BAKE-HARDENING OF DUAL-PHASE STEELS

by

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SUMMARY

The influence of bake-hardening on the flow strengths and dent-resistance of two lean-alloy dual-phase steels which had been lightly prestrained in uniaxial tension, plane strain and balanced biaxial tension were investigated. 0.5% proof stress and UTS values were measured at 0° and 90° to the direction of maximum extension in prestraining after ageing at five different temperatures in the range 20°-220°C for 30 minutes. The proof stress values of the unidirectionally dual-phase steels showed much greater directionality than that developed in ordinary batch annealed, low-carbon steels. With the as-received dual-phase steels, which had been preaged at 260°C for one minute, substantial increases in flow strength were developed by bake-hardening above 135°C but after ageing at 180°C the directionality in proof stress was only partly removed. Bake-hardening at 220°C was required to remove most of the directionality. In sheets stretched in equibiaxial tension the changes in 0.5% proof stress values with bake-hardening were essentially similar to those observed with the 90° tests on the unidirectionally stretched sheets. The influence of variations in ageing potential was examined by re-heat treating with preageing treatments, usually at 200°, 300° and 400°C for 3 minutes. These tests showed the expected reductions in bake-hardenability after preageing at 300° and 400°C and, with these higher preageing temperatures, the directionality of proof stress values was not removed even by bake-hardening at 220°C.

Static dent resistance was investigated on similarly prestrained and bake-hardened flat panels and also on curved panels stretched in equibiaxial tension. The results show that the influence of bake-hardening on dent resistance is particularly important in relation to the formation of shallow dents in flat panels. The results as a whole suggest that measurements of yield strength in tensile tests will not always provide reliable predictions of relative dent resistance.

KEY WORDS: Dual-phase
bake-hardening
static dent resistance

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NOTATIONS

BE	Bauschinger Effect
BH	Bake-hardening
DPL	Inland Steel Company's designation for dual-phase steels
$M P_a$	Mega-Pascals
N	Newton
P	Load
PS	Proof Stress
α	Ferrite
γ	Austenite
σ	Stress
ε_1	The maximum principal strain in the sheet plane
0°	Longitudinal direction
90°	Transverse direction

CHAPTER 1

INTRODUCTION

The development of high strength low-carbon steels in the past two decades has been important in relation to the requirement for fuel economy in automobiles. Low-carbon dual-phase sheet steels are of particular interest for the manufacture of body pressings which require good stretch forming properties, because they offer attractive combinations of strength, ductility and formability. Dual-phase steels can be processed to give good strain hardening properties and high ratios of yield strength to ultimate tensile strength. However, because of the rather flat shape of many body panel pressings the strain developed in the central area of the pressings is usually less than 5 percent. Even with dual-phase steels the increment in strength by strain hardening is not large. Thus the reduction in thickness of the panels, relative to those made using ordinary low-carbon steels, is limited by strength requirements and particularly by dent resistance which is sensitive to sheet thickness. For this reason there has been considerable interest in the extent to which panel strength and dent resistance can be increased by strain age-hardening after pressing at the elevated temperature used for paint-baking. This ability to increase strength during ageing at temperatures of about 170°-$180^{\circ}C$ is usually called bake-hardenability. The bake-hardenability of continuously annealed steels which are cooled rapidly from the annealing temperature can be varied by variations in the overageing treatment which is applied to reduce dissolved carbon and in this way prevent ageing at room temperatures.

In the past bake-hardenability has usually been investigated using uniaxial tensile tests in which the directions of prestraining and test straining are the same. Such tests give no information about directionality in strength properties. Directionality is more strongly developed in prestrained high strength steels than in softer low-carbon steels and it is expected that such directionality will be reduced during bake-hardening.

The investigations described in this thesis were made in order to identify the effects of bake-hardening on the tensile strength and its directionality in two lean-alloy dual-phase steels which had been lightly pre-strained in different modes of stretching. In addition to the industrially processed condition, tests were made after overageing at several different temperatures in order to vary the bake-hardening potential of the steels. In the second part of the research the static dent resistance of similarly lightly stretched sheets was examined so that the effects of a wide variety of bake-hardening treatments on the tensile properties and dent resistance could be compared.

CHAPTER TWO

LITERATURE REVIEW

2.1 Dual-Phase Steels

Among the many different high strength low-alloy steels which have been developed in the past ten years, dual-phase low-carbon steels are regarded as the most promising for application in the stretched-formed panels which are being used in some light-weight motor cars because they can provide the best combinations of strength and stretch formability(1). Their excellent strength-ductility combination has opened up the possibilities of utilising thinner gauge, higher strength steels in a number of severely stretch formed parts where component weight is of great importance.

Dual-phase steels consist essentially of a dispersion of moderately high-carbon martensite in a fine-grained ferrite matrix. Dual-phase may be produced by a variety of processing routes. A number of recent studies(2) have shown that the detailed nature and dispersion of both the hard and soft phases can vary with changes in composition and processing route. The structure is developed by inter-critical annealing in the $\alpha + \gamma$ region of phase equilibrium followed by cooling rapidly enough to transform most of the austenite to martensite. A typical microstructure of a low carbon dual-phase steel designed for cold forming consists of an alloyed ferrite matrix which is dispersion hardened with 10 to 20% of martensite or lower bainite[3], containing a small proportion of retained austenite[4]. The inter-critical annealing temperature is selected to provide the required proportions of the two principal phases and the alloy content must be sufficient to suppress the pearlite reaction on cooling so that, preferably, martensite is formed by low-temperature transformation of the islands of austenite[5]. The principal alloying elements are usually manganese and silicon but small additions of more expensive elements such as V, Mo and/or Cr may be required to match the transformation characteristics to available cooling rates[6].

In terms of mechanical behaviour, the dual-phase steels exhibit combinations of tensile strength and total elongation better than those attained by other high strength sheet steels[7]. It has been found that to a good first approximation the strength of dual-phase steel is linearly proportional to the percentage of martensite in the structure[8]. It has also been shown that the tensile strength of dual-phase steel is approximately proportional to both fatigue[9] and crash resistance[10].

Since the dual-phase structure is obtained as a final heat-treatment step, these steels can be obtained over a wide range of thicknesses. The thicker gauges of dual-phase steels are produced from hot-rolled strip and the thinner gauges by final cold rolling and continuous annealing. For such materials hardenability (of the small austenite patches) becomes an important factor which must be matched to the available cooling rate and thus alloying additions such as Mo, Cr, etc, have been investigated[11,12].

Since the dual-phase steels in common use consist of at least 80% ferrite, it is to be expected that the properties of this ferrite will have a major influence on the work hardening of the steel. The ferrite in dual-phase steels is usually fine-grained and preferably low in interstitial elements and precipitates, and often strengthened by the addition of substitutional alloying elements such as Si and/or P[13]. The other 20% or less of the structure is usually of a high carbon (up to 0.6% C) martensite which may contain a small amount of retained austenite[3,14]. Dual-phase steel can provide substantial improvements in stretching performance relative to those of original HSLA steels which are strengthened with the help of carbide dispersions[8,15].

Since the introduction of dual-phase steels in the mid-1970's[16,17] much research has been aimed at defining the microstructural parameters that contribute to their improved strength-ductility balance[6,18]. Several investigations have related the tensile strength of the dual-phase steels to the amount of martensite present[19,20]. Other studies[21,22] provided evidence that the carbon content of the martensite and hence the strength of the second phase, must be considered in assessing the overall strength of these ferrite-martensite composites. Along with the effects of martensite on tensile elongation[21,22] there is a clear trade-off between tensile strength and total elongation in dual-phase steels and several correlations of this type have been developed[19,20,23,24]. Generally an increasing amount of martensite results in reduced ductility but other factors such as ferrite composition or retained austenite are also significant. Transformation of austenite to martensite produces a high dislocation density in the ferrite and a sufficient cooling rate below the MS temperature preserves dislocation mobility enough to ensure continuous yielding. Transformation of as little as 3% martensite can produce continuous yielding behaviour when a sufficiently fast cooling rate is employed below the MS temperature.

Dual-phase steels exhibit a high initial work hardening rate which decreases rapidly with increasing plastic strain[22] and the initial work hardening is increased when either the amount of martensite or its carbon content is increased. Essentially, the work-hardening behaviour is typical of dispersion-hardened alloys but it may be influenced to some extent by the ability of martensite to deform plastically at high applied strains particularly when the carbon content of the martensite is low.

2.2 Strain Ageing

The main feature of strain ageing in low-carbon steels are described in a comprehensive review by Baird[26]. Strain ageing shows itself chiefly by an increase in yield

or flow stress on ageing after or during straining, age-
ing after straining being classed as 'static' and ageing during straining as 'dynamic' strain-ageing. The important change in properties are: return of sharp yield point; rises in the ductile/brittle transition temperature and tensile strength; reduction in ductility.

Following the work of Cottrell and Bidby[27] many of the effects observed can be explained qualitatively as arising from the segregation of solute atoms to dislocations producing pinning of the dislocations by solute atmospheres. However, in the later stages, increased frictional resistance to the passage of unpinned dislocations and a higher strain-hardening rate, may also occur as segregation of solute to dislocations become more extensive. In steel, strain ageing arises largely from the presence of the interstitial solutes, carbon and nitrogen, although hydrogen can also produce a limited amount at low temperatures[28].

Strain-ageing is of commercial importance in a number of ways. It may lead to an undesirable decrease in ductility in cold forming operations, reduce toughness in cold-worked metals, or give rise to unsightly stretcher-strain markings in pressing operations[26]. The changes in tensile properties in a low-carbon commercial steel after tensile prestraining, ageing at 60°C and re-straining in the same direction, may be described in terms of several stages, as suggested by Wilson and Russell[29,30] and are summarised in Fig 1. In stage I the yield stress and the Lüder's strain or yield point elongation rise but other properties remain unaltered. In stage II, while yield stress continues to rise, the Lüder's strain remains roughly constant and the flow stress beyond the Lüder's strain rises but with an unaltered rate of strain hardening. Stage III is similar to stage II except that the initial strain-hardening rate increases, with resultant increase in UTS and decrease in elongation. Stages II and III are referred to as strain-age-hardening since they reflect a permanent hardening rather than a purely yield-point effect as in Stage I. In stage IV overageing effects lead to softening, although the Lüder's strain is maintained and even rises in fine-grained conditions. Increasing the solute level accelerates the ageing processes but also has a particularly marked effect on the magnitude of the changes in Stage II and Stage III.

The general interpretation of the above results put forward by Wilson and Russell was that in stage I dislocation atmospheres are formed, which at the completion of the stage, are sufficient to lock the dislocations fully and hence to re-establish the full Lüder's strain, but which are dispersed on straining through the Lüder's strain as the dislocations are unpinned, so that they have no effect on other properties. In stage I the value of Ky in the Petch relationship[31] $\sigma y = \sigma i + Ky\ d^{-\frac{1}{2}}$ rises and the friction stress, σi , remains unaltered

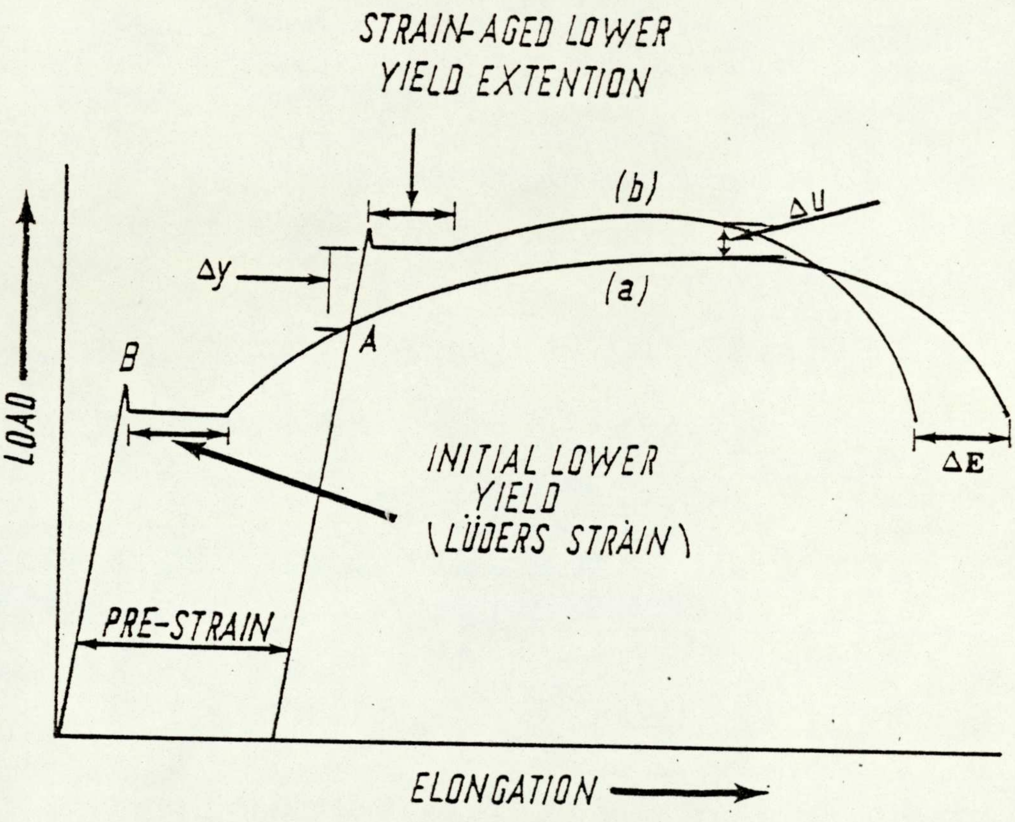


Fig 1(a) Schematic diagram showing the changes in tensile properties of a low-carbon steel caused by strain ageing. Δy and Δu are the increases in yield strength and UTS respectively and ΔE is the reduction in total elongation.(26)

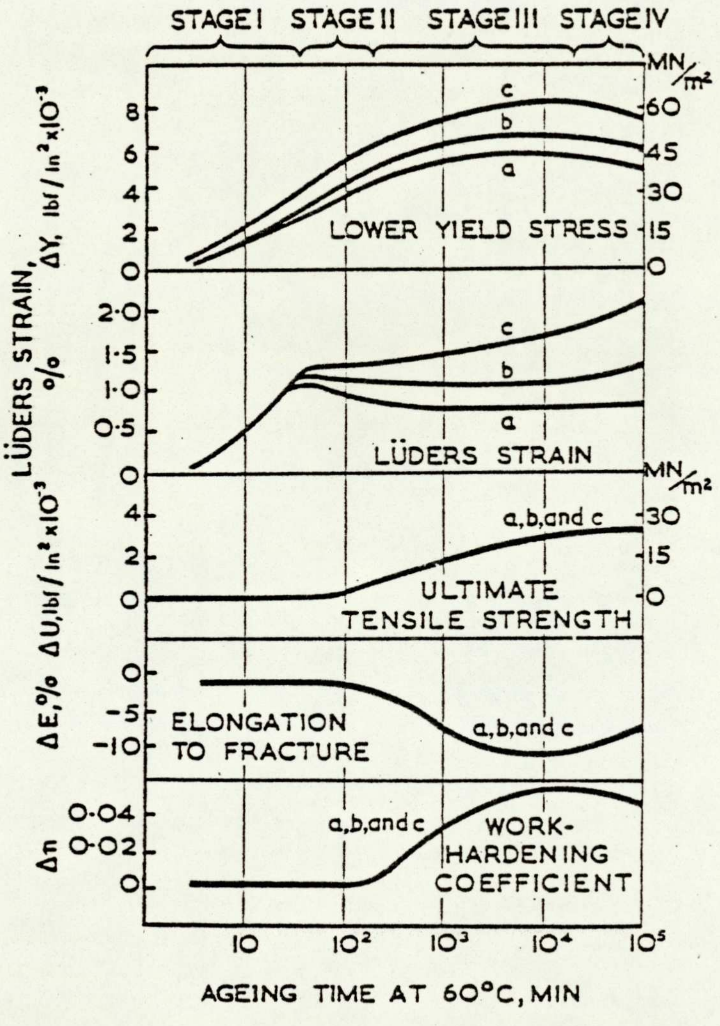


Fig 1(b) Effects of grain size and ageing time at $60^{\circ}C$ on changes in tensile properties due to ageing at $60^{\circ}C$ after 4% prestrain, for a low-carbon rimmed steel containing approximately 0.0022 dissolved carbon plus nitrogen. Grain sizes in grains/mm^2 were: (a) 50; (b) 195; and (c) 1850. Δn is the change in the work hardening exponent derived by fitting the power law relationship $\sigma = k\varepsilon^n$ to the part of the σ/ε relationship observed beyond Lüders strain, after Wilson and Russell (29) (d = average grain diameter). In stage II the permanent hardening reflected by an increase in σi may be due to the pinned arrays of dislocations or to fine precipitates on dislocations[26]. The transition from stage II to stage III probably coincides with the point at which the precipitates on dislocations become of sufficient size to increase strain-hardening.

The effect of increasing ageing temperature is, first to accelerate the development of the above stages without greatly altering the magnitude of the changes that occur; the accelerating effect of temperature can be computed by Hundy's equation using the appropriate activation energy for the diffusing solute[32].

2.2.1 Effect of Amount and Type of Prestrain

It has been shown[26] that the change in yield stress on strain ageing after tensile prestraining is not very sensitive to the amount of prestrain, but that the increase in UTS and decrease in elongation on strain-ageing are larger, the higher is the prestrain. If prestraining and ageing are carried out in increments, there is a larger increase in yield stress than if the same prestrain is effected in a single operation. Of greater interest and importance is the type of prestrain and its relationship with the direction of straining after ageing. Although changes in UTS and elongation are unaffected by type and direction of prestrain, the rate of return of yield plateau is markedly affected. If prestraining is carried out in tension and the direction of straining after ageing is the same as that in prestraining, then the yield plateau returns rapidly on ageing. However, if prestraining is carried out by rolling or if the direction of extension in the test strain is transverse to the direction of extension in prestraining, after ageing, the yield plateau returns much more slowly during ageing.

The apparent retardation of ageing caused by temper-rolling as compared with stretching appears to be confined to the yield-point elongation[33] and it was suggested that this is due to the pattern of oriented micro-stresses imposed by the method of prestraining, since unless the direction of testing after ageing is the same as that of prestraining, the early stages of the return of the yield point elongation are obscured. Wilson and Ogram[34] have carried out an intensive study of this subject which has thrown more light on this problem. They showed in stretched sheets that, in a very coarse-grained steel, no yield plateau appeared even on prolonged ageing at 89^{o}C when the test strain was perpendicular to the prestrain direction. Similar results were obtained when prestraining and final straining were carried out in torsion in opposite directions. They also found that reducing the interstitial content has the same effect as coarsening the grain size. It is noticeable that the later stage of the increase in Lüder's strain, which is obtained particularly at higher solute levels in forward straining, is still obtained in reverse straining and that it is only the earlier stages of yield-plateau return that are affected by the direction of straining.

The conclusions which can be obtained from these results are that, firstly, the return of yield point in 'reverse straining' conditions is controlled by segregation of solute atoms to dislocations or dislocation sources and not to other effects such as recovery or relief of internal microstresses. Secondly, the dislocation sources operative in forward and reverse straining are different; the latter requiring much more extensive segregation of solute to lock them than the former.

