Coefficient).

The equation of motion may be written:

$$\frac{W}{g} \frac{d^2x}{dt^2} = -Sx - C \frac{dx}{dt} \quad \text{_____} \quad (D.1)$$

which may be rewritten

$$\frac{d^2x}{dt^2} + 2\mu \frac{dx}{dt} + \omega^2 x = 0 \quad \text{_____} \quad (D.2)$$

where $2\mu = \frac{Cg}{W}$

$$\omega^2 = \frac{Sg}{W}$$

The solutions to the above equations are:

(1) For Heavy Damping; $\mu > \omega$

$$\begin{aligned} x &= Ae^{(-\mu + \sqrt{\mu^2 - \omega^2})t} + Be^{(-\mu - \sqrt{\mu^2 - \omega^2})t} \\ &= Ae^{-m_1 t} + Be^{-m_2 t} \quad \text{_____} \quad (D.3) \end{aligned}$$

where

$$m_1 = \mu - \sqrt{\mu^2 - \omega^2}$$

$$m_2 = \mu + \sqrt{\mu^2 - \omega^2}$$

$$\therefore m_1 + m_2 = 2\mu$$

(2) Critical Damping; $\mu = \omega$

$$x = (A + Bt)e^{-\mu t} \quad \text{_____} \quad (D.4)$$

(3) Light Damping:-----

$$x = e^{-\mu t} (C \cos. \sqrt{(\omega^2 - \mu^2)t} + D \sin. \sqrt{(\omega^2 - \mu^2)t}) \text{---(D.5)}$$

Where

$$C = (A + B) \text{ and } D = j(A - B)$$

$$\text{The periodic time} = \frac{2\pi}{\sqrt{(\omega^2 - \mu^2)}} = t_p \text{---(D.6)}$$

$$(\text{c.f. Periodic time of undamped vibrations, } t_p = \frac{2\pi}{\omega})$$

The ratio of successive amplitudes is $e^{\mu t}$.

or

$$\log_e \frac{a_1}{a_2} = \mu t_p = \frac{2\pi\mu}{\sqrt{(\omega^2 - \mu^2)}} \text{---(D.7)}$$

(b) Experimental Determination of the Damping Constants

The dashpot was disconnected from the system but since some viscous damping, due to windage, was still present the system could still be represented by Fig. (D.2). The constant speed paper drive system was set up on the recorder and the lever given a displacement. With the paper running at constant speed the lever was released and the pen traced out a record of the type shown in Fig. (D.3).

From records of this type the natural frequency was determined to be 3.081 c/s

$$\text{i.e. } t_p = 0.3246 \text{ sec.}$$

Also the ratio of successive amplitudes was found to be 1.185.

Hence from equation (D.7)

$$\log_e 1.185 = .3246 \mu$$

$$\therefore \mu = .5234 \text{ rad./sec.}$$

Also from equation (D.6)

$$t_p = .3246 = \frac{2\pi}{(\omega^2 - .5234^2)}$$

$$\therefore (\omega^2 - .274) = 19.36$$

$$\therefore \omega = 19.35$$

(i.e. virtually no change from the undamped frequency)

The dashpot was now reconnected and a similar test was carried out. In this case instantaneous release was most important, and this was achieved by tying the lever back with a thread to give the initial displacement and burning the thread to release the lever. Records of a type similar to those shown in Fig. (D.4a) were obtained. It is seen that this is heavy damping and equation (D.3) will apply.

$$\therefore x = Ae^{-m_1 t} + Be^{-m_2 t}$$

Now as will be seen later, $m_2 > m_1$ and as soon as t becomes at all large $e^{-m_2 t} \rightarrow 0$

Hence after point P say

$$x = Ae^{-m_2 t}$$

Consider any two points 1 and 2 on the curve after P

$$x_1 = Ae^{-m_1 t_1}$$

$$x_2 = Ae^{-m_1 t_2}$$

$$\therefore x_1/x_2 = e^{-m_1(t_1 - t_2)}$$

Thus by measuring x_1 and x_2 , t_1 and t_2 (taking an arbitrary

origin at P) the value of m_1 could be calculated. Thus since ω is unchanged the value of μ may be calculated from

$$m_1 = \mu - \sqrt{\mu^2 - \omega^2}$$

The value of μ was found to be approximately 45.