In commercial practice for producing deep-drawing sheet steels, suppression of yield-point extension (to avoid stretcher-strain markings) is normally carried out by temper rolling $\sim 1\%$. This is an effective method of eliminating the initial yield plateau by the application of a small strain and of delaying the return of yield plateau on subsequent ageing[35]. At small strains in temper rolling alternate lamellae of deformed and undeformed material are developed on a fine scale. Elimination of the yield plateau is ascribed partly to the effects of multiple nucleation of yield fronts in this structure and partly to the directional effects described earlier[37]. Normally sheet steel is subjected to enough cold deformation in temper rolling to eliminate the yield point. In a rimmed steel, ageing after temper-rolling will eventually cause a return of the yield point. Wilson and Butler[36] studied the effect of temper rolling and strain ageing on formability of a rimmed steel. They found that 5% temper rolling reduced general elongation in a uniaxial tensile test but 1.2% rolling reduction did not; both these temper rolling reductions decreased performance in biaxial stretch forming and deep drawing tests. Strain ageing after 1.2% rolling reduction reduced uniaxial tensile ductility and stretch formability. After 5% rolling the effect of ageing on the stretching properties were more severe. They suggested that the change in stretching performance due to temper rolling and strain ageing can be explained qualitatively in terms of their effect on work hardening rates.

2.2.2 Carbon and Nitrogen

Carbon and nitrogen have very similar diffusion coefficients in steel[26] and produce identical distortions of the ferrite lattice[37]; hence they would be expected to produce very similar strain-ageing effects in steel when present in solution in the same amounts. The main differences between the strain ageing effects of carbon and nitrogen arise from their widely differing solubilities in steel. The solubility of nitrogen is higher in the temperature range in which rapid precipitation can take place (say, $> 200^{o}$), but at $200^{o}C$ the solubility of carbon in equilibrium with cementite has already fallen to $< 10^{-4}\%$. As a result, provided that well-dispersed nuclei are present onto which carbon can precipitate, carbon strain-ageing is normally negligible at room temperatures in slowly cooled steels[38]. However, on ageing above 100°C there is evidence that fine carbide particles (especially ε carbide) can redissolve to cause strain-ageing[39,40,41]. Nitrogen strain ageing is generally considered to be fairly directly related to the 'free' nitrogen content i.e. the nitrogen not combined with strong nitride formers, and to be much less dependent than carbon strain-ageing on thermal treatment in the ferritic range. However, there appear to be some exceptions to this generalization. Quench ageing at 100°C or slow cooling below 250°C reduces the strain-ageing rate somewhat at high nitrogen levels[42,43].

2.2.3 Effect of Other Alloying Elements

It has been suggested[26] that elements going into solution in ferrite with no affinity for carbon and nitrogen have little effect on strain ageing, whereas those with some affinity for carbon and nitrogen probably slow down strain-ageing. Silicon and manganese are known from internal friction studies[44,45] to interact with nitrogen in solid solution, although under certain conditions, they can also precipitate nitrides[46,47]. Under conditions where they would be largely in solution, both silicon and manganese have been found to retard strain-ageing[48]. The effect of manganese is particularly important from a practical point of view in low-carbon steels, since in such steels the effective manganese content is that in solid solution, which is given by the total manganese content less the amount tied up in oxides and sulphides.

2.2.4 Nitride Formers

Since strain-ageing below 100°C in slowly cooled steels is largely due to nitrogen, low-temperature strain-ageing can be reduced to a low level by nitride formers provided that the earlier heat treatment is such as to precipitate nitrogen fully as alloy nitrides and that carbon is allowed to precipitate fairly completely during final cooling. The influence of prior-heat treatment in the case of aluminium-containing steels has shown that strain-ageing rate in the as-rolled condition may be almost the same as in a similar aluminium-free steel, but the ageing rate is much lower after annealing treatments which cause the nitrogen to combine with aluminium as aluminium nitride[49].

Some strong nitride formers are also strong carbide formers, for example, zirconium appears to form a nitride in as-rolled steels before combining extensively with carbon[50] and titanium is expected to have a similar effect. Vanadium also probably forms nitrides. In the case of weaker nitride formers, heat-treatment in the upper ferrite range is necessary to precipitate alloy nitrides sufficiently fully to eliminate strain-ageing.

2.2.5 Formability

The effect of strain ageing on cold formability is of particular importance in the use of sheet steel in press forming applications e.g. in the pressing of car-body panels. Three types of deformation occur in such pressing operations; deep drawing, stretch forming, and bending (usually under tension)[51]. In sheet steels which are not fully stabilised strain ageing may occur at ambient temperatures after temper rolling to suppress the sharp yield point extension (and/or to improve shape and surface texture).

Pure deep drawing in sheet steel is governed largely by the R value[52] and as strain ageing has no effect on the R value it does not alter deep drawability significantly after normal temper rolling reductions[35,52] but performance under stretch-forming conditions is generally reduced by strain ageing. In the unclamped Erichsen test, where a certain amount of draw-in of the blank occurs, an appreciable drop in performance takes place when the yield point returns; this is attributed, at least in part, to the reduction in the amount of draw-in, but a further drop may occur if the elongation is reduced[36]. Similar behaviour is found in some types of commercial pressings[53]. In a firmly clamped stretch-forming test using a hemispherical punch or hydraulic pressure, where no draw-in occurs, the strains are large throughout the deformed region of the blank and the drop in ductility in the later stages of strain ageing is the important factor controlling performance[35,51,53].

In general, return of the sharp yield point will have the largest effect in pressings where the minimum strain is near zero, but the effect of strain ageing on ductility, which is important at high strains, will increase rapidly with increasing temper-rolling reductions.

2.2.6 Commercial Control of Strain Ageing

As indicated in the previous section, the control of strain ageing is particularly important in the case of temper-rolled sheet steel for pressing or cold-forming applications; where strain ageing can give rise to undesirable stretcher-strain markings and to a reduction in stretch formability. On the other hand, strain ageing after pressing can provide a valuable strengthening of a pressed panel with an improvement to its resistance to denting[54]. The ideal sheet steel for pressing applications is therefore one in which the strain-ageing in the temper-rolled condition is low at room temperature, to restrict the changes in properties before pressing, but which strain-ages appreciably after pressing or during any subsequent heating operation required for paint-baking.

2.2.7 Strain-Ageing of Dual-Phase Steels

Dual-phase steels are usually aluminium-killed but rapid cooling increases the amount of dissolved carbon relative to that in batch annealed steels. Dual-phase steels can be considered to provide a somewhat different environment for the diffusion of interstitial atoms than that in the ferrite-carbide structure of ordinary low-carbon steels. Dislocation density in dual-phase steels appear to be a maximum in the ferrite-martensite interface[55] while dislocations are more randomly distributed in ferrite-carbide structures. The effect of prestrain on the ageing of dual-phase steels must also be considered in relation to the residual stresses which are developed by the volume expansion associated with the formation of martensite within the ferrite matrix[55]. Krupitzer[56] found that in a gas jet cooled dual-phase steel containing 0.075 V and 1.40 Mn, at room temperature no significant change in properties occurred for at least one year. Although this stability is not unusual for aluminium killed batch-annealed mild steels, it is a significant observation for a continuously annealed product, especially for one not processed with an over-ageing thermal cycle. In sub-critically annealed steels rapid cooling after annealing can result in super-saturated carbon in the ferrite and can cause appreciable room temperature strain-ageing[26,57] in otherwise 'non-ageing' steels. In the gas cooled V-Mn dual-phase steel, Krupitzer observed significant increases in yield strength and Lüder's strain after ageing at elevated temperature and the increment depended on temperature and time. However, the changes in yield phenomena were generally associated mainly with the classical stage I of strain-ageing related to atmosphere formation at dislocations.

It has been shown that, in as-quenched dual-phase steels, dislocation density is relatively low in the ferrite except near the ferrite-martensite interfaces[55] where dislocations form on cooling because of the volume expansion associated with the martensite transformation. Rapid cooling is believed to allow insufficient time for pinning of dislocations by interstitials and therefore they remain mobile after annealing. Similarly, micro-residual stresses associated with the volume expansion of the martensite, contribute to a low value of the as-annealed yield strength. Dual-phase steels are invariably treated with aluminium and carbon, not nitrogen, is mainly responsible for the ageing. A parallel can be drawn between the ageing behaviour of temper-rolled plain carbon steels and the ageing of dual-phase steels.

Krupitzer[56] concluded that, in dual-phase steels, many of the characteristics of temper-rolled steels exist immediately after annealing and rapid cooling; particularly significant is the fact that the return of Lüder's strain is noticeably retarded in as-annealed dual-phase steels such that it may not occur at all at room temperature and very sluggishly at temperatures up to 260°C. On the other hand, a slight tensile prestrain radically accelerates the development of Lüder's strain so that slight yield plateaus return even in ageing at room temperature. In a sense, then, dual-phase steels behave like temper-rolled (plain carbon) steels for which it has been demonstrated that retarding effects of rolling can be eliminated by a tensile prestrain.

It is clear that the diffusion of interstitials (mainly carbon) are responsible for the ageing which occurs in dual-phase steels[26]. The process is not a simple one; the strain-ageing of dual-phase steels suggests that, in the as-annealed condition, micro-residual stresses are present, probably as a result of the volume expansion of martensite during transformation on cooling. Tensile prestrain will significantly alter this residual stress distribution to a directional one and allow the easy formation of Lüder's strain on ageing if the direction of test is the same as the prestraining direction. It appears that rolling prestrain of dual-phase steel develops residual stresses which have different directional characteristics from those developed by tensile prestraining and causes further retardation of Lüder's formation in subsequent tensile tests.

Tempering of dual-phase steels changes the properties of both martensite and the ferrite which can lead to a degradation of the mechanical strength and changes in formability. It is expected that the lean and more highly alloyed dual-phase steels will respond differently to tempering. The amount of supersaturation in the ferrite and martensite will be a function of the cooling rate from the intercritical annealing temperature; the faster the cooling rate the greater will be the supersaturation. It has been observed by R G Davies[58] that a dual-phase steel, when water-quenched, exhibited reduced elongation but a low temperature (130°C) temper restored the ductility.

2.3 Directionality

When sub-critically annealed low carbon steel containing uncombined nitrogen or dissolved carbon is deformed plastically and then aged before continuing straining in the same direction, ageing for less than a day at room temperature is usually sufficient to restore a sharp yield point[34]. This rather rapid effect of dislocation repinning is consistent with the theory of Cottrell and Bilby[27]. However, if the direction of re-straining is different from that of the first strain, the sharp yield point is found to return very slowly, if at all[36,59]. Such directionality is of considerable practical importance, for example, it contributes usually to the slow return of the yield point in temper rolled sheets[33,36] but it can also limit the application of strain-ageing as a strengthening mechanism[60].

The work of Tipper and Tardif and Ball[33,59], using tensile test pieces cut transversely to the direction of the original stretching, showed that the ageing times required to restore the yield point were so long that solute segregation to dislocations was thought to have been almost completed before a yield point emerged. It was suggested by Hundy[32] that directional internal stresses, developed in prestraining, acted with the applied stress to promote easy dislocation unpinning in transverse tests. A difficulty with this explanation is that σ_p, the local stress required to unpin fully locked dislocations, is expected to be large compared with the internal stresses affecting dislocations which were mobile before ageing. When a test piece is unloaded after prestraining 'back' stresses will cause mobile dislocations to bow out between local obstacles in the opposite direction to that of their movement in prestraining. To this extent there is a component of internal stress acting on potential dislocation sources which can assist yielding in reverse straining, but the effective magnitude of such a stress must be less than that of the flow stress at the end of prestraining. Residual stresses of this kind contribute to the Bauschinger effect. Higher residual stress could affect immobile obstacles in the prestrained structure, such as grain boundaries and dislocations which remain locked, but these stresses are expected to be of the same sign as prestrain stresses[61]. This difficulty can be avoided if it is assumed that nucleation of yielding occurs with the help of stress concentrators, such as second-phase particles and that the effectiveness of stress concentrators in a strain-aged structure depends on the direction of restraining. Kennet and Owen [62] suggested that inclusions elongated during rolling could promote such directionality. However, a more general explanation in terms of the effects of local stress concentrators is possible. Stress concentrators which operate in prestraining will be relatively ineffective on restraining in the same direction because the elastic stress fields have been changed to plastic strain fields[63]. The same, or differently orientated, stress concentrators can be effective when the direction of straining is changed. Indeed, they could become more effective in activating dislocation sources in a prestrained structure.

Following the observations that rather high dislocation densities are built up at grain boundaries and other strong obstacles after a few percent prestrain, Ogram and Wilson[64] suggested that the dislocation sources which are active in reverse deformation are not necessarily strongly locked after ordinary periods of strain-ageing. They argued that if dislocation locking were effectively complete before the transfer yield point was restored, restoration must depend on a recovery process, rather than on strain-ageing. In this event, return of the transverse yield point should be accompanied by a reduction in flow strength and the magnitude of the Bauschinger effect; also increasing the dissolved carbon content might be expected to delay recovery, since precipitation on dislocations could inhibit dislocations rearrangement. On the other hand, if restoration of the transverse yield point depends on a progressive increase in the strength of locking of the dislocations which were sources in reversed deformation, then restoration could occur independently of a recovery process. In this case, increasing the dissolved carbon content would be expected to increase the rate of restoration of the transverse yield point. Wilson and Ogram(64) showed that the effect of variations in dissolved carbon on the behaviour of a low-carbon steel which was tested in reversed deformation supported the latter explanation.

2.4 Bauschinger Effect

When work hardenable materials are deformed, first by forward and then reverse loading, they usually exhibit a decrease in the reverse yield stress. This decrease of yield is identified as the Bauschinger Effect (BE). It is generally believed that this effect is caused by internal stresses that develop as a result of inhomogeneous deformation(65).

For quantitative assessment of the BE it is convenient to plot both forward and reverse stress in the positive direction as a function of cumulative strain. Fig 2 shows

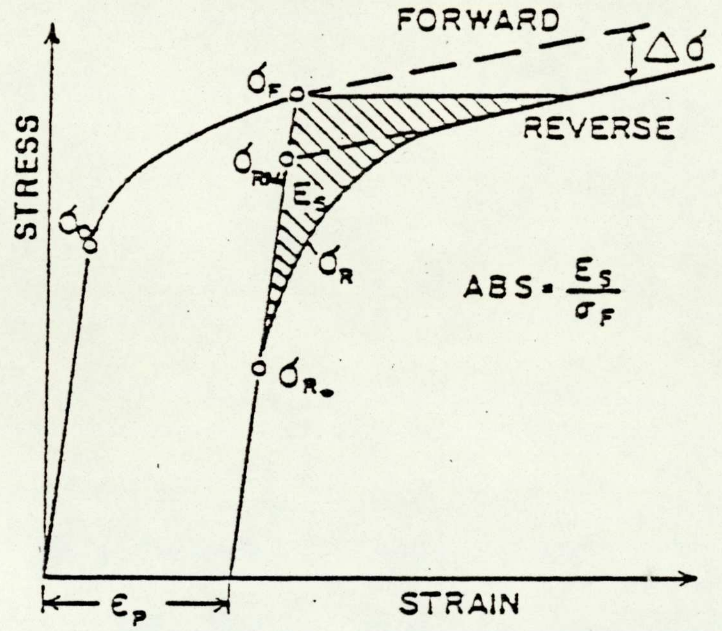


Fig 2. Schematic illustration of the flow curve for forward and reverse straining. E_s is the area between the elastic unloading curve and the reverse curve up to a stress equal to the forward stress. ABS is the average Bauschinger strain, $\Delta\sigma$ is permanent softening, σ_O is the initial yield stress, σ_{RO} is the yield stress on reverse flow, σ_{Rm} is the reverse flow stress extrapolated from the parallel section of the forward and reverse flow curves.(65)

the significant stresses and strains which are used in evaluation of the BE. The curve shows the typical reduction in yield stress in reversal straining and the well-rounded nature of the initial plastic portion of the flow curve. In some instances the reverse curve remains below that for continued forward straining and eventually becomes parallel with the forward curve. The stress increment corresponding to the parallel displacement is termed 'permanent softening'. Orowan[66] suggested that the rounding of the initial portion of the curve was caused by weak, permeable obstacles to dislocations and that the permanent softening resulted from back stresses. He also suggested that the back stress should vanish after about 1 to 3% reverse plastic strain. This has been verified by Wilson[67] using X-ray diffraction to estimate the mean internal stress. He found that the stress difference $\Delta \sigma_p$ between the forward and reverse curve was approximately twice the mean internal stress which existed prior to reverse straining.

2.5 Dent Resistance

A significant reduction in the weight of automobiles would be made by substituting thinner, high strength steels for the outer body panels. However, the thickness reduction of outer panels used, for example in the door, causes deterioration of panel stiffness and dent resistance. The primary concern is the ability of panels to resist moderate denting.

The yield stress of panels is of particular importance when considering resistance to casual damage such as that from pressing too hard on the panel when slamming doors in normal service. Butler et al[68] did some tests on rimming and stabilised steel and showed that the ageing characteristics of rimming steel offer considerable advantages over non-ageing varieties particularly for shallow dents. They also showed that, after ageing, not only is the dent produced more easily in a stabilised steel panel, but the damage is considerably more severe. The dent is spread over a much wider area and is therefore more difficult to knock out and reblend to the original curves of the panel in repair. This difference in dentability results can be attributed to the beneficial effect of strain-age hardening. Dent resistance is a function of the material's yield strength, modulus of elasticity, thickness and panel geometry or stiffness[69,70,71]. Dent resistance increases with an increase in yield strength and radii of curvature, but decreases with an increase in modulus of elasticity and stiffness.

Yoshida[72] suggested that dent resistance is approximately proportional to $\sigma_p t^2$ where σ_p is the flow stress and t is the thickness. Therefore, in order to achieve a reduction in the thickness of the sheet used without losing dent resistance, a large increase in strength is required. Miyahara et al[73] showed that the deflecting behaviour of the curved panel is different from that of a flat panel. In the case of curved panels a nodal point of deflection (Fig 3) is formed and, as the load increases, this point moves from the centre of dent towards the periphery of the panel[73,74]. In the case of a flat panel, on the other hand, a nodal point of deflection does not form (Fig 3). Y utori et al[74] showed that the load required to form a given depth of dent, P_d, is proportional to t^m where t is the panel thickness. The strength of the sheet had no effect on the value of m. In the previous study on steel panels they showed that $P_{0.1}$, the load to give a 0.1mm depth of dent, is proportional to $t^m \sigma_y$[73] where σ_y is the yield strength of the panel. They found that the value of m was 2.3 - 2.4 for curved panels and 1.4 - 1.6 for flat panels. Y utori et al[74] showed that, in general, the thinner the sheet is and the larger the panel radius of curvature, the easier it is for the dent to occur. However, for a panel subjected to a load P, applied to a denting tool of spherical form, the plastic dent is initiated when distortion under the denting tool reaches the elastic limit of the steel. They derived an empirical formula for the relation between load and deflection in the elastic region where δ = deflection, k,n = constants, t = thickness and P = load.

$$\delta = K_1 \frac{1}{t^n} P$$

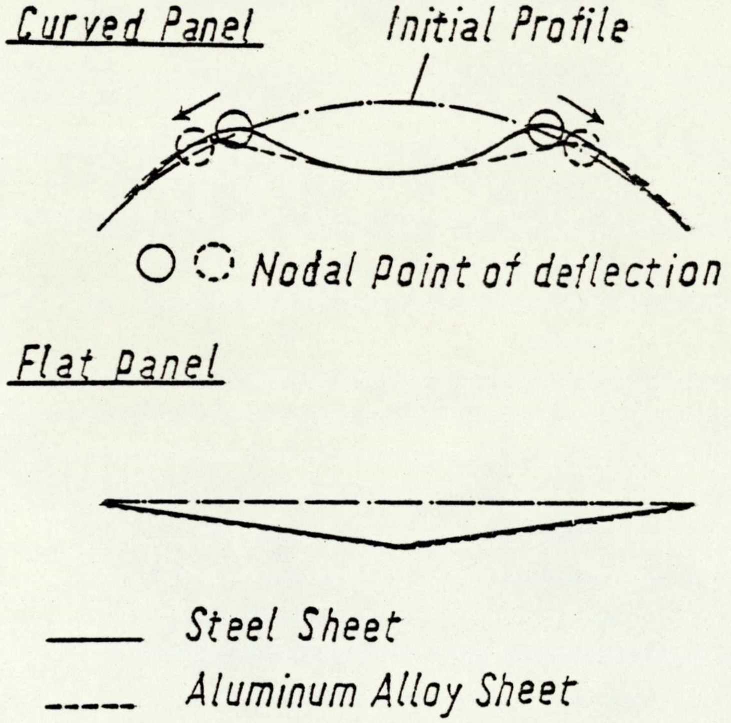


Figure 3 Schematic illustration of the distortions developed in denting curved and flat panels; after Miyahara et al[73].