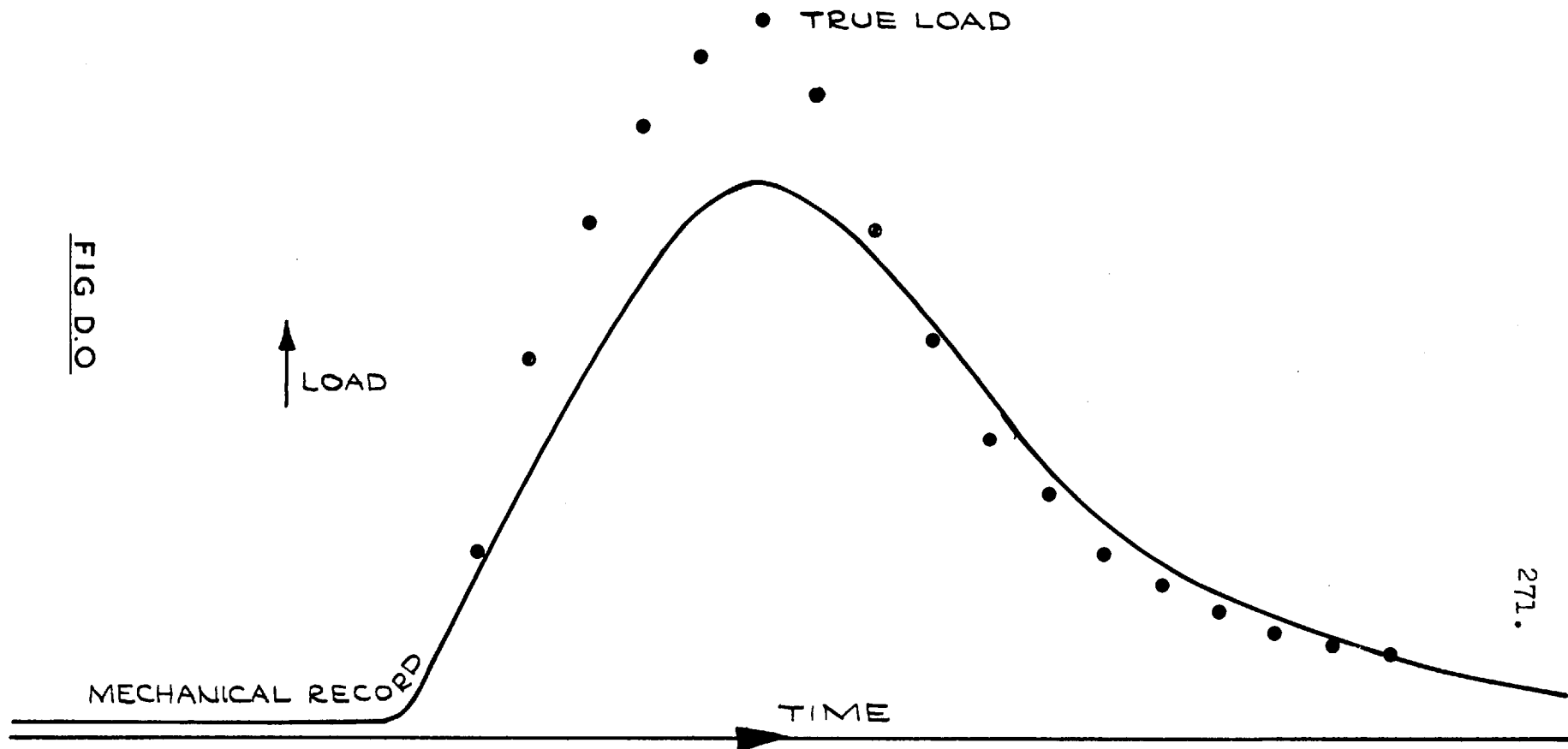
For critical damping we require

$$\mu = \omega = 19.4$$

The existing piston of diameter 1.113 in. was replaced by a piston of 1.108 in. diameter and similar tests were carried out giving $\mu = 22$ rad./sec. See Fig. (D.4b)

A further 0.001 in. was removed from the new piston and the tests repeated once more. It was obvious from the record (Fig. (D.4c)) that the damping was just less than critical and the recorder response could be improved no more without the danger of over-shoot.