M J Painter and R Pearce[75] studied the effect of thickness, strain hardening and the rate of straining on dentability of CR1, a dual-phase steel, a HSLA steel and an Al-Cu-Mg alloy. They measured the elastic and plastic components of the dents. The larger the elastic contribution, the greater will be the spring back on removing the load. They also showed that with increasing prestrain of the panel, the resistance of the dual-phase steels improved more rapidly than that of the HSLA steel due to the more rapid strain hardening behaviour of the former.

2.6 The Strength and Dent Resistance of Dual-Phase Steel Panels

Dual-phase steel usually has high initial rates of strain hardening which help to give more uniform distribution of strain, leading to larger limiting depths of pressing than those obtained with other kinds of high strength low-carbon steel. However, because of the rather flat shape of many body panel pressings and the influence of tool friction, high strains are usually developed only in the outer part of typical body pressings. Thickness reduction within the central area of the pressing is often less than 5% and therefore the increase in strength of the central area caused by strain hardening is rather small. After forming, two of the most important characteristics of the part are strength and dent resistance. Considerable interest has developed in the past few years into possibilities of increasing strength by strain ageing at an elevated temperature after pressing[7]. Because motor-car pressings are usually paint-stoved at $\sim$ 170°C for about 20-30 minutes, no additional ageing treatment is required.

In order to develop maximum ductility after rapid cooling from the annealing temperature, dual-phase steels are often over-aged at about 300°C to 400°C for 1 to 3 minutes. This treatment precipitates most of dissolved carbon as cementite particles on a reasonably coarse scale and it leads to a very small potential for strain-ageing in paint stoving after pressing. However, if the short time over-ageing treatment is carried out at 200°C-250°C, it leads to a much finer precipitation of carbide particles which may be mainly of the intermediate precipitate. Removal of most of the carbon from solid solution prevents quench-ageing or strain-ageing in the temper rolled conditions at room temperature[76] but, after stretching, the fine carbides are able to dissolve during heating at about 170°C and this gives a substantial strain-ageing potential at the paint stoving temperature. Such steels are called 'bake hardening' steels and the two dual-phase steels which have been used in the research are representative of current Japanese practice in this respect.

Nearly all the published data on the flow strengths of stretched and bake-hardened dual-phase steels is confined to behaviour in uniaxial tensile tests. Such uniaxial tensile tests give no information about the lowering of flow stress in directions other than that of the tensile extension and about the extent to which bake-hardening can reduce directionality in flow strength. Due to the influence of dispersion hardening by the martensite particles the Bauschinger effects which are developed in dual-phase steels are much larger than those which occur in ordinary batch annealed low-carbon steels. It is known that the Bauschinger effect can be reduced substantially in ordinary low-carbon steels by strain ageing[77].

A principal aim of the present research was to investigate the influence of bake-hardening on directional variations in flow strength which are developed by different modes of prestraining and also the effect of bake-hardening on the dent resistance of dual-phase steels.

CHAPTER 3

EXPERIMENTAL PROCEDURE

3.1 Materials

The two dual-phase steels used in the investigation were supplied by the Inland Steel Company of the United States. They are representative of modern lean-alloy dual-phase materials which are produced using a continuous annealing cycle which incorporates very rapid cooling (in this case water-quenching from the annealing temperature). Such rapid cooling allows a volume fraction of the strong phase of up to about 0.2 to be retained as martensite without the necessity of large manganese or other alloying element additions.

The chemical compositions of the two steels in weight percent were:

steel 60
0.06C, 0.35Mn, 0.013 P, 0.005S

steel 80
0.09C, 0.52Mn, 0.06P, 0.003S

Both steels were killed with sufficient aluminium to remove most of the nitrogen from the solid solution as aluminium nitride. Thus the composition of steel 60 used was similar to that of a conventional aluminium-stabilised steel. The strength of steel 80 was enhanced by small increases in carbon, manganese and phosphorus. The annealing cycle used for the two steels is shown in Fig 4. After the overageing treatment at 260°C the steels

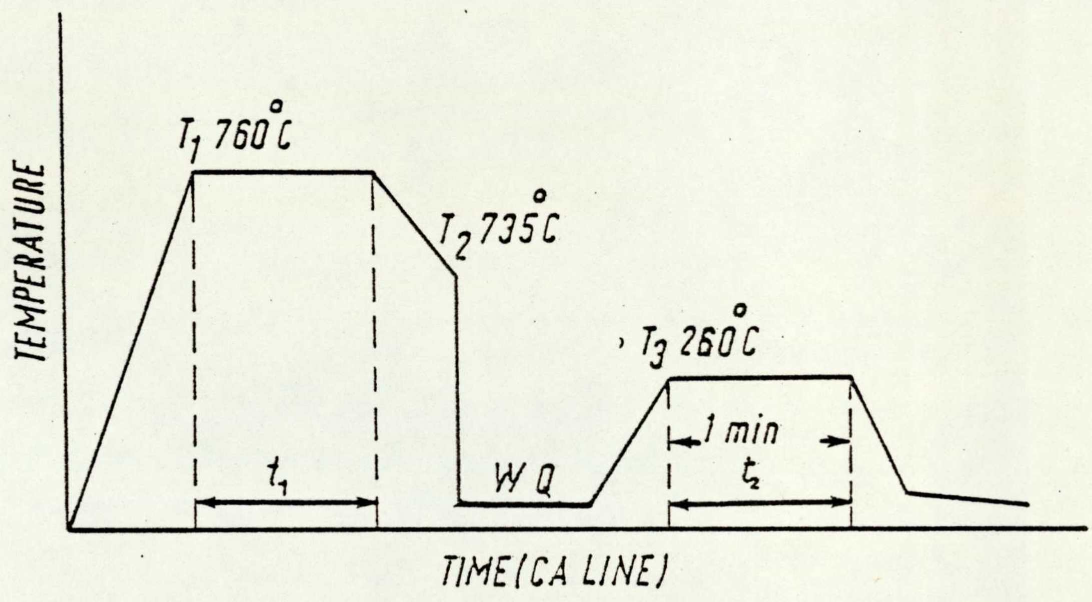


Figure 4 Heat Treatment Cycle

Commercial heat-treatment cycle applied to dual-phase steels 60 and 80.

were temper-rolled $\sim$ 1%. The relatively low overageing temperature of 260°C is selected to produce steels having good bake-hardenability. In comparison with the effects of an overageing treatment at about 400°C, which would be applied to maximise stretch formability without too great a loss in strength caused by tempering the martensite, the overageing temperature of 260°C provides higher initial strength at the expense of some reduction in tensile ductility. Release tests on the commercially processed steels quoted the tensile strength of the 60 and 80 steels as 483 MPa and 620 MPa respectively with total elongation measured on a 50mm gauge length, in the range 20 to 24%.

The average volume fractions of martensite measured by the linear intercept method on several different sheets were 0.11 for steel 60 and 0.14(5) for steel 80. The average grain diameters measured in plane sections were 5.6 μm for steel 60 and 4.6 μm for 80. Typical microstructures showing the phase distributions are given in Figs 5a and b. The samples examined were annealed at 400°C to give the martensite dark etching behaviour.
In the first part of the investigation the tests were made on the dual-phase steels in the "as-received" conditions. To provide a basis of comparison a parallel series of measurements were made on a conventional aluminium-killed low-carbon CR1 steel supplied by the British Steel Corporation. This was a sub-critically box-annealed and temper-rolled aluminium-stabilised steel of the following

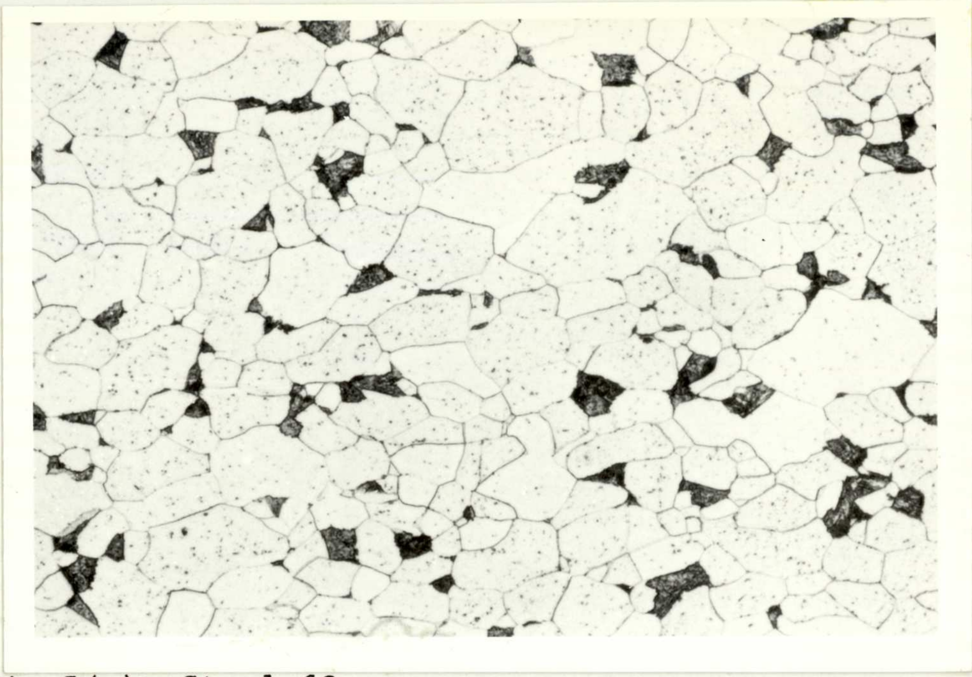


Fig 5(a) Steel 60

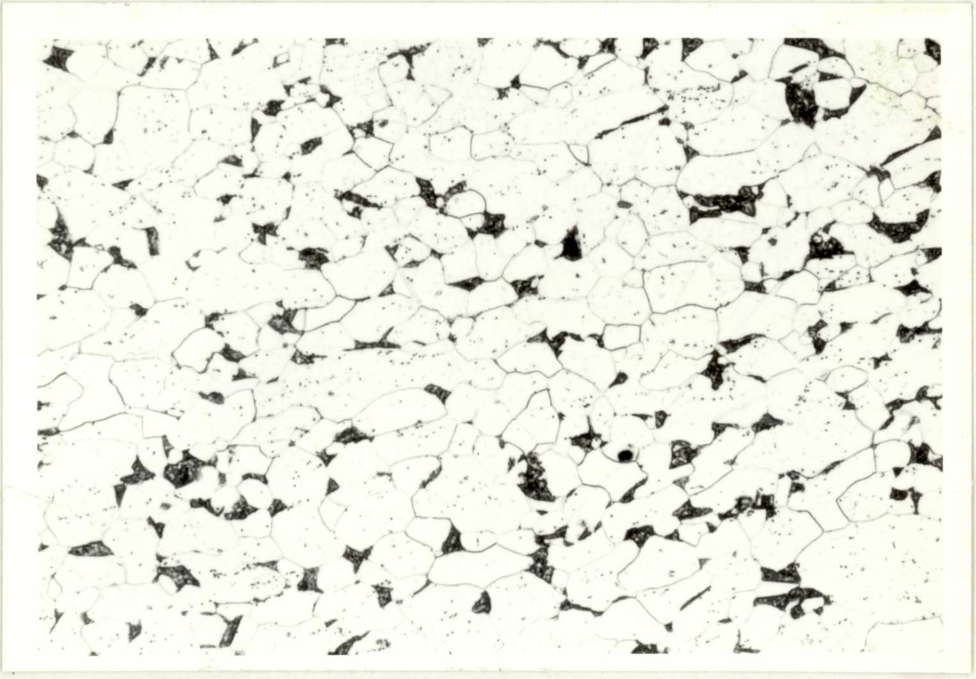


Fig 5(b) Steel 80

Fig 5 Phase distribution in as-received steels 60 and 80. The samples were tempered at 400°C for 3 minutes to darken the martensitic area. Mag X 980.

composition in weight percent.

C	Mn	Ni	Cr	Mo	S	P	Si
0.06	0.25	0.03	0.02	0.01	0.016	0.012	0.02

Figure 6 shows a typical microstructure of the CR1 steel in the "as-received" condition. The average grain diameter in a plane section was 14 μm. The thickness of the three batches of materials were:

Dual phase 60	0.89mm
Dual phase 80	0.89mm
CR1	0.77mm

3.2 Mode of Prestraining

The modes of prestraining by stretching used were uniaxial tension, plane-strain stretching and balanced biaxial stretching. Some tests were also made on cold-rolled sheets.

3.2.1 Stretching in Uniaxial Tension

Uniaxial prestraining was carried out on an Instron 50 ton testing machine using a cross-head speed of 0.5mm/min. The rectangular prestrained sheets were usually 13" long and they had to be at least 6" wide to allow standard tensile testpieces to be cut at right angles to the direction of maximum extension in the sheet plane. The amount of prestraining was measured in the directions of major and minor extension in the sheet plane, using a lightly-scribed grid of 1 inch squares. A large proportion of uniaxial prestrainings were made to give an extension of 4% but a few tests were also carried out with extensions

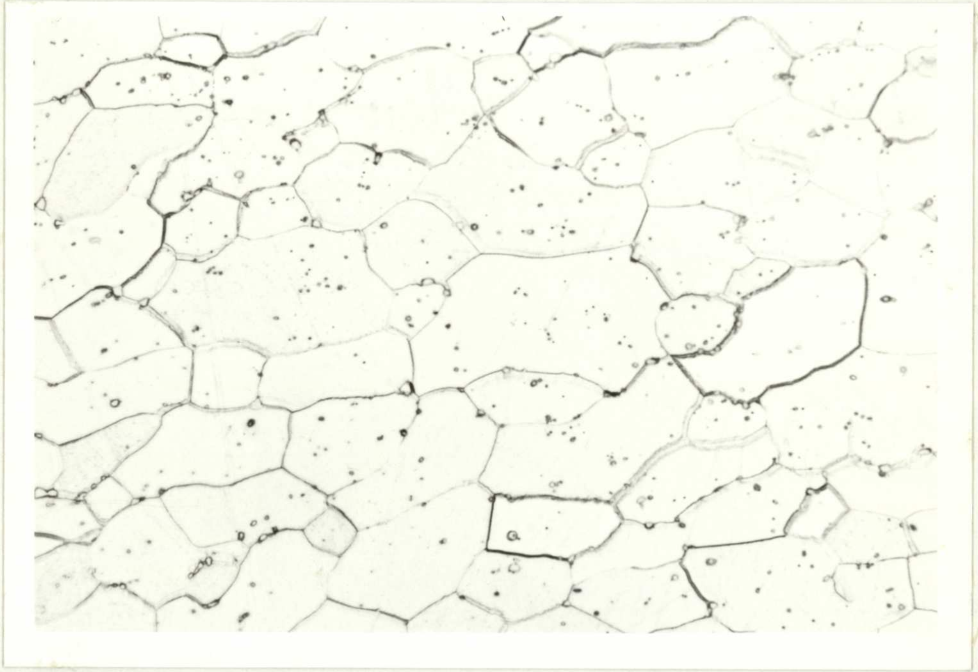


Fig 6 Phase distribution in as-received steel CR1, Mag X 980.

of 8 and 16%. Standard tensile specimens were cut in directions at 0° and 90° to the direction of maximum extension.

3.2.2 Plane Strain Stretching

For plane strain stretching sheets 20" wide by 14" long were bent to form a 6-inch wide box section as shown in Fig 7. One side of the box was spot welded. The box was then fitted into a specially constructed rig which provided lateral constraint by means of an internal former as shown. The extension applied in prestraining in plane strain was 4% in all cases.

3.2.3 Balanced Biaxial Stretching

Balanced biaxial stretching of flat sheets was carried out in the 100 ton Mays hydraulic press over a flat-faced punch of 8 ins diameter with draw-in fully restrained by a locking bead. The amount of prestraining was measured over a 4" gauge length. The linear surface strains applied in this case were 2, 4 or 8%. Tensile test pieces were taken in two directions at right angles.

3.2.4 Prestraining by Cold Rolling

6-inch wide sheets were reduced in thickness by 5 and 10% in a cold rolling mill. Tensile specimens were taken at both 0° and 90° to the rolling direction.

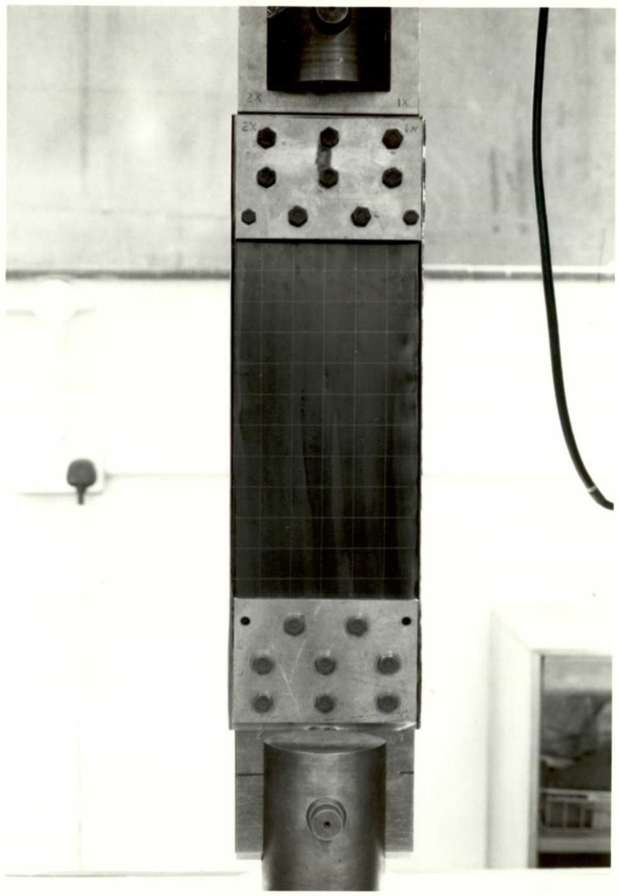


Fig 7 Testpiece and rig used for stretching in plane strain.

3.3 Mechanical Testing

3.3.1 Uniaxial Tensile Testing

Tensile tests were carried out on a 5 ton Instron tensile testing machine model TT-CM in the "as prestrained" condition and after ageing at 90°, 135°, 180° and 220°C for 30 minutes. Specimens were marked with a 50mm gauge length as shown in Fig 8. Specimen width and thickness were measured at three positions along the gauge with an accuracy of 0.005mm using a micrometer. The specimens were subsequently pulled to fracture. Values of 0.5% proof stress were measured using a mechanical extensometer. From the load extension curve values of UTS were obtained and total elongation was measured after fracture. Quoted UTS and proof stress values are both nominal stresses calculated by dividing the observed loads by the initial cross sectional area.

The original aim was to measure proof stress values at both 0.2% and 0.5% offsets. However, in the early programmes of testing it was found that the 0.2% proof stress values were erratic, particularly in the case of the 90° tests. It was concluded that this variation was caused by small differences in ageing during cutting and machining the testpieces from the large sheets. At least for the 90° tests it was not possible to avoid some fluctuations in the 0.2% proof stress values. The 0.5% proof stress values were much less sensitive to small differences in ageing close to room temperature. In the case of 0° tests were made on uniaxially stretched sheets a comparison was made between samples which had been cut

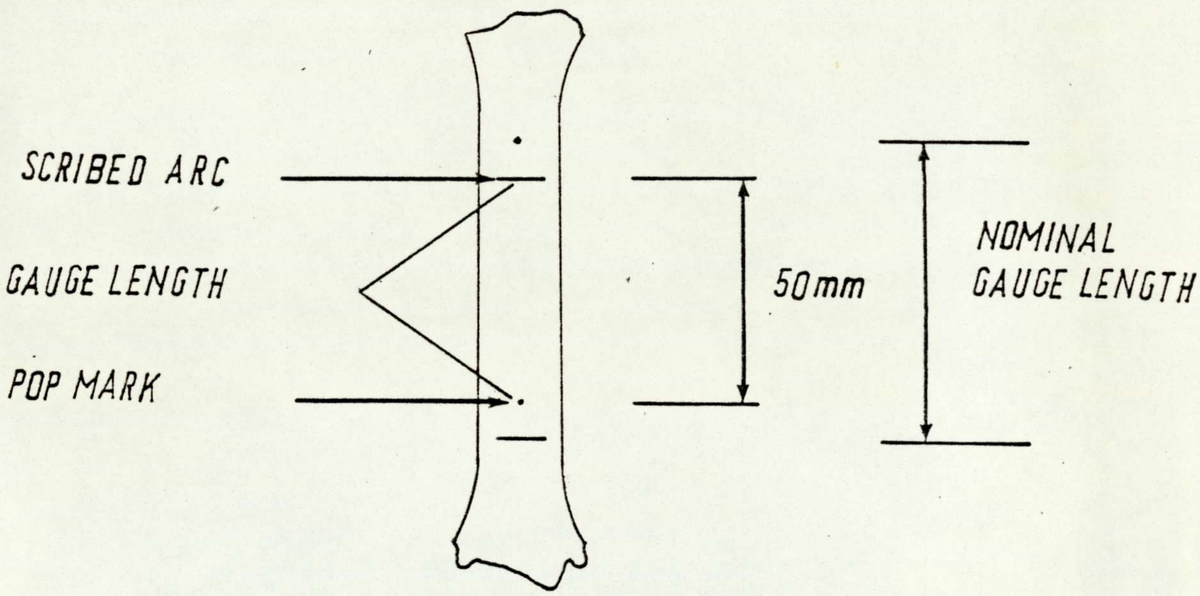


Fig 8. Dimensions and marking of tensile test specimens.

from large sheets and results with pre-machined tensile test pieces which were not machined between prestraining and final testing. In this case the 0.5% proof stress values for the unaged condition were not significantly different in tests made using the two different procedures. Following these observations it was decided to abandon attempts to measure the 0.2% proof stress values of the dual-phase steels and the results for dual-phase steels which are presented in the next section refer only to 0.5% proof stress values.

3.4 Final Ageing Treatment

In order to explore the influence of the final ageing temperature on the strength increment due to strain-ageing, the prestrained specimens were aged in an air circulating furnace at 90°, 135°, 180° or 220°C for periods of 30 minutes.

3.5 Heat-Treatments Made to Vary the Strain Ageing Potential

In the second stage of the work, the steels were re-heat treated in order to allow an investigation of the effects of variations in the overageing treatment which is applied after rapid cooling from the annealing temperature.

The heat-treatment consisted of re-heating the dual-phase sheets to 740°C for 15 minutes in a chloride salt bath, quenching in water and tempering in a nitrate-nitrite bath at 200°, 300° or $400^{\circ}C$ for 3 minutes. To allow comparison with a conventional aluminium killed drawing quality steel which had been rapidly cooled from a sub-critical annealing temperature, sheets of the CR1 steel were heated to $695^{\circ}C$ for 15 minutes in a salt bath and quenched in water and then tempered as above.

Some representative microstructures of the re-heat treated 60 and 80 steel are shown in Figs 9a, b and c. Fig 9a shows steel 80 as-quenched from $740^{\circ}C$. Figs 9b and c show the quenched 80 steel after pre-ageing for 3 minutes at 150° and $400^{\circ}C$. The martensite is relatively light-etched in the as-quenched and aged at $150^{\circ}C$ conditions but after ageing at $150^{\circ}C$ the ferrite grains show clear evidence of carbide precipitation on a very fine scale. This was seen even more clearly in the re-heat treated 60 steel after ageing at $150^{\circ}C$ (Fig d). Fig c shows that ageing at $400^{\circ}C$ for 3 minutes was sufficient to give the martensite areas dark-etching properties caused by general carbide precipitation during decomposition of the martensite, but the carbide precipitation within the ferrite grains was on a relatively coarse scale. This was also seen in the as-received 60 and 80 steels after ageing at $400^{\circ}C$ (Figs 5a and b). The volume fractions of martensite obtained after quenching from $740^{\circ}C$ were about 0.19 in the 80 steel and 0.11 in the 60 steel.

The heat-treated steels were stretched to an extension of 4% uniaxially as before, either in the as quenched condition or after overageing at one of the temperatures

Figs 9a - 9d

Microstructures of the re-heat treated dual-phase steels 60 and 80, Mag X 2000.

9a Steel 80 quenched from 740°C.

9b Steel 80 quenched from 740°C, overaged at 150°C.

9c Steel 80 quenched from 740°C, overaged at 400°C.

9d Steel 60 quenched from 740°C, overaged at 150°C.

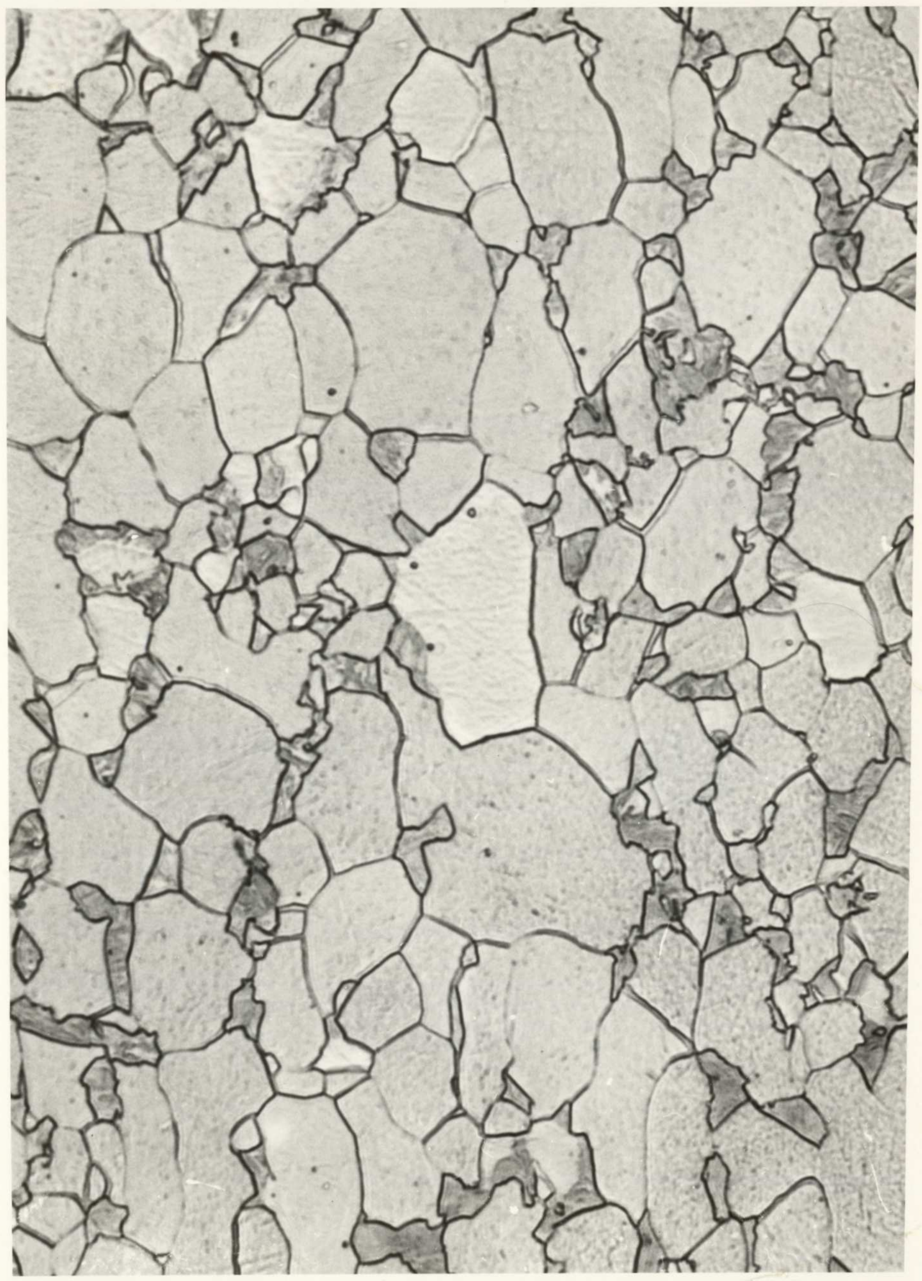


Fig 9a

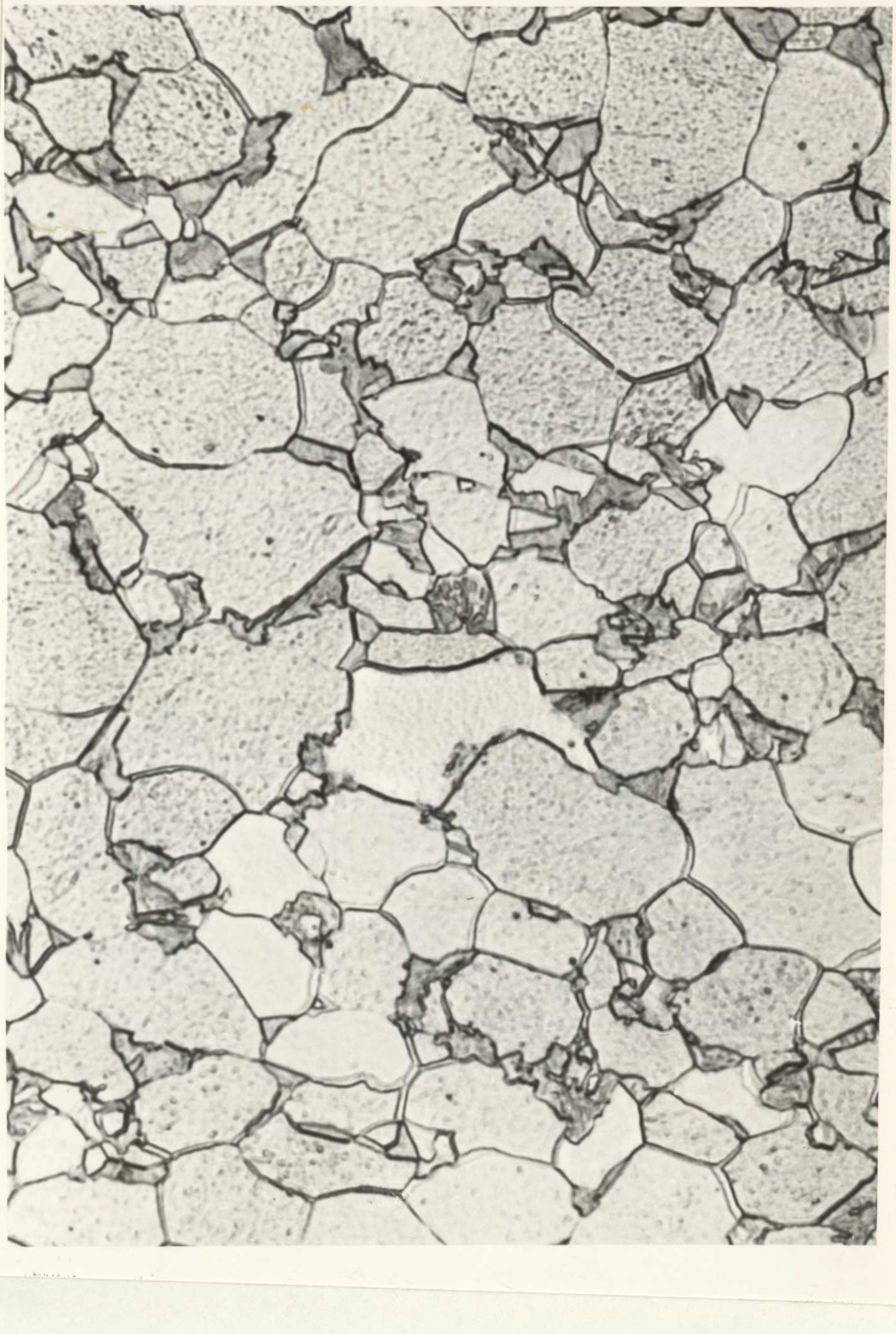


Fig 9b

Fig 9c

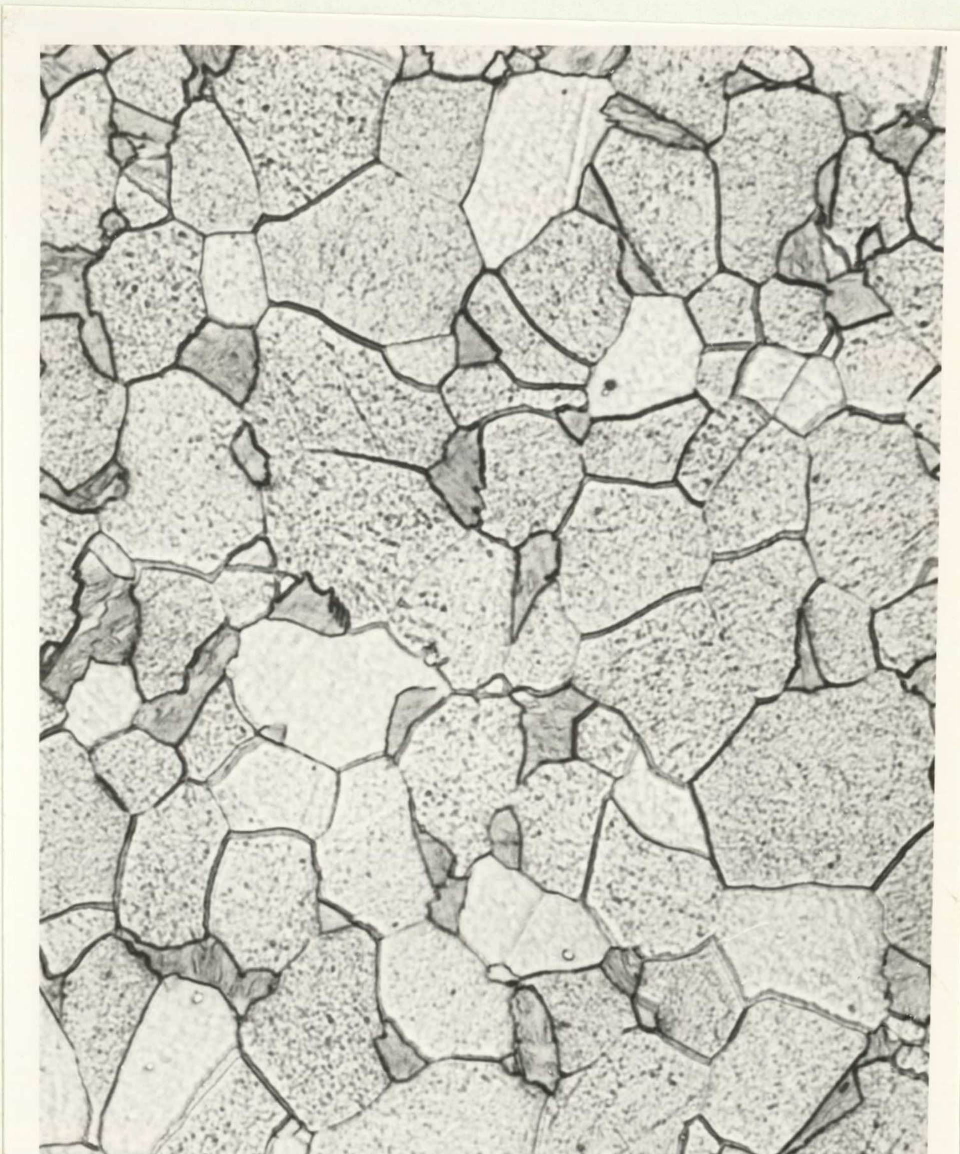


Fig 9d

mentioned above. In all cases the stretched sheets were finally tensile tested at 0° and 90° to the direction of major principal strain after ageing at 20° for several hours or at 90°, 135°, 180° or $220^\circ C$ in the air circulating furnace for 30 minutes.

3.6 Temper Rolling

Although a high ageing potential is advantageous after pressing it is important that the sheet should not age significantly during storage at room temperature before pressing. In order to investigate the influence of variations in the overageing temperature on the stability of properties at room temperature after temper rolling, sheets of two dual-phase steels and CR1 were quenched from 740° and $695^\circ C$ respectively. The sheets were then temper rolled $\sim 1\%$ in the as-quenched condition and also after pre-ageing for 3 minutes at 200°, 300° and $400^\circ C$. Finally tensile pieces were cut from the rolled sheets and aged at $47^\circ C$ for different periods of days. This last temperature was chosen to provide an acceleration of the kind of ageing which can be expected to occur during prolonged periods of storage at room temperature.

3.7 Dent Resistance Measurements

Dent resistance of the sheets were measured from the point of view of resistance against localised static loads. The effect of strain ageing on dent resistance was studied by means of a denting test applied to panels stretched by uniaxial tension, plane strain, hydraulic bulging and plane equibiaxial stretching. After stretching some of the panels were aged at 180° or 220° for 30 minutes and subjected to denting tests using the apparatus shown in Fig 10. The panels were clamped on the surface plate of the apparatus and the denting was performed by impressing a 50mm diameter steel ball into the surface of the panel. The permanent dent depth of the panel after releasing the load was measured by means of a micrometer dial gauge. Dent resistance tests were carried out in the 50 ton Instron machine. These tests were made on flat panels which had been prestrained in uniaxial tension, plane strain or equibiaxial stretching and also on curved panels equibiaxially stretched in hydraulic bulging.

In all cases the tests were made on sheets prestrained in the as-received condition but in some cases tests were also made on sheets which had been reheat-treated to give different strain ageing potentials. The tests made on flat panels will be described first.

3.7.1 Flat Panels

In the first stage of the work flat sheets of dual-phase 60, 80 and CR1 were prestrained in the as-received conditions to give a strain of 4% in the direction of maximum extension in uniaxial tension or in plane strain. After prestraining the steels were cut into panels 6 inches square, dent tested in the as-stretched condition and after BH for a period of half an hour at temperatures

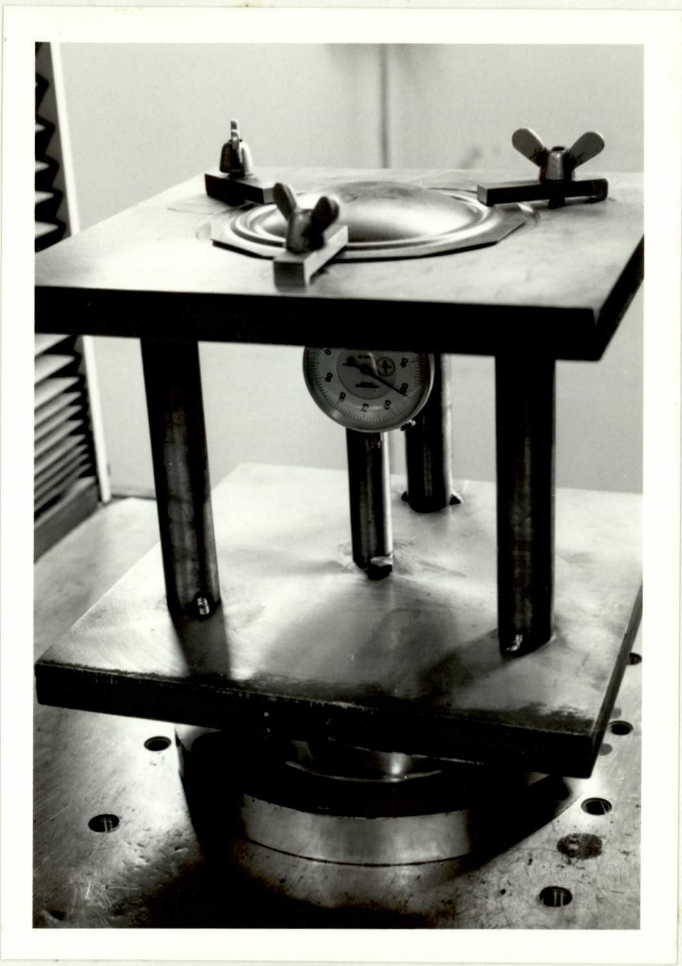


Fig 1O. Apparatus used in dent resistance tests.

of 180° or 220°C. It should be noted that the panels for the as-stretched and aged at 180°C conditions were prepared from the same prestrained sheets. A few tests were made on as=received sheets which were not prestrained before denting.