ERROR OF MECHANICAL RECORDER AT HIGH LOADING RATES
BEFORE REDUCTION IN DAMPING



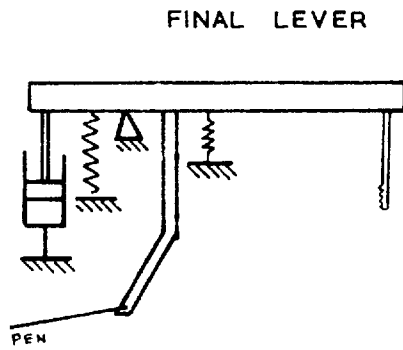


FIG. D.1

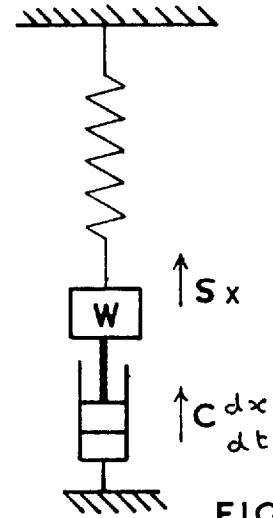


FIG. D.2

SPRING SYSTEM IN THE DENISON MACHINE

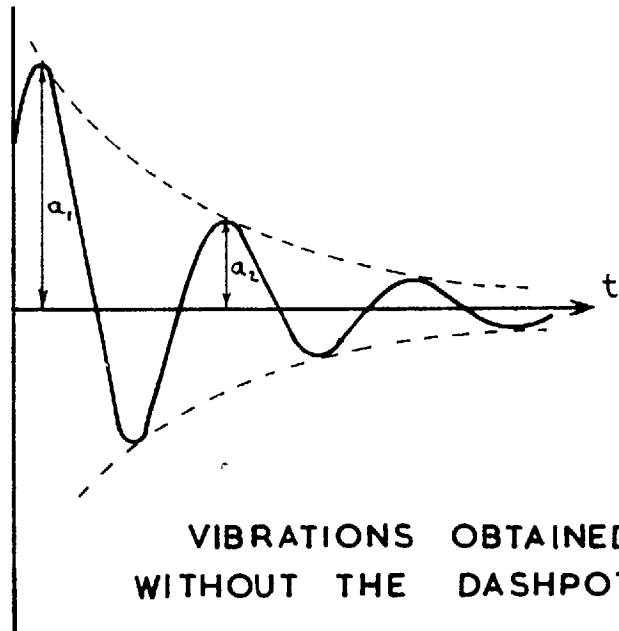
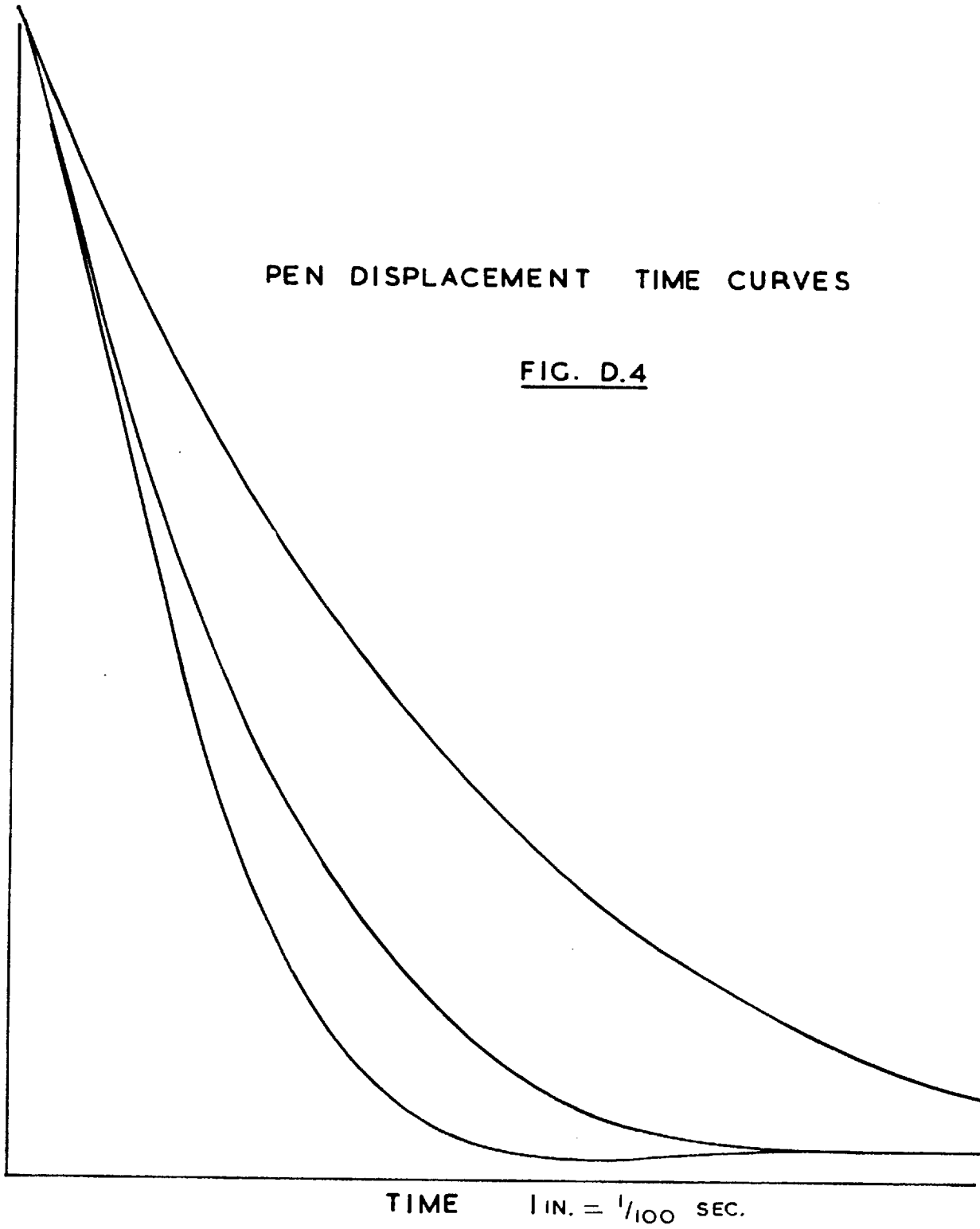


FIG. D.3

VIBRATIONS OBTAINED
WITHOUT THE DASHPOT

PEN DISPLACEMENT TIME CURVES

FIG. D.4

APPENDIX 'E'AN ESTIMATION OF THE ELASTIC STRESSES AND STRESS RATES
AT THE NOTCH ROOTS OF THE SPECIMENS TESTED1. Notch-Bend Specimens

The nominal bending stress at the root of the notch is:-

$$\sigma_{nb} = \frac{PL}{4Z'}$$

where P = Applied load (lb.)

L = Distance between specimen supports

Z' = Section modulus of the notched cross-section.

Therefore, if k is the elastic stress concentration factor, the true elastic stress at the notch root is given by:-

$$\sigma = k\sigma_{nb} = \frac{kPL}{4Z'} \quad \text{-----} \quad (E.1)$$

There is some doubt as to the value of k for the V-notch in question and estimates as high as 5.5 have been made, but the value of 3.71 which was calculated for the Izod notch by Hendrickson, Wood and Clark (23) will be used.

Substituting the various specimen dimensions in equation (E.1) gives:-

For the Standard Size Specimens $\sigma = 224.4P$ p.s.i.

Similarly the rate of elastic stressing at the root of the notch is given by:-

$$\dot{\sigma} = 224.4 \dot{P} \text{ p.s.i./sec.}$$

where $\dot{P}$ is the rate of applied loading (lb./sec.).