In the second group of tests, the sheets were reheat-treated (water quenched from 740°C for the dual-phase steels and from 695°C for CR1) and then pre-aged at 100-150-200-300 or 400°C for 3 minutes. The sheets could not be pre-aged at 100°, 150°C in the salt bath therefore the pre-ageing at 100, 150, 200°C was carried out in an oil bath. These reheat-treated sheets were stretched 4% in uniaxial tension, then dent tested in the "as-stretched" condition and after final ageing treatments of 180° or 220°C for a period of half an hour.

In-plane balanced biaxial stretching was carried out in the Mays press. The sheets were stretched to give a thickness strain of 4% or 8%. The stretched sheets were then dent tested after final ageing treatments at 180°C or 220°C.

3.7.2 Curved Panels

In the second phase of the work the steels were stretched biaxially in a Mand hydraulic bulge tester in the "as-received" and reheat-treated conditions. The overall diameter of the stretched area was about 11-12cms. The sheets in "as-received" conditions were stretched to give

4% and 8% thickness strain at the pole of the bulge. Curved panels were then dent tested in the as-stretched condition and after the final ageing treatments of 180°C and 220°C. The radius of curvature at the pole was approximately 180mm after 4% strain and 105mm after 8% strain. In the second part the steels were reheat-treated as described previously with a full range of preageing treatments in the range of 100°-400°C and then pre-strained in the bulger to 4% thickness strain and dent tested as-stretched and after the final ageing treatments.

CHAPTER 4

EXPERIMENTAL RESULTS

4.1 Investigation of As-Received Sheets Using Tensile Tests

4.1.1 Uniaxial Pre-Straining

Figures 11 to 13 show the results of tensile tests made on the two dual-phase steels and aluminium stabilised steel of CR1 quality after prestraining 4% in uniaxial tension in the as-received condition. The tests were made in the as-stretched condition and after final ageing treatments of 90°, 135°, 180°, 220°C for 30 minutes. The figures show 0.5% proof stress and UTS values as a function of the final ageing temperatures. After prestraining in uniaxial tension the 0.5% proof stress of the dual-phase was directional for tests at 0° and 90° to the direction of maximum extension in the prestrain.

The influence of elevated temperature ageing is evident with dual-phase steels (Figs 11, 12). The results show that there were useful increases in both 0.5% proof stress and UTS values after ageing above 135°C, but for the transverse (90°) tests an ageing temperature above 180°C was required to eliminate the directionality in the 0.5% proof stress. The effect of ageing on the 0.5% proof stress values were rather different from those on the UTS values. The most important difference is that the increase in the proof stress values are much greater than those in the UTS particularly after ageing at 180°C and 220^{o}C.

The corresponding plot for the CR1 steel is shown in Fig 13. No significant change in the values of 0.5% proof stress and UTS were observed after ageing treatments up to 220^{o}C, but the 0.2% proof stress (Fig 14) indicated a small rise in ageing.

4.1.2 Pre-Straining in Plane Strain

The effect of ageing temperatures after 4% pre-strain in plane strain on the 0.5% proof stress and UTS values of the dual-phase steels is shown in Figs 15 and 16. In these tests the results are very similar to those with pre-straining in uniaxial tension. As may be seen, directionality in proof stress values was reduced on ageing above 135^{o}C. There was a significant reduction in directionality after ageing at 180^{o}C but ageing at 220^{o}C was required to develop a large reduction in directionality of the 0.5% proof stress values.

Fig 17 shows the corresponding results obtained from the as received steel CR1. As with uniaxial prestraining the results show little difference in the flow stress values measured at 0^{o} and 90^{o}C to the direction of maximum extension in stretching and very little change after ageing.

4.1.3 Pre-Straining in Equibiaxial Stretching

Tensile test results obtained from the dual-phase steels after stretching equibiaxially to 4% thickness strain followed by ageing treatments in the 20° to 220°C range for 30 minutes are shown in Figs 18 and 19. In the equibiaxially stretched conditions there was no significant variation in flow strength with direction in the sheet plane. However, considerable increases in the 0.5% proof stress and UTS values were achieved on ageing at 180°C and ageing at 220°C gave the highest strengths. However, the 0.5% proof stress values remained considerably lower than the UTS values after all the ageing treatments. A similar plot for biaxially stretched CR1 steel is shown in Fig 20. No significant change was observed in the values of 0.5% proof stress and UTS after ageing treatments up to 220°C.

4.1.4 Pre-Straining in Rolling

Figs 21 and 22 relate to results obtained from dual-phase steels after prestraining 5% in rolling and then ageing at the selected temperatures. Directionality of the 0.5% proof stress was less than those observed after uniaxial and plane strain stretching but the reduction in directionality on ageing was rather small.

4.2 Effect of Variation in Pre-Ageing Temperature

This section relates to the investigation of the influence of variations in the pre-ageing temperature applied before pre-straining. The dual-phase steels were re-heat treated by water quenching from 740°C and then overageing for 3 minutes at 200°, 300°or 400°C. The CR1 steel was water quenched from 695°C before ageing at the above temperatures. Figs 23 to 34 show the tensile test results obtained after prestraining 4% in uniaxial tension, for the as-stretched condition and after final ageing treatments of 180°C and 220°C for 30 minutes. Figs 23, 24, 25 and 26 show the 0.5% proof stress values of the dual-phase steels 60 and 80, which were pre-strained in the as-quenched condition and after the three pre-ageing treatments, and finally bake-hardened at 180°C or 220°C. As may be seen, with the as-quenched condition, bake-hardening of 180°C was sufficient to eliminate most of the directionality in the proof stress values.

With pre-ageing at 300°C or 400°C significant directionality remained even with bake-hardening at 220°C. With pre-ageing at 200°C much of the directionality was eliminated in bake-hardening of 180°C and bake-hardening of 220°C eliminated the directionality almost completely. The corresponding relationship for the CR1 steel are given in Figs 27 and 28. The changes in strength of the as-quenched CR1 on ageing in the 90°C to 220°C range were less than in the dual-phase steels. The UTS values were not significantly directional and they showed only a modest improvement with the bake-hardening treatments. Directionality of the 0.5% proof stress value was relatively small and it was practically eliminated after bake-hardening at 220°C.

4.3 Re-heat treated Conditions given Low Temperature Ageing after Temper-Rolling

This section shows the tensile results of dual-phase steels and CR1 steel which have been re-heat treated as described in the previous section and then temper-rolled about 1% and finally aged at 47°C for a prolonged period of time. Equivalent ageing times at 20°C were calculated using Hundy's equation[32]:

$$\text{Log}\left(\frac{t_r}{t}\right) = 4400\left[\left(\frac{1}{T_r}\right) - \left(\frac{1}{T}\right)\right] = \text{Log}\left(\frac{T}{T_r}\right)$$

where t_r and t are the times to give the same degrees of strain ageing at room temperature (T_r) and a higher temperature (T) respectively.

Results with the three steels are presented in the form of stress and total elongation values as a function of Log time at 47°C in Figs 35 to 37. According to Hundy's equation ageing at 47°C is approximately 16.9 times more rapid than at 20°C. From the changes observed in samples which had been temper-rolled in the as-quenched condition, it appears that considerable strain-ageing occurred during specimen preparation in this condition. The initial strengths of the samples were high and the effect of prolonged ageing at 47°C was to reduce this strength and to cause some improvement in the total elongation from initially low values. However, preageing at temperatures in the range 200°C-400°C stabilised the steels and the changes in flow strengths and elongation values detected during ageing at 47°C for times up to 45 days were extremely small. It is notable that increases in the preageing temperatures in the 200°C to 400°C range caused significant increases in the total elongation values, particularly for steel 60 and CR1.

4.4 Dent Resistance Tests

4.4.1 Dent Resistance Results in the As-Received Condition

4.4.1.1 Flat Panels

Figs 38 to 40 show the dent resistance results of the two dual-phase steels and CR1 in the as-received condition. They are compared with results obtained with the same steel after prestraining 4% in uniaxial tension. With the dual-phase steels there is no significant change between the panels, dent tested after 4% prestrain and the one which is not prestrained. With the CR1 steel (Fig 40) there is a rise in dent resistance with the prestrained panel.

Fig 41 relates to flat panels of the dual-phase steel 60 which were prestrained 4% in uniaxial tension before dent testing in the 'unaged' and BH conditions. As may be seen, the resistance to dents of 0.2mm depth is quite low after 4% prestrain and this value is greatly improved by bake-hardening. The corresponding results for dual-phase 80 are shown in Fig 42. The behaviour of steel 80 was rather similar to steel 60. Fig 43 shows results for the steel CR1. There is some improvement in dent resistance on BH but it is relatively small. The dent resistance relationships of panels prestrained 4% in plane strain (Figs 44, 45 and 46) were essentially similar to those with uniaxial stretching.

The results for flat panels of the as-received steels which have been stretched equibiaxially to 4% and 8% thickness strain prior to denting in the as-stretched and BH conditions are shown in Figs 47 to 52. Fig 47 relates to dual-phase steel 60. It was observed that the dent resistance of low dent depths was quite low in the as-stretched panels and this was almost doubled after the BH treatment of 180°C. There was only a very small improvement in increasing the BH temperature from 180°C to 220°C. Fig 48 shows a similar plot for the dual-phase 60 stretched to 8% thickness strain.

The corresponding results for the dual-phase 80 are shown in Figs 49 and 50. Apart from the higher dent resistance which is expected from steel 80 the results are similar to those for steel 60. Figs 51 and 52 show the dent resistance results obtained with CR1 steel, stretched equibiaxially to 4% and 8% thickness strain. As after uniaxial stretching, the panels showed a small improvement in dent resistance after the BH treatments. The panel with the higher prestrain showed a larger resistance to denting.

4.4.1.2 Curved Panels

The dent resistance results for curved panels, which were prestrained equibiaxially in a hydraulic bulge tester to give 4% and 8% polar thickness strains, are shown in Figs 53 to 58. The radius of curvature at the pole of these bulged test pieces were 180mm and 105mm respectively. Dent resistance was increased substantially with the panels which had the higher prestrain and smaller radius of curvature. It is shown that raising the BH temperature from 180°C to 220°C increased the dent resistance substantially and it is also shown that the dent resistances of the curved panels were superior to those of the flat panels. The results for steel CR1 are shown in Figs 57 and 58. It was observed that the BH effect on the CR1 was relatively small.

4.4.2 Examination of the Dent Resistance of Re-Heat Treated Panels

4.4.2.1 Flat Panels

The results of dent resistance tests made on re-heat treated flat panels which have been prestrained in uni-axial tension and dent tested in the as-stretched and BH conditions are shown in Figs 59 to 68.

Figs 59 to 61 relate to the results of DPL 60, pre-aged at 200°, 300° and 400°C. The results with a pre-ageing temperature of 200°C (Fig 59) are rather similar to those for the as-received uniaxially stretched panels, but with the pre-ageing at 300° and 400°C the improvement in dent resistance is rather small.

The corresponding results for DPL 80, shown in Figs 62, 63 and 64, follow rather similar trends to those for DPL 60 steel. Bake-hardening at 180°C improved the dent resistance substantially; with the BH temperature of 220°C the dent resistance showed further improvement.

The results for CR1 steel are shown in Figs 65 to 68. These include tests made on sheets which were pre-aged below 200°C. An examination of the plot obtained with pre-ageing at 150°C (Fig 65) shows that BH reduced the denting load; this is believed to be due to the overageing of the fine precipitates which are formed during pre-ageing at 150°C. With pre-ageing at 200°C dent resistance increased on ageing of 180°C and 220°C. With sheets pre-aged at 300°C and 400°C (Figs 67, 68) BH at 180°C and 220°C also increased the resistance to denting; the two temperatures having rather similar effects. It was also observed that the dent resistance obtained with heat-treated panels were higher than with the as-received box-annealed condition even with the pre-ageing temperature of 300°C and 400°C.

4.4.2.2 Curved Panels of Re-Heat Treated Sheets

Figures 69 to 86 show the results of dent resistance tests made on circular hydraulically bulged testpieces of about 180mm radius of curvature with polar thickness of $\sim$ 4%. The sheets used for these tests were reheat-treated as described earlier but preageing temperatures below 200°C were examined in this case.

The results for dual-phase 60 steels are shown in Figs 69 to 74. With the as-quenched condition and pre-ageing treatment of 100°C (Fig 70) the resistance to denting was improved by BH at 180°C. However, after BH of 220°C the dent resistance, although higher than in the as-stretched condition, was lower than with BH at 180°C. This is believed to be due to overageing during the bake-hardening treatment at 220°C. Preageing at 150°C (Fig 71) with BH of 180°C and 220°C made similar improvements in dent resistance. With a preageing temperature of 200°C (Fig 72) BH made a relatively larger increase in dent resistance. After preageing treatments of 300°C and 400°C (Figs 73 and 74) dent resistance was increased usefully but the general level of dent resistance was low in comparison with that obtained from preageing at 200°C. Results with dual-phase steel 80 are shown in figs 75 to 80. The relations are rather similar to those for dual-phase 60, except in the as-quenched condition, where BH at 180°C and 220°C had clearly similar effects in increasing the dent resistance; corresponding results for CR1 steel are shown in Figs 81 to 86. Figure 81 relates to the results of the as-quenched condition. As may be seen, the dent resistance was increased with a BH treatment at both 180°C and 220°C, but the effect of BH at 220°C was lower than with BH at 180°C. Examination of the plot obtained with a preageing temperature of 100°C (Fig 82) shows that BH at 180°C and 220°C reduced the dent resistance in this precipitation-hardened condition. Figs 83 and 84 show the results with preageing temperatures of 150°C and 200°C. Dent resistance was increased by BH treatment and the two BH temperatures had rather similar effects. The results with pre-ageing temperatures of 300°C and 400°C are shown in Figs 85 and 86. Dent resistance was increased with BH treatment of 180°C and 220°C. In this case, a higher dent resistance was obtained with BH at 220°C.

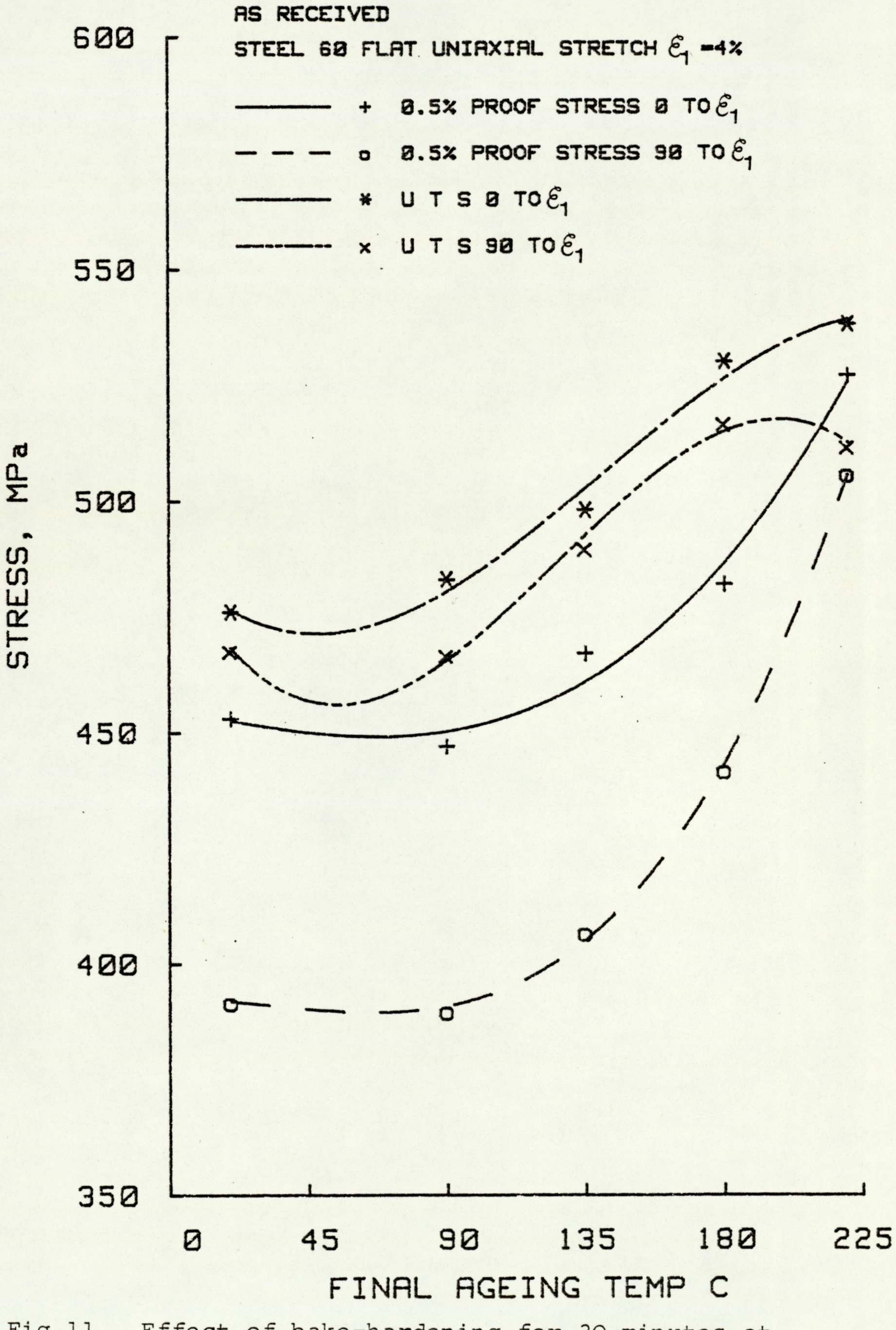


Fig 11. Effect of bake-hardening for 30 minutes at temperature ranges 90°C to 220°C on 0.5% proof stress and UTS values measured at 0° and 90° to the direction of extension in prestrain: for as-received steel 60, prestrained 4% in uniaxial tension.

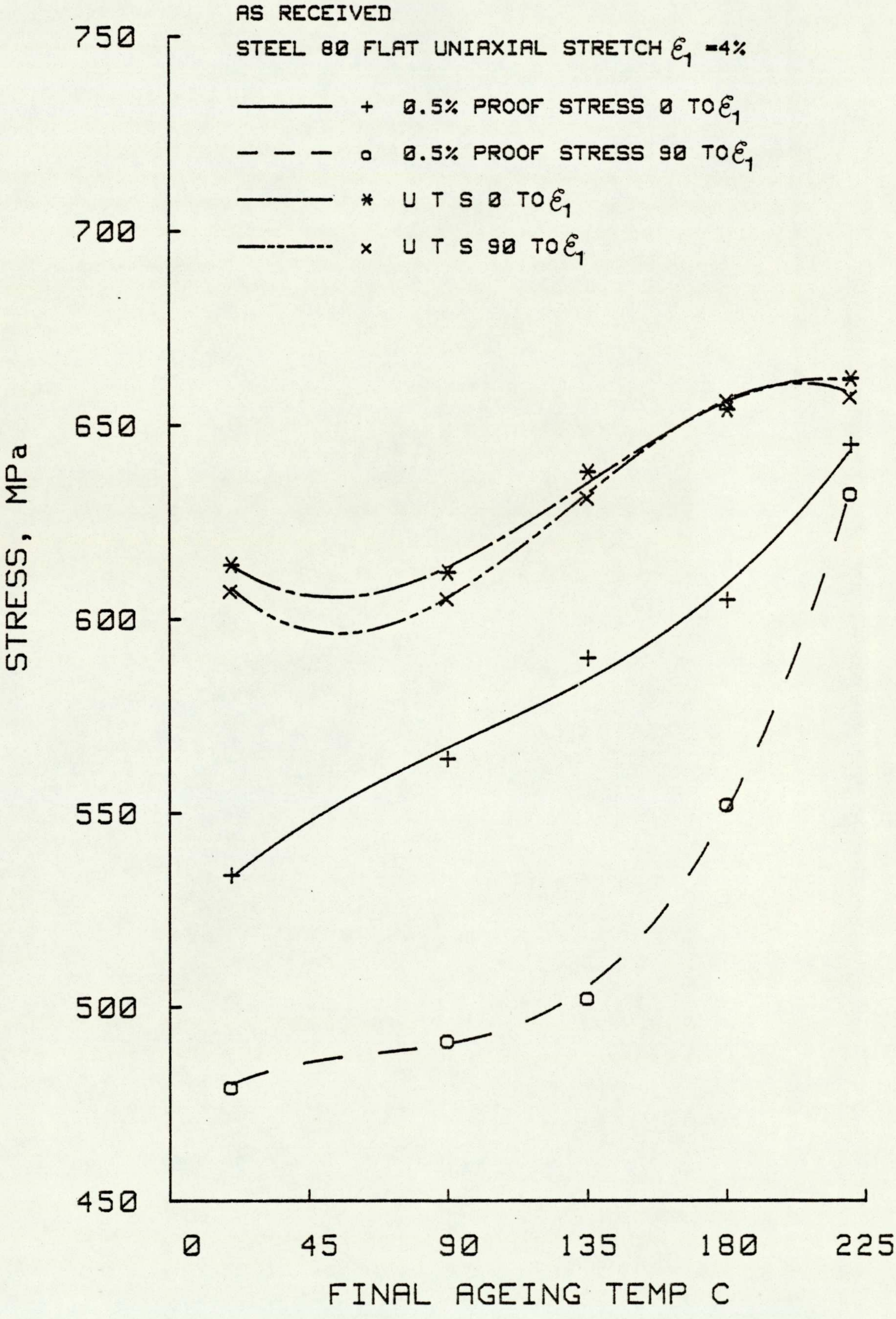


Fig 12. Effect of bake-hardening for 30 minutes at temperatures in the range 90° to 220°C on 0.5% proof stress and UTS values measured at 0° and 90° to the direction of extension in prestrain: for as-received steel 80, prestrained 4% in uniaxial tension.

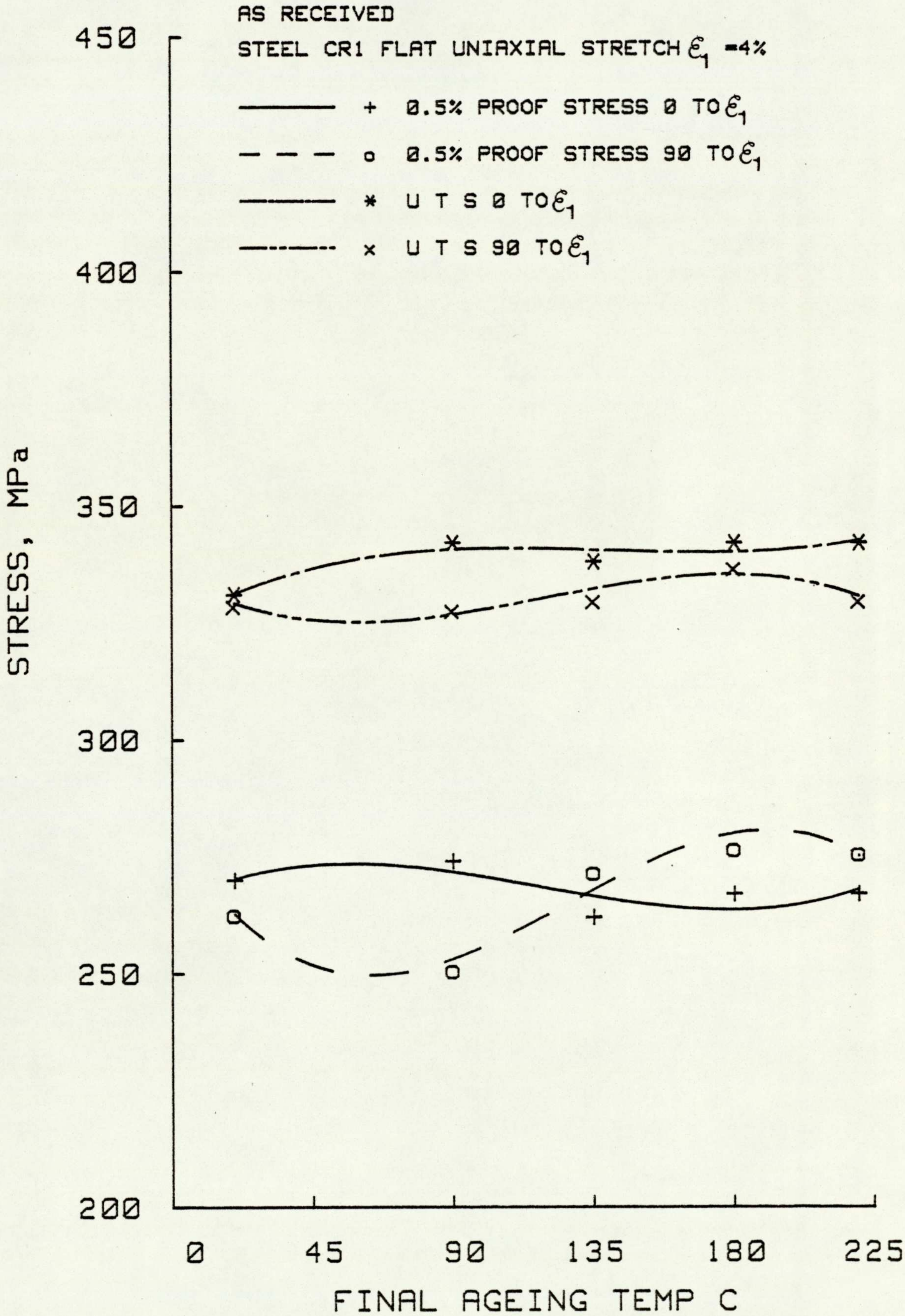


Fig 13. Effect of bake-hardening for 30 minutes at temperatures in the range 90° to 220°C on 0.5% proof stress and UTS values measured at 0° and 90° to the direction of extension in prestrain: for as-received steel CR1, prestrained 4% in uniaxial tension.

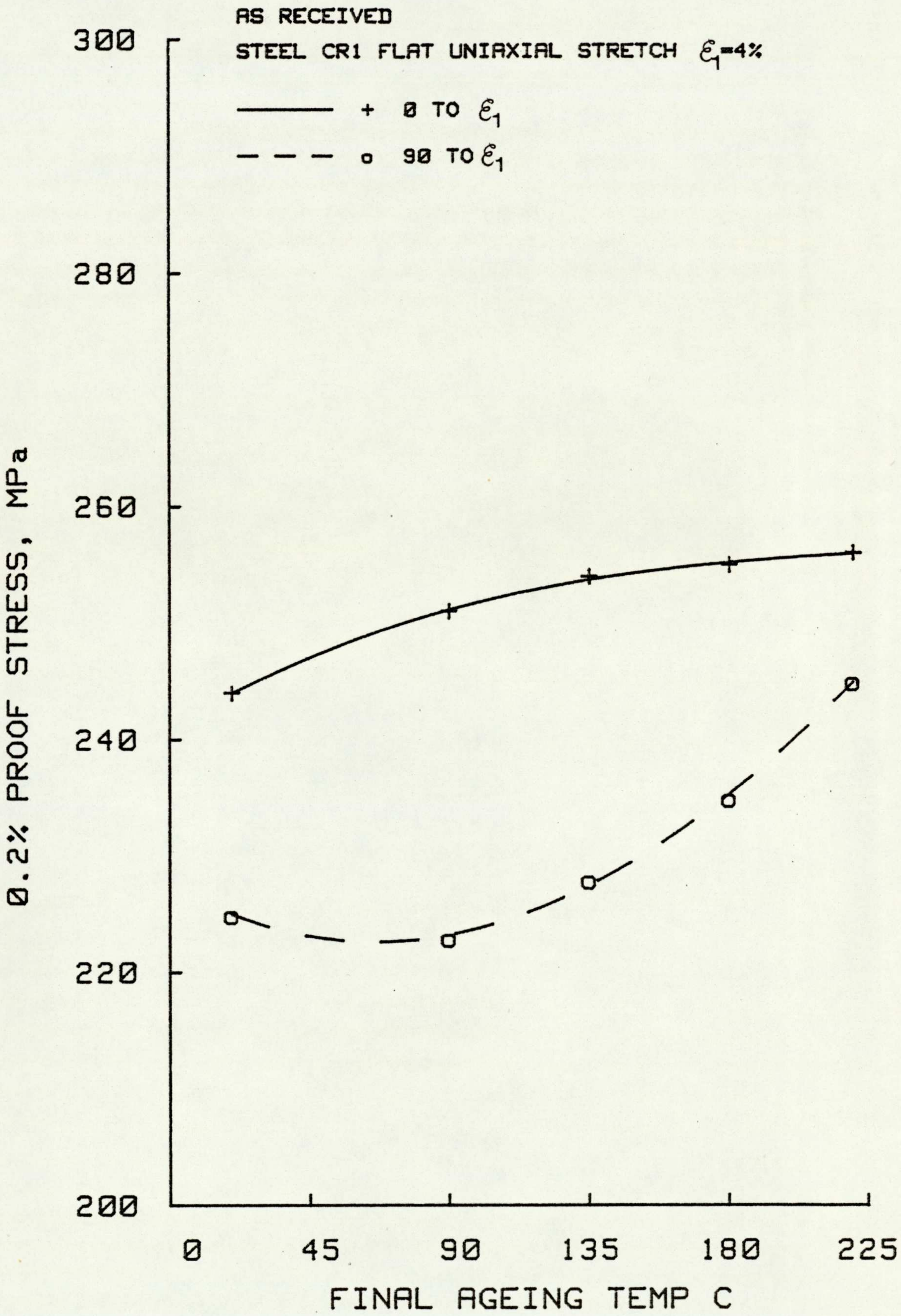


Fig 14. Effect of bake-hardening for 30 minutes at temperatures in the range 90° to 220°C on 0.2% proof stress values measured at 0° and 90° to the direction of extension in prestrain: for as-received steel CR1, prestrained 4% in uniaxial tension.

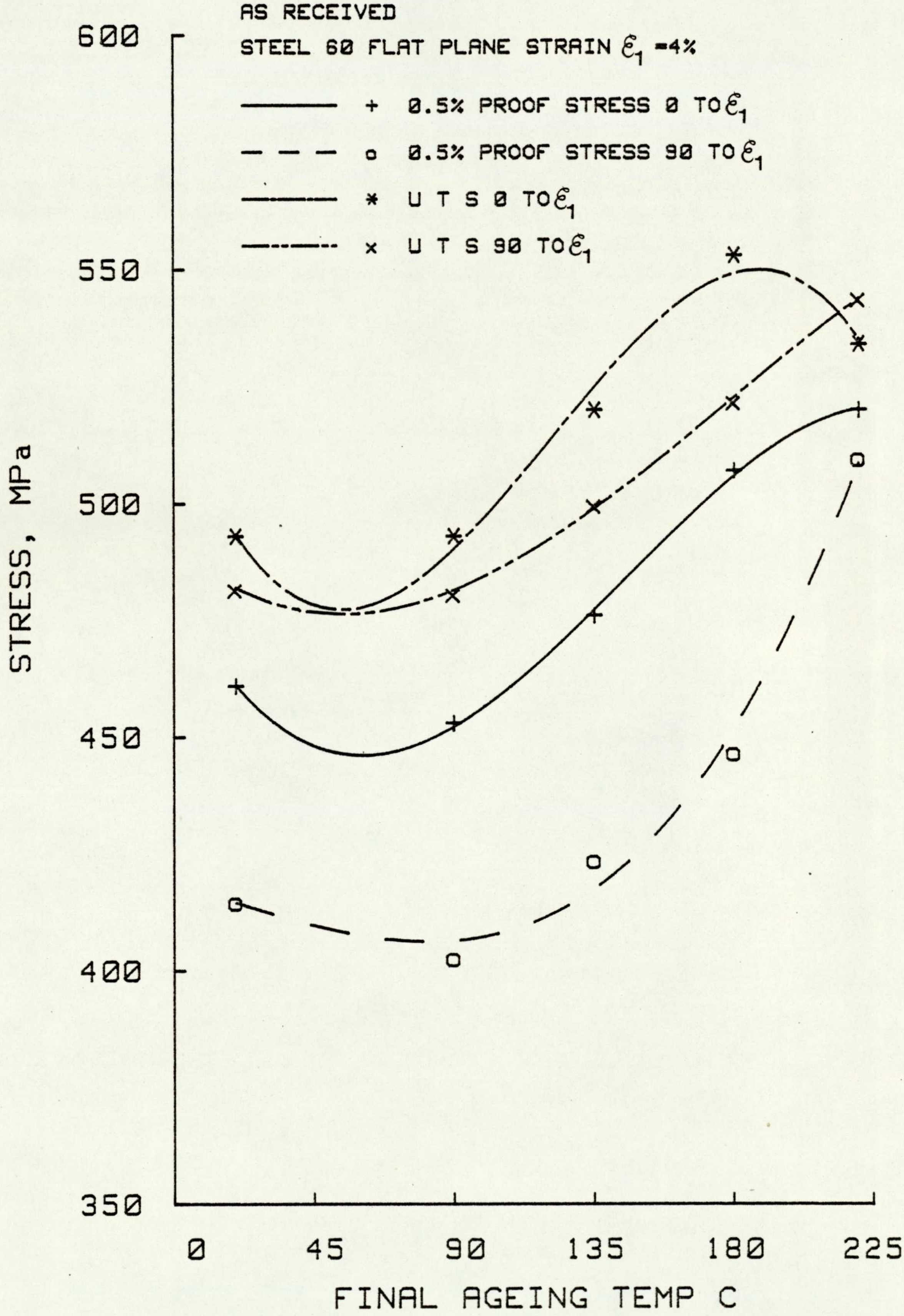


Fig 15. Effect of bake-hardening for 30 minutes at temperatures in the range 90° to 220°C on 0.5% proof stress and UTS values measured at 0° and 90° to the direction of extension in prestrain: for as-received steel 60, prestrained 4% in plane strain.

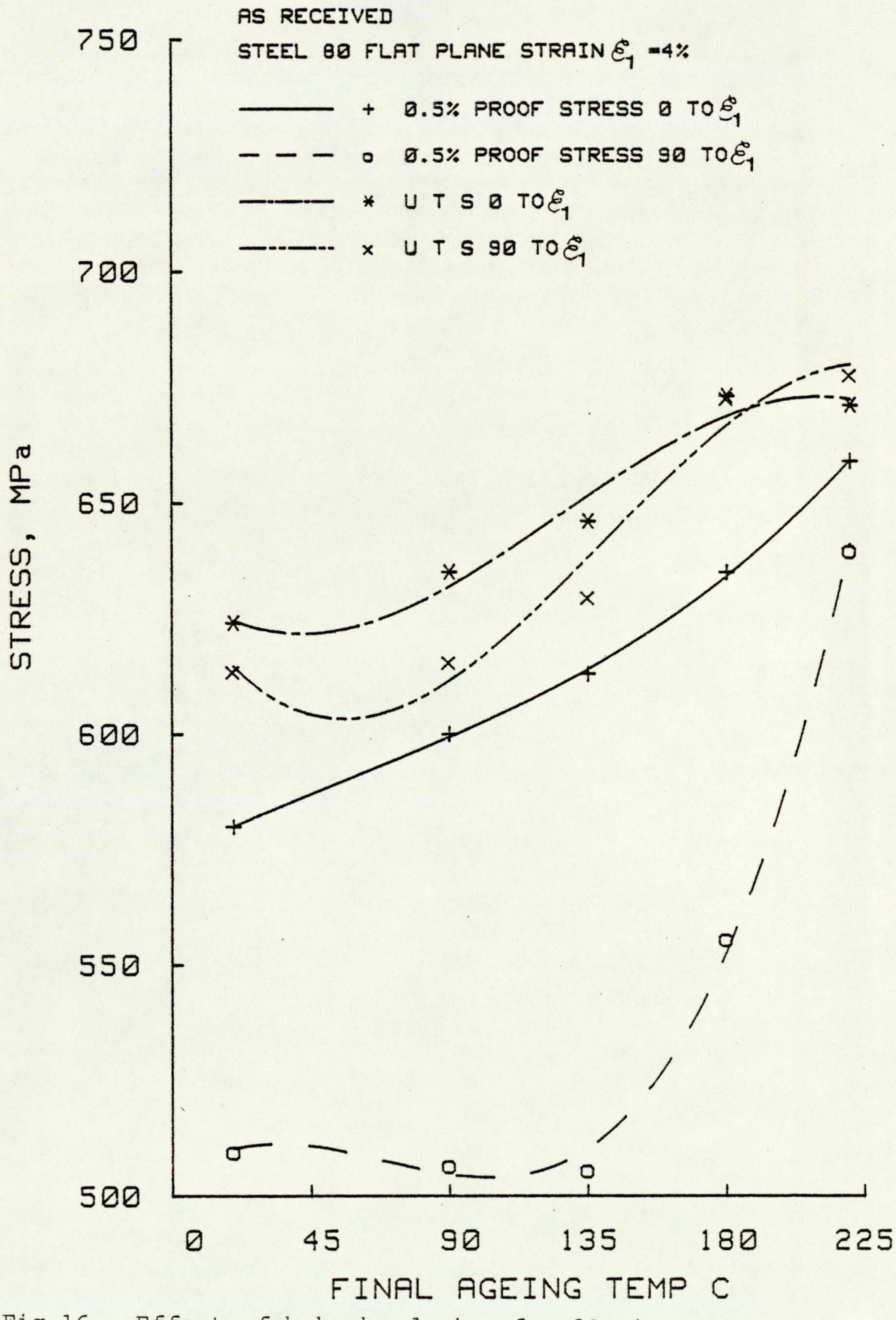


Fig 16. Effect of bake-hardening for 30 minutes at temperatures in the range 90° to 220°C on 0.5% proof stress and UTS values measured at 0° and 90° to the direction of extension in prestrain: for as-received steel 80, prestrained 4% in plane strain.