For the Half-size Specimens $\sigma = 8.97.6 P$ p.s.i.

For the Double-size Specimens $\sigma = 56.1 P$ p.s.i.

The measured rates of loading and the resultant estimated rates of stressing are shown below.

(a) Tests on Size-Effect in Slow Notch-Bend Tests

Half size: $\dot{P} = 230 \text{ lb./sec.}$ $\dot{\sigma} = 2.02 \times 10^5 \text{ p.s.i./sec.}$
 Standard size: $\dot{P} = 900 \text{ lb./sec.}$ $\dot{\sigma} = 2.02 \times 10^5 \text{ p.s.i./sec.}$
 Double size: $\dot{P} = 4010 \text{ lb./sec.}$ $\dot{\sigma} = 2.25 \times 10^5 \text{ p.s.i./sec.}$

(b) Tests on the Effect of Strain-Rate in Standard Size Notch-Bend Tests

Slow notch-bend: $\dot{P} = 900 \text{ lb./sec.}$ $\dot{\sigma} = 2.02 \times 10^5 \text{ p.s.i./sec.}$
 Fast notch-bend: $\dot{P} = 12,000 \text{ lb./sec.}$ $\dot{\sigma} = 2.70 \times 10^6 \text{ p.s.i./sec.}$
 Impact notch-bend: $\dot{\sigma} \doteq 1.7 \times 10^9 \text{ p.s.i./sec.}$ (see below)

It is extremely difficult to estimate the stress rate at the notch root because, as is pointed out in Section 5, the load gauges indicate the strain in the striker nose and not necessarily the load applied to the specimen. Also the complex stress wave effects within the specimen itself produce a system which is very difficult to analyse. Although an estimation of the stress rate from external conditions may be grossly misleading there is, at present, no alternative.

Hollomon (105) used the impact velocity together with an elastic analysis connecting the curvature of the beam and the strain at the outer fibres and obtained an elastic strain rate at the notch root of 600/sec. (This value of $600 = 10^{2.8}/\text{sec.}$ gave rise to the frequently quoted value of $10^3/\text{sec.}$). Since this analysis does not, however, take into account any specimen indentation it would seem more reasonable to take the ratios

of the notch root stresses in the impact and fast notch-bend tests to be in the same ratio as their striker velocities. This gives $\dot{\sigma} = 1.7 \times 10^9$ p.s.i./sec.

2. Notch-Tensile Specimens

The nominal elastic tensile stress at the notched cross-section is:-

$$\sigma_{nt} = \frac{P}{A} \text{ p.s.i.}$$

where: P = Applied load (lb.)

A = Cross-sectional area at the notched section (in.²).

Thus the true elastic stress at the root of the notch is

$$\sigma = k\sigma_{nt} = \frac{kP}{A} \text{ (E.2)}$$

The notches employed in the notch-tensile specimens were identical in profile and depth with the standard Charpy V-notch, and it will be assumed that the stress concentration factor is also the same. Substituting the specimen dimensions in equation (E.2) gives:-

$$\dot{\sigma} = 29.92 \dot{P} \text{ p.s.i./sec.}$$

The measured rate of loading was $\dot{P} = 6,500$ lb./sec.

Therefore $\dot{\sigma} = 1.95 \times 10^5$ p.s.i./sec.

APPENDIX 'F'CALIBRATION ROUTINE FOR THE INSTRUMENTED IMPACT TEST
(See Fig. 20)

(1) The external Y-shift voltage supply was connected to amplifier No. 2 and the amplifier gain set at 10. The deflection calibration was then performed using the external Y-shift to traverse spot No. 1 (zero brightness on spot No. 2).

(2) The oscilloscope Y-plate input was switched from amplifier No. 2 to the oscillator and the brightness increased on spot No. 2 and extinguished on spot No. 1. The time-base calibration was then performed.

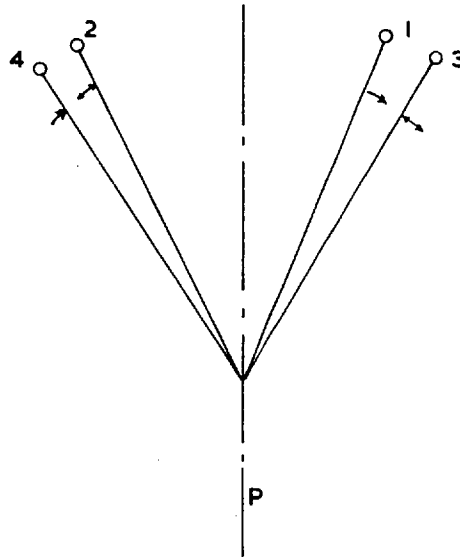
(3) The oscillator was replaced by amplifier No. 2 once more, this time switching the amplifier input from the external shift to the load-cell. The amplifier gain was increased to 250.

Amplifier No. 1 was then disconnected and the time-base was connected to X_1 , as well as X_2 the coarse setting on the time-base being adjusted to give a slow traverse across the screen. This traverse was used in conjunction with the built-in calibration resistors in the load cell bridge to give a load calibration for each spot in turn.

(4) The coarse time-base setting was returned to its operational position and amplifier No. 1 reconnected to X_1 , in place of the time-base. The brightness of the two spots was then reduced to zero until the next test was ready.

Corrections for Friction and Windage Losses in the Notched
Impact Test.

Let the initial position of the pendulum be the point 1 and let 2, 3 and 4 be the positions of the pendulum at the end of each swing.



If the energy lost in swinging from 1 to 2 is E_1
 then " " " " " 1 to P $\div \frac{1}{2} E_1$
 and if " " " " " 2 to 4 is E_2
 then " " " " " 2 to P $\div \frac{1}{4} E_2$

Thus the total energy lost in swinging from 1 to 2 $= (\frac{1}{2} E_1 + \frac{1}{4} E_2)$

Initially the pencil leads were removed and the deflection contacts held back. The pendulum was released from its maximum position and allowed to swing freely, the friction pointer being continually reset to allow the points of swing

through, such as 2, 4, 6 etc., to be read and recorded. Thus the value of E_1 could be determined from the initial swing, and by plotting the points of swing through against the number of swings the value of E_2 could be calculated for any given final angle.

The losses due to friction of the contacts on the carbon strips and to energy absorbed in breaking the pencil leads, E_F , was determined by releasing the pendulum from 1 and recording the final angle. The difference between the indicated energy absorbed and E_1 gave E_F .

The results obtained were as follows:

$$\frac{1}{3}E_1 = 0.35 \text{ kg.m.}$$

$$\frac{1}{4}E_2 = 0.16 \text{ to } 0.37 \text{ kg.m.}$$

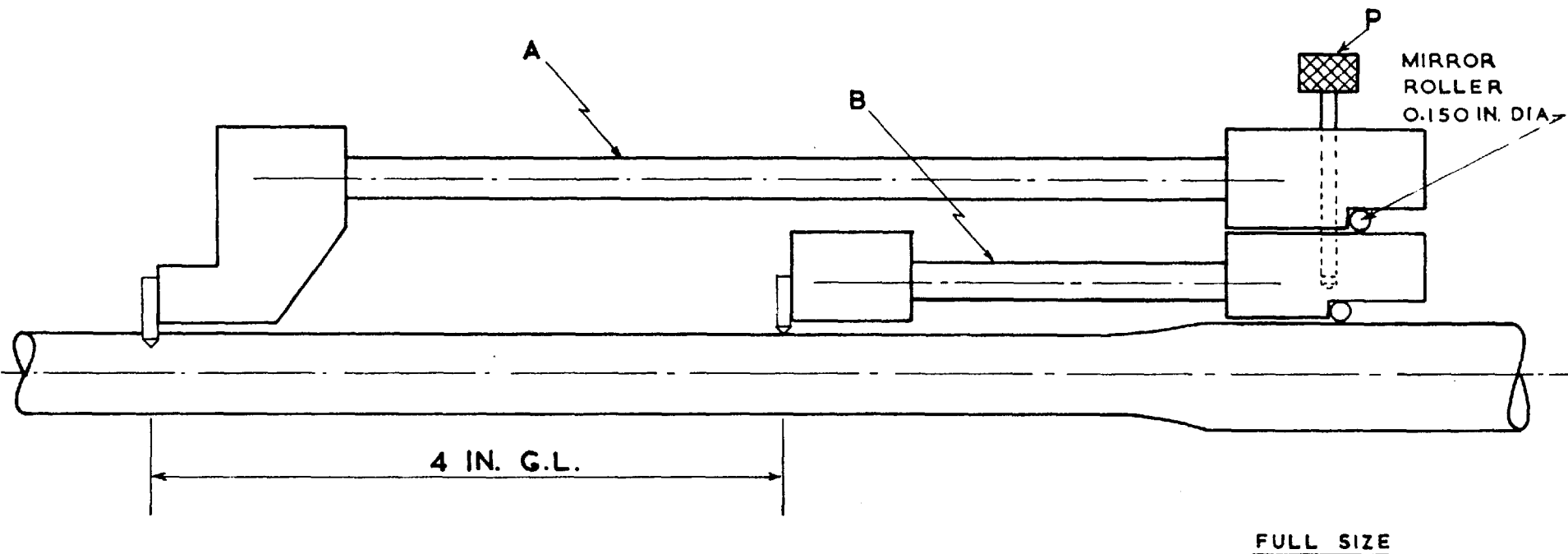
$$E_F = 0.18 \text{ kg.m.}$$

APPENDIX 'G'FURTHER DETAILS OF THE STATIC YIELD STRESS TESTS1. The Triple Extensometer Unit

This unit was designed to measure the extensions on three 4 in. gauge lengths set at 120° to each other around the circumference of the specimen. It was capable of performing this with the entire gauge length immersed in a temperature bath. A single extensometer is shown in Fig. G.1. As shown this has, of course, no means of support, but when all three units were fixed together the clamping force of the springs held the unit firmly in place. The locating pin P ensured that the Vee knife-edge carried by the top part A and the straight knife-edge carried by the bottom part B were initially 4 in. apart. The three gauge lengths were checked with a travelling microscope and found to be accurate within ± 0.001 in. A mirror was carried on the accurately ground roller R which separated the end blocks of A and B. Once the extensometer was set up the locating pin could be removed. Any extension of the gauge length caused relative motion between A and B, thus rotating the mirror. The change in temperature between the room temperature when the unit was assembled and the testing temperature caused a differential expansion (or contraction) between the two parts. Those parts of A and B above the liquid level suffered the same amount of expansion, but the greater immersed length of steel tube in part A caused part A to expand more. The whole of any

$$\text{Extension} = \frac{1}{2} \left(\frac{d}{D} \right) S$$

where d = Roller Dia.
D = Mirror - Scale Distance
S = Scale Reading



ONE UNIT OF THE TRIPLE MIRROR-EXTENSOMETER

one test, however, was carried out at a constant temperature and thus the only effect of the differential expansion was (1) a slight change in gauge length (less than 0.1% for 100°C.) which had a negligible effect, and (2) the necessity for a zero adjustment of the mirrors before the test started, which was usually necessary in any case. The entire unit was made as light as possible, the end blocks being made of duralumin connected by thin stainless steel tubes. The knife edges, however, were made of hardened steel.

To assemble the unit the two parts of each extensometer were clamped together temporarily with the locating pins in place. Special spacers were employed to set the extensometers in place around the specimen and then the nine springs were clipped over dowels on the end blocks of the extensometers. The clamps and spacers could then be removed and the mirror rollers and idler rollers carefully slipped into place. The assembly is shown in Plate (11). Also shown in this photograph is a device which enabled the whole unit to be rotated through any desired angle. The two parts of each extensometer were clamped together as before, using small tool-maker's clamps and the graduated disc lowered on to them. Slots had been milled in the underside of the disc and the tops of the extensometers were a sliding fit in these slots. Thus rotating the disc rotated all three extensometers at the same time, the depth of the slot ensuring that there was no twisting

of the extensometers during rotation. By this means the position of the unit could be set to within $\pm 2^\circ$. When not in use the disc screwed on to the upper part of the specimen as shown.