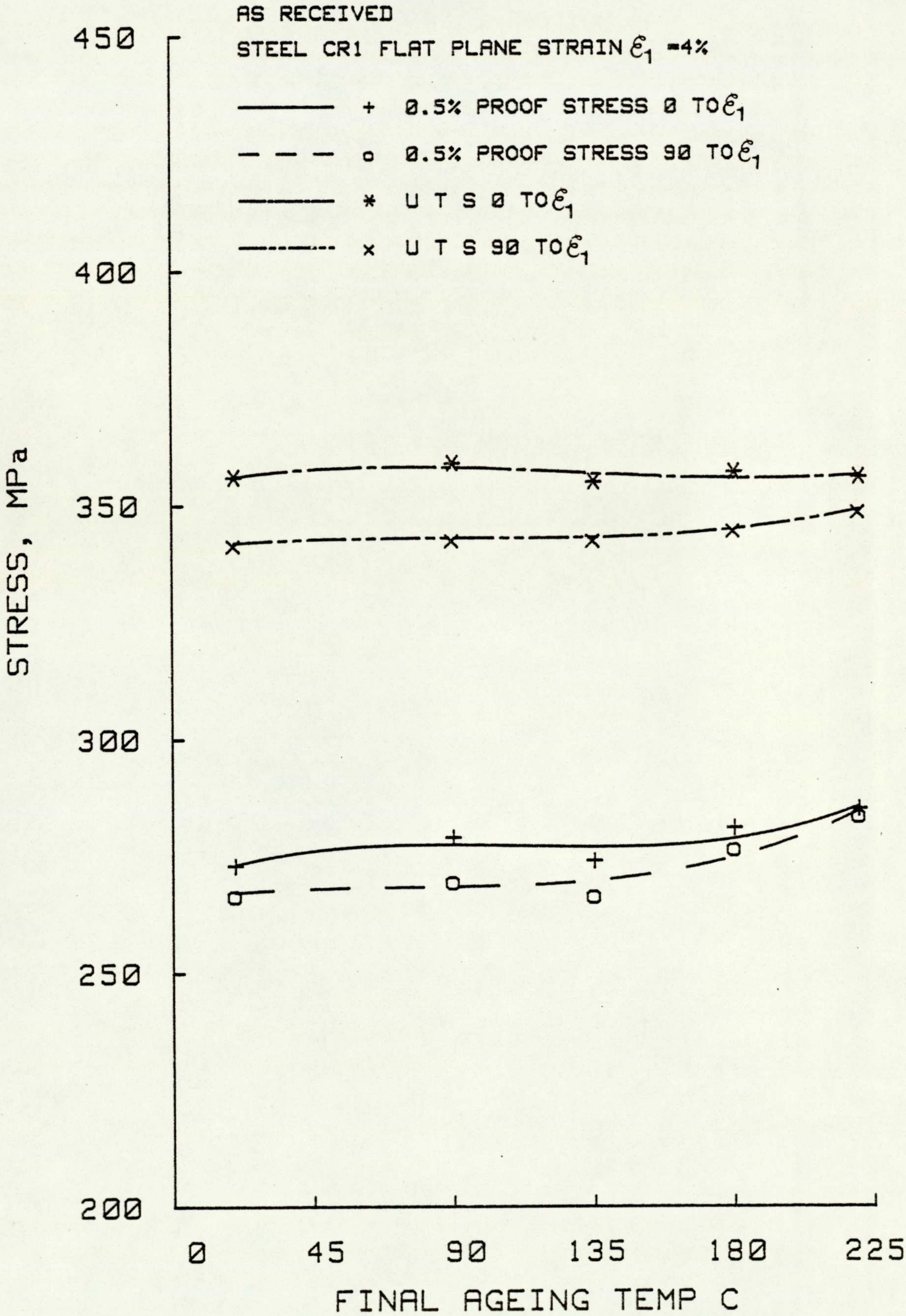


Fig 17. Effect of bake-hardening for 30 minutes at temperatures in the range 90° to 220°C on 0.5% proof stress and UTS values measured at 0° and 90° to the direction of extension in prestrain: for as-received steel CR1, prestrained 4% in plane strain.

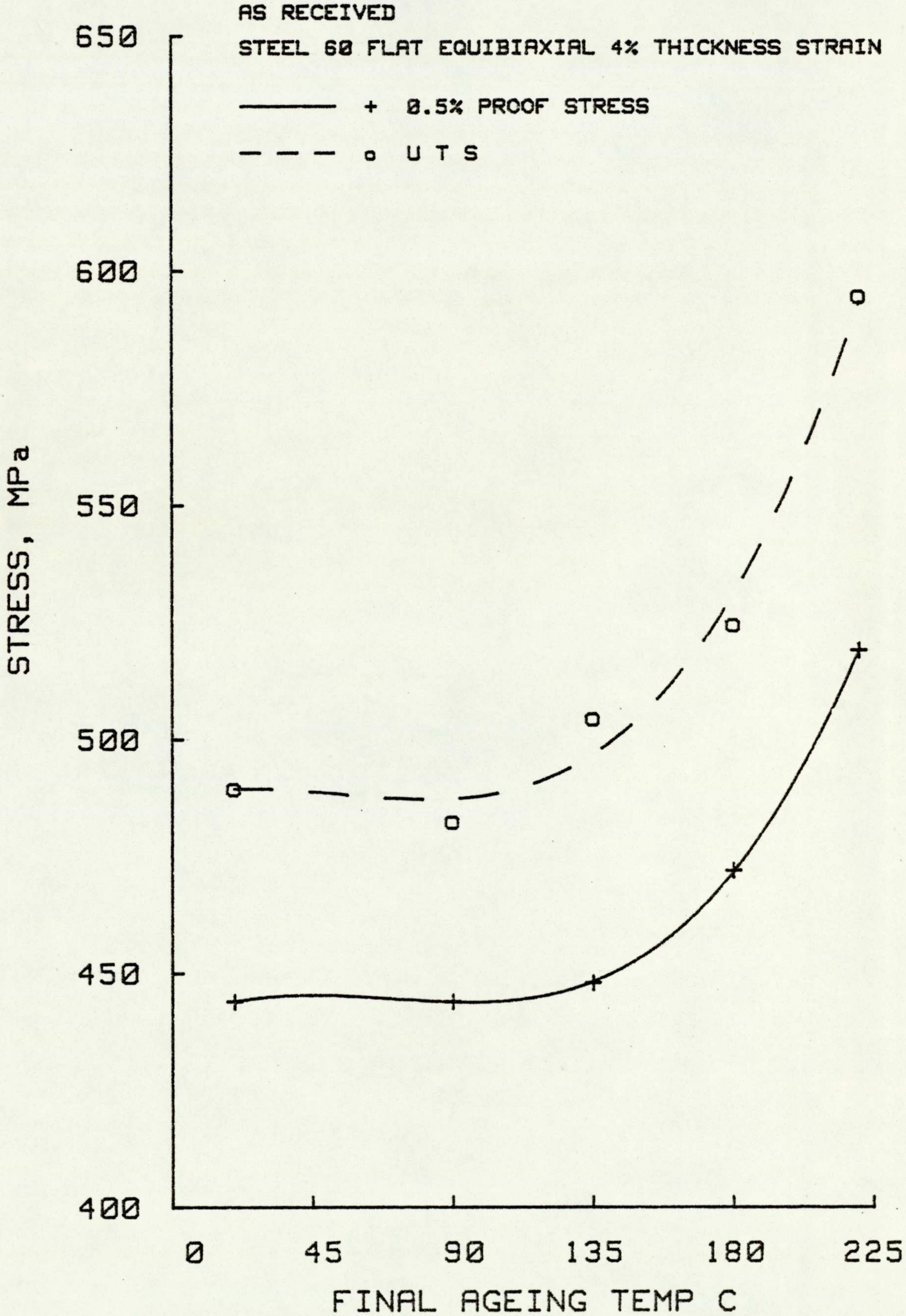


Fig 18. Effect of bake-hardening for 30 minutes at temperatures in the range 90° to 220°C on 0.5% proof stress and UTS values measured in two directions at right angles in the sheet plane. For as-received steel 60 stretched equibiaxially to a thickness strain of 4%.

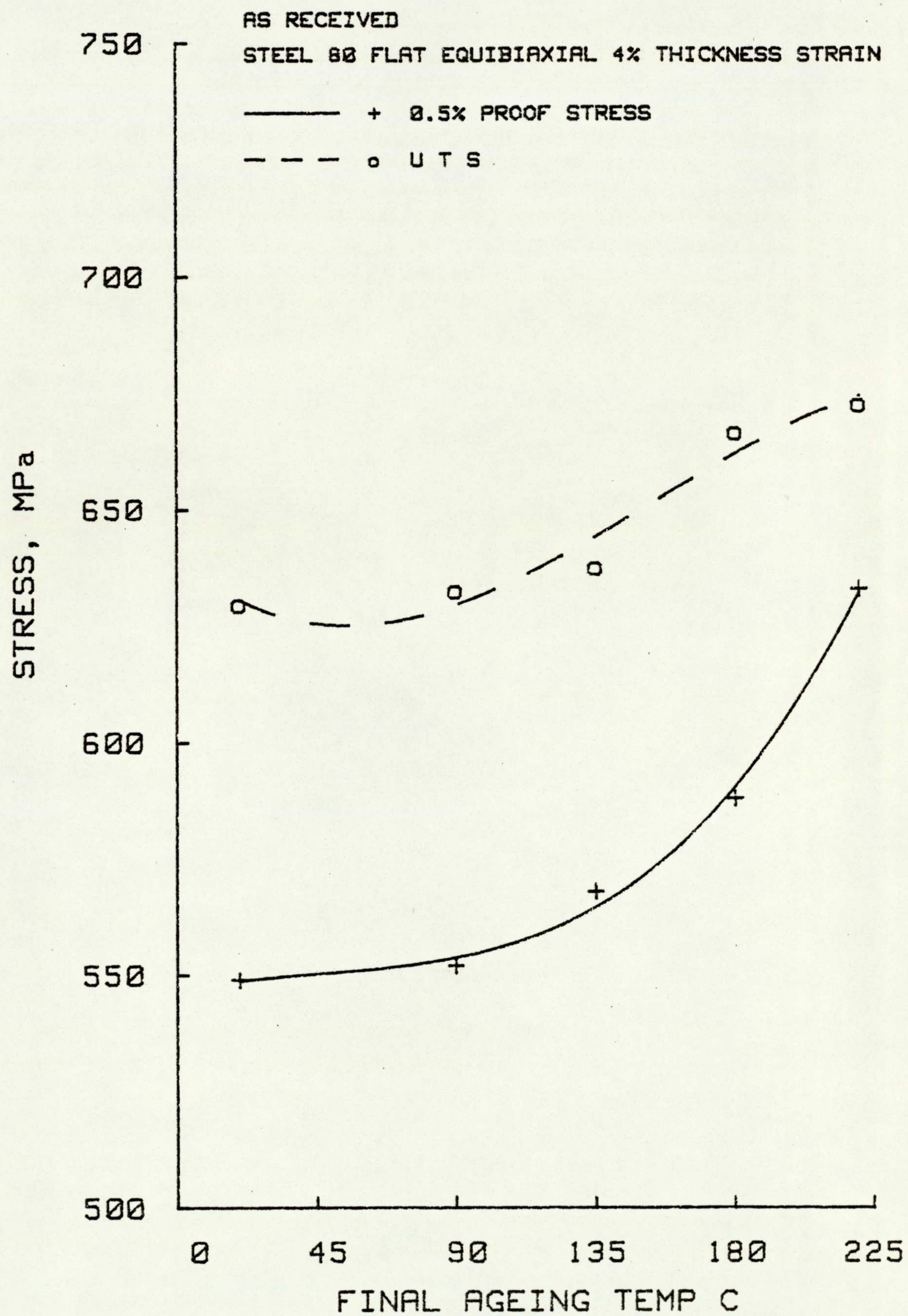


Fig 19. Effect of bake-hardening for 30 minutes at temperatures in the range 90° to 220°C on 0.5% proof stress and UTS values measured in two directions at right angles in the sheet plane. For as-received steel 80 stretched equibiaxially to a thickness strain of 4%.

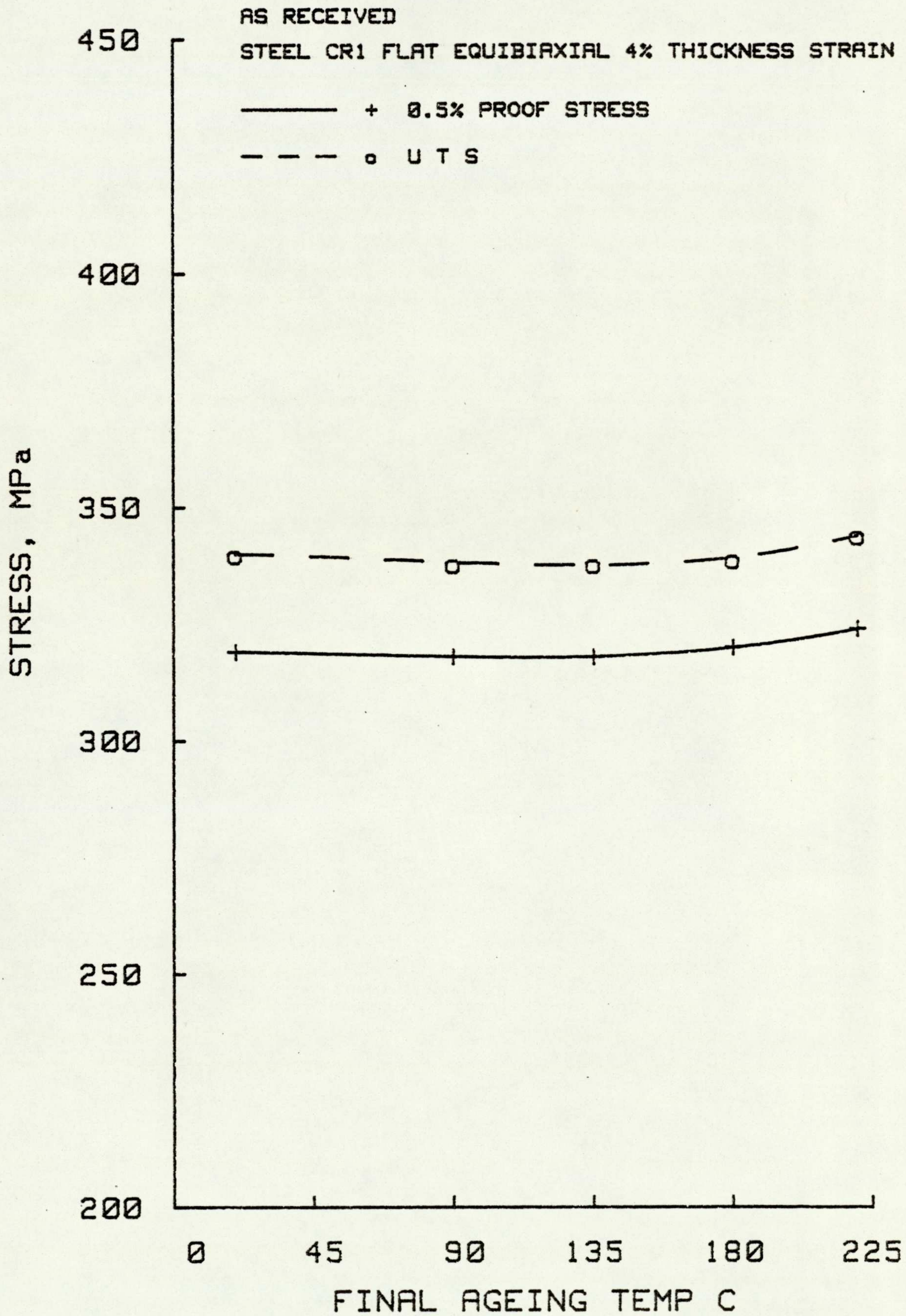


Fig 20. Effect of bake-hardening for 30 minutes at temperatures in the range 90° to 220°C on 0.5% proof stress and UTS values measured in two directions at right angles in the sheet plane. For as-received steel CR1 stretched equibiaxially to a thickness strain of 4%.

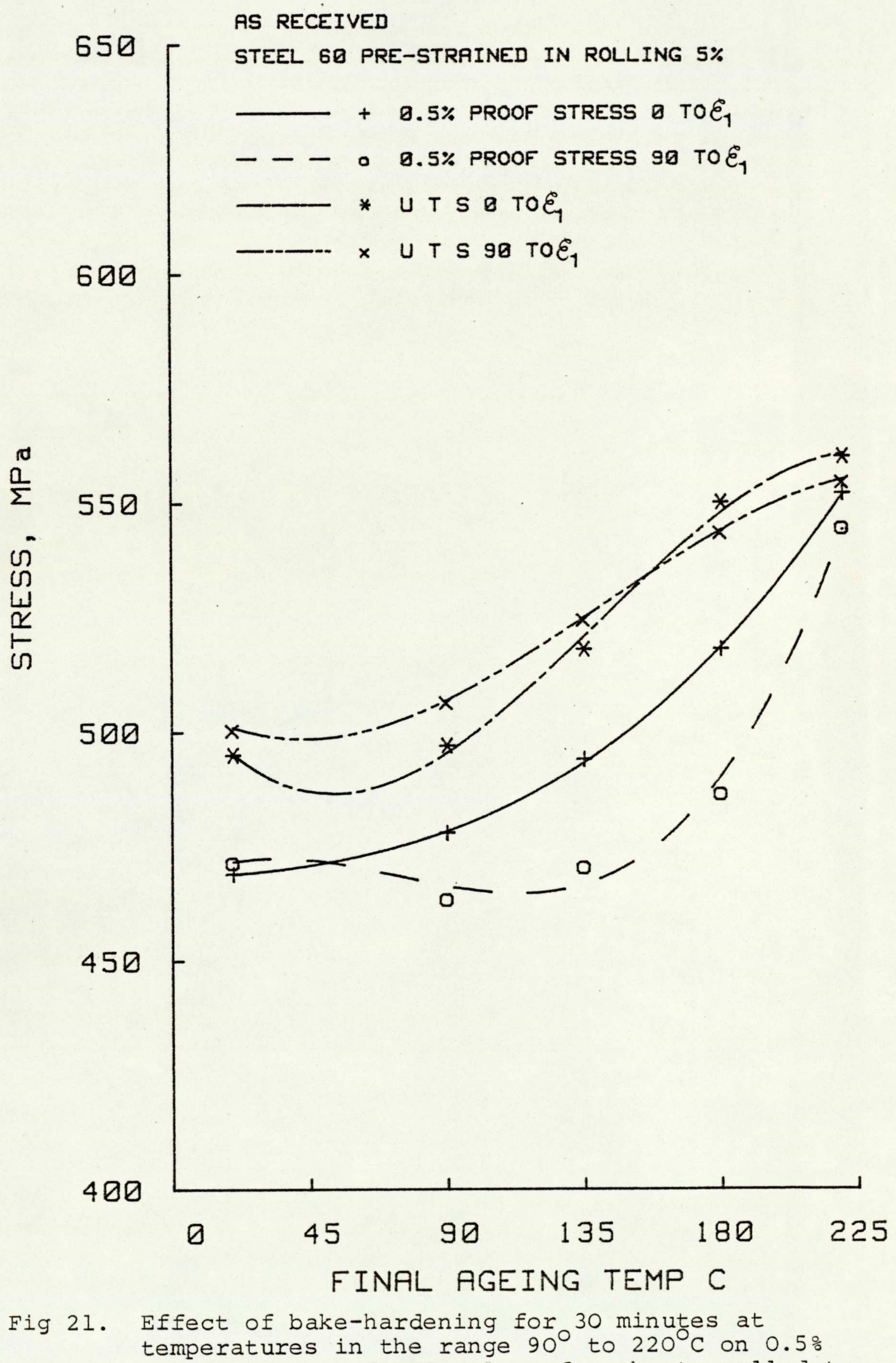


Fig 21. Effect of bake-hardening for 30 minutes at temperatures in the range 90° to 220°C on 0.5% proof stress and UTS values for sheets rolled to give 5% reduction in thickness, measured at 0° and 90° to the rolling direction: for as-received steel 60.