Three telescopes and scales were used to measure the rotation of the mirrors and hence the extension on each of the three gauge lengths. For small angles of rotation the extensions could be measured to an accuracy of 10^{-5} in. The method of calculation is shown in Fig. G.1.

2. Horrison's Method to Determine the Plane of Maximum Strain (106)

Let AB be a diameter of the specimen (Fig. G.2) such that A is the position of the minimum strain and B that of the maximum. Let C, D and E be the positions of the three extensometers, set mutually at 120° , and let the measured strains at these points be x , $(x + y_1)$ and $(x + y_2)$ respectively. The triangle MNQ shows the distribution of strain across the specimen.

The average strain is $1/3(3x + y_1 + y_2)$ or $x + 1/3(y_1 + y_2)$ and the maximum strain will be $x + 1/3(y_1 + y_2) + \frac{1}{2}h$.

Where h is the height NQ as shown.

$$\text{But } \frac{h}{MN} = \frac{y_1}{LO} = \frac{y_2}{LP}$$

$$\text{and } LP = V \sin. \theta$$

$$LO = V \sin. \theta + V \sin.(60 - \theta) = 1/2 V \sin. \theta + 3/2 V \cos. \theta$$

$$MN = \frac{2V}{\sqrt{3}}$$

$$\begin{aligned}\text{Therefore } h &= \frac{LN}{LP} y_2 = \frac{2}{\sqrt{3}} (1/\sin. \theta) y_2 \\ &= \frac{2}{\sqrt{3}} y_2 \operatorname{cosec}. \theta = \frac{2}{3} \sqrt{3} y_2 \operatorname{cosec}. \theta \quad \text{---(G.1)}\end{aligned}$$

$$\text{Also } \frac{y_1}{y_2} = \frac{LO}{LP} = \frac{\frac{1}{2} V \sin. \theta + \frac{\sqrt{3}}{2} V \cos. \theta}{C \sin. \theta} = \frac{1}{2} + \frac{\sqrt{3}}{2} \cot. \theta \quad \text{---(G.2)}$$

Using the transformation $1 + \cot.^2 \theta = \operatorname{cosec}.^2 \theta$ equation (G.2)

may be rearranged to give $\sqrt{3} y_2 \operatorname{cosec}. \theta = 2(y_1^2 - y_1 y_2 + y_2^2)^{\frac{1}{2}}$

combining with equation (G.1) gives:-

$$\frac{1}{2} h = \frac{2}{3} (y_1^2 - y_1 y_2 + y_2^2)^{\frac{1}{2}}$$

$$\therefore \text{Maximum strain} = \text{Average strain} = \frac{2}{3} (y_1^2 - y_1 y_2 + y_2^2)^{\frac{1}{2}}$$

3. Warnock and Brennan's Construction for the Theoretical Upper Yield Point (107)

Whilst the lower yield point of mild steels is in general unaffected by the shape of specimen the upper yield point is regarded as an experimental phenomenon. If the stress concentration due to the shape of the specimen or to the eccentricity of loading is just greater than that due to local yielding, then the upper yield point will vanish and the stress at the limit of proportionality will be equal to or less than the lower yield stress.

Fig. G.3 has been exaggerated to illustrate the stages in the deformation of a mild steel in tension up to the end of plastic yield. The rate of loading is assumed to be so slow that there is no speed effect.

The graph EABC represents the ordinary tensile test in which little precaution has been taken to ensure axiility.

From A to B elastic deformation only takes place. At A, the limit of proportionality in this case, local yielding commences and produces a condition equivalent to a stress concentration at the yielded portion. At B general plastic yielding commences and continues to C. By reducing the stress concentration due to the specimen shape etc. the stress-strain curve may be raised to some point F, or with great care still further to G. At F and G the maximum stress in the specimen has reached a value sufficient to commence the propagation of yield along the gauge length, and the external stress falls rapidly to B where general yield may be said to commence.

The ideal situation with no stress concentration at all is shown by the straight line EAD. At A, the theoretical limit of upper yield the uniform stress in the specimen has the same value as the maximum stress due to the stress concentrations at points F, G and B.