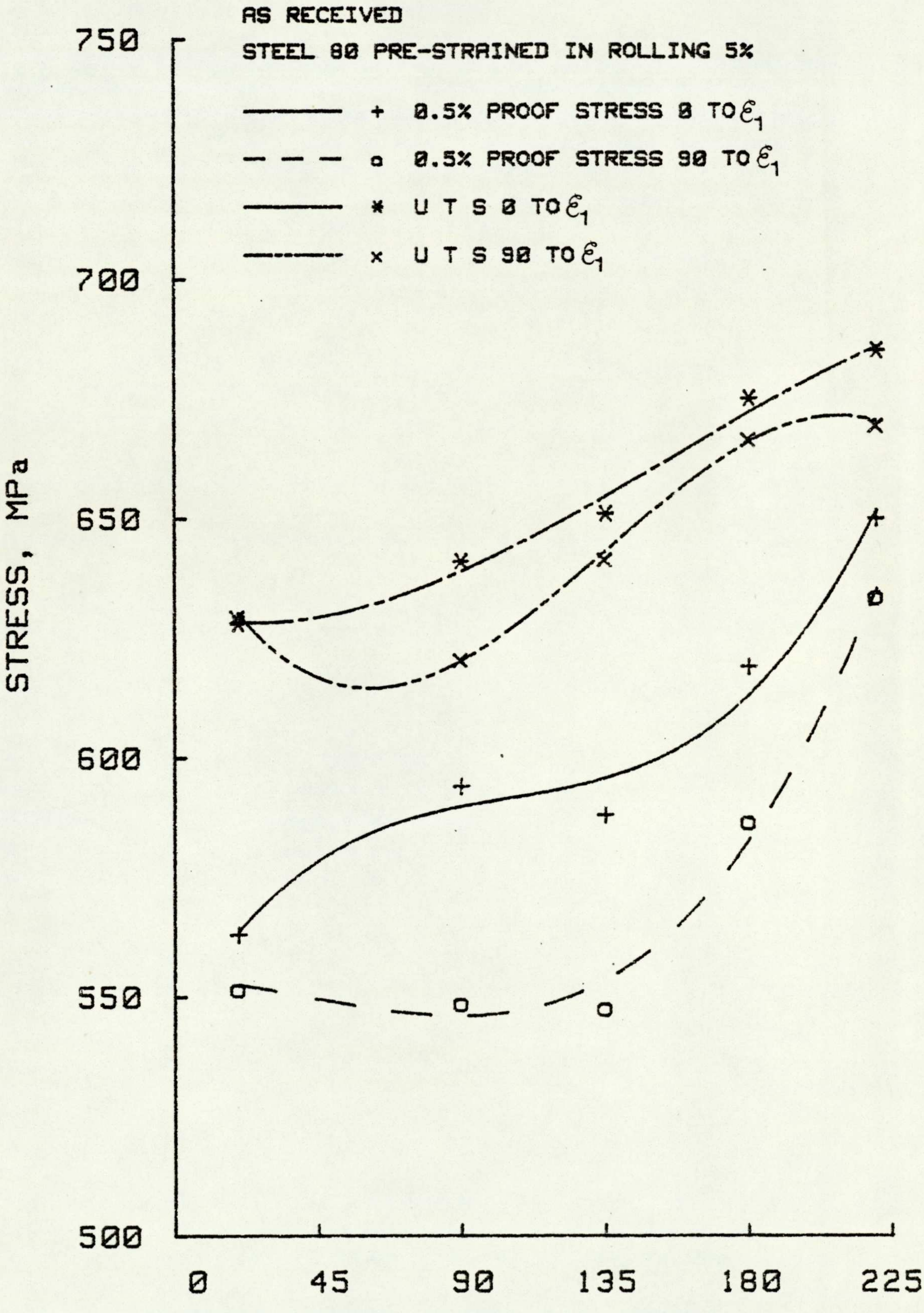


Fig 22. Effect of bake-hardening for 30 minutes at temperatures in the range 90° to 220°C on 0.5% proof stress and UTS values for sheets rolled to give 5% reduction in thickness, measured at 0° and 90° to the rolling direction: for as-received steel 80.

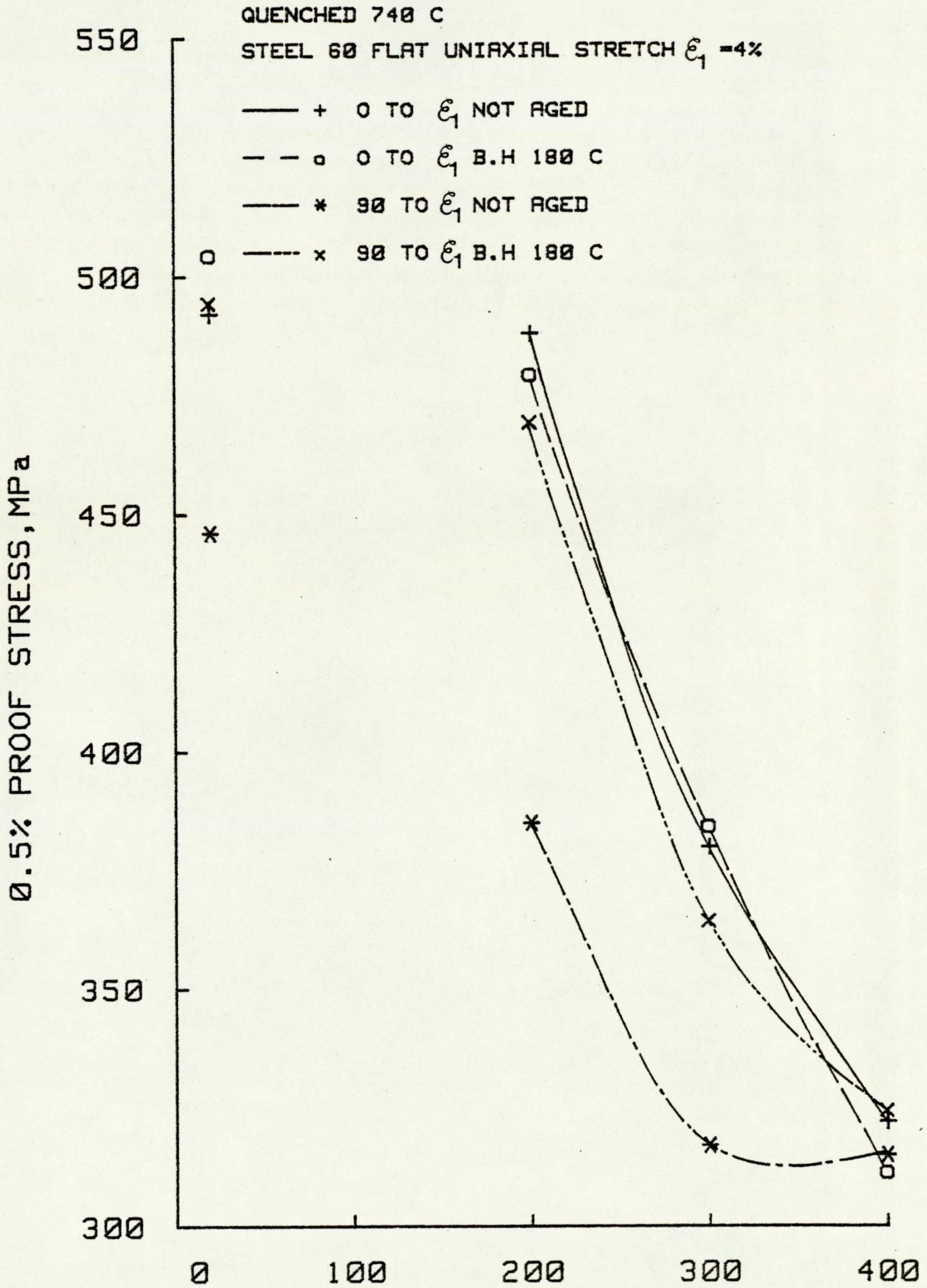


Fig 23. Effect of pre-ageing temperature on 0.5% proof stress values for re-heat treated steel 60, uniaxially prestrained 4% and tested at 0° and 90° to the direction of extension in prestraining. The tests made on samples as-prestrained and after bake-hardening at $180^{\circ}C$ for 30 minutes.

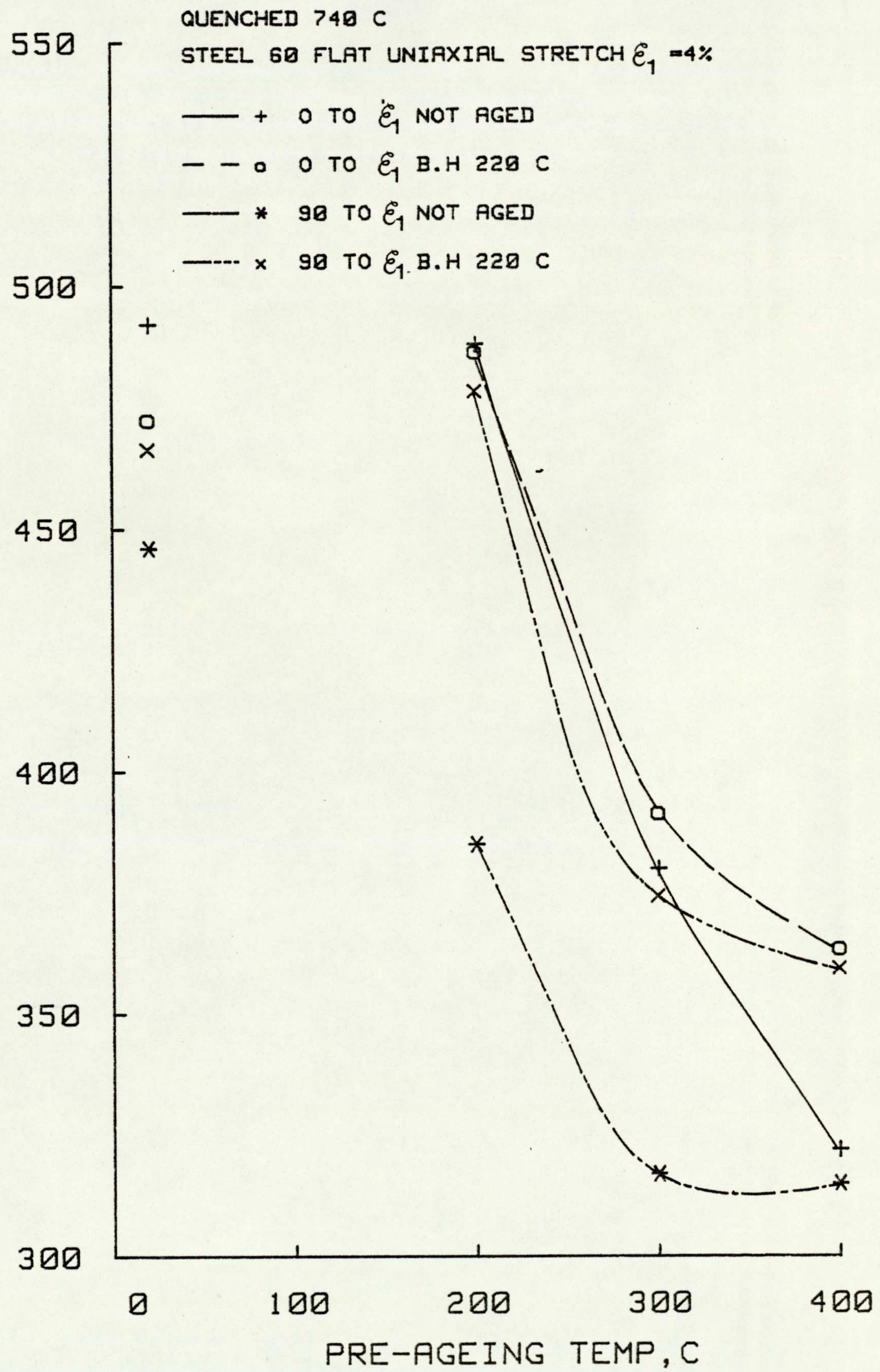


Fig 24. Effect of pre-ageing temperature on 0.5% proof stress values for re-heat treated steel 60, uniaxially prestrained 4% and tested at 0° and 90° to the direction of extension in prestraining. The tests made on samples as-prestrained and after bake-hardening at 220°C for 30 minutes.

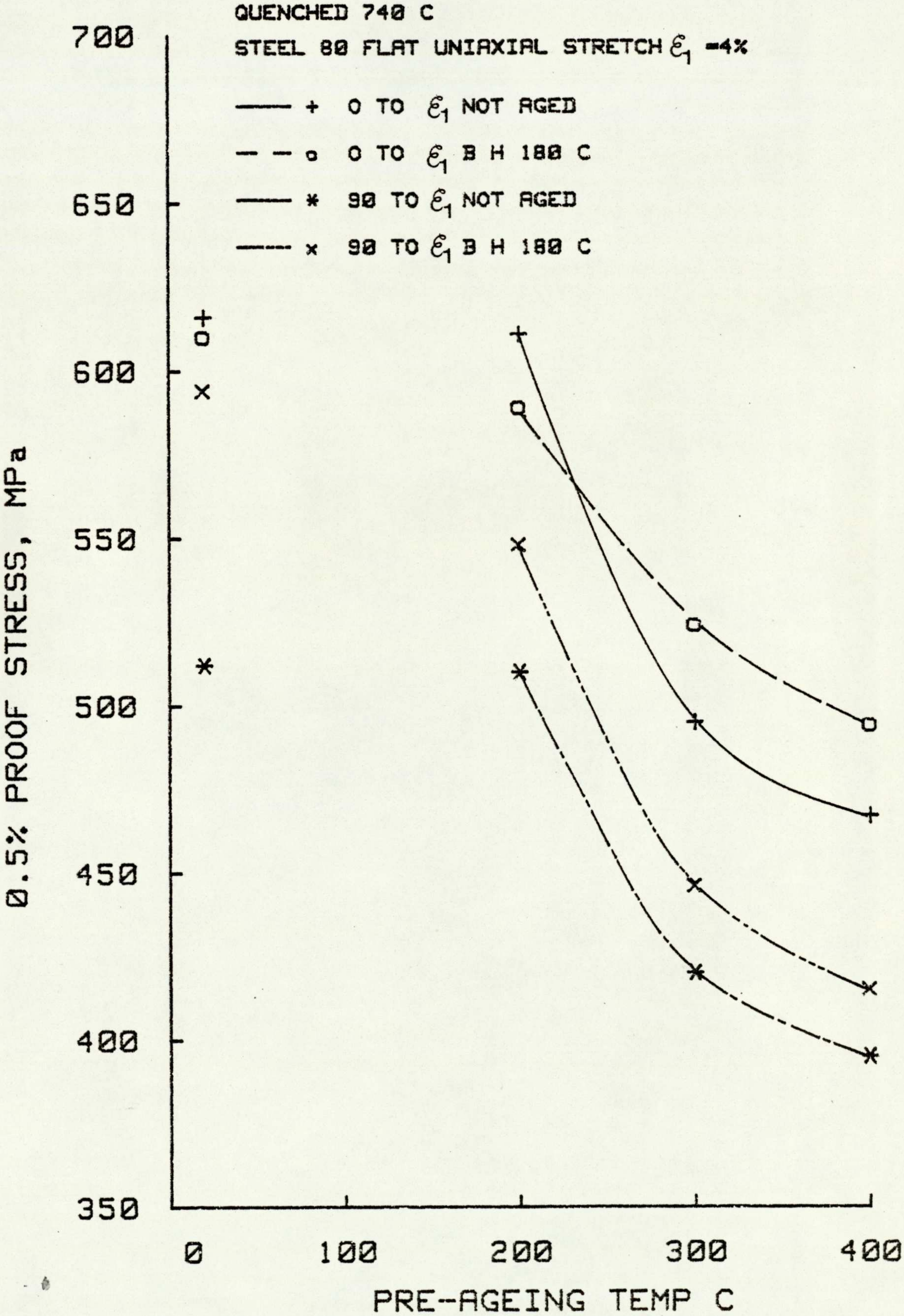


Fig 25. Effect of pre-ageing temperature on 0.5% proof stress values for re-heat treated steel 80, uniaxially prestrained 4% and tested at 0° and 90° to the direction of extension in prestraining. The tests made on samples as-prestrained and after bake-hardening at 180°C for 30 minutes.

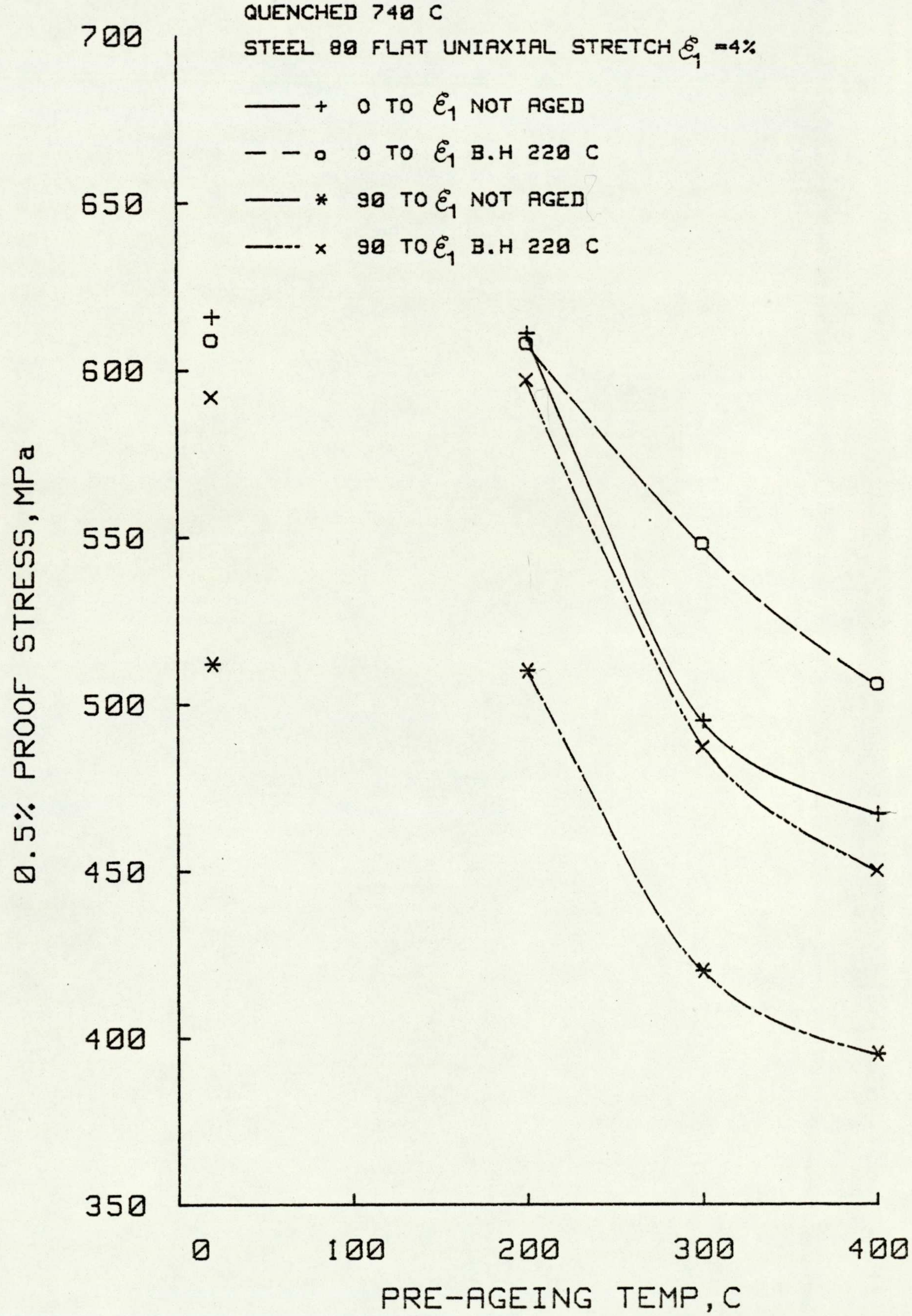


Fig 26. Effect of pre-ageing temperature on 0.5% proof stress values for re-heat treated steel 80, uniaxially prestrained 4% and tested at 0^{o} and 90^{o} to the direction of extension in prestraining. The tests made on samples as-prestrained and after bake-hardening at $220^{o}C$ for 30 minutes.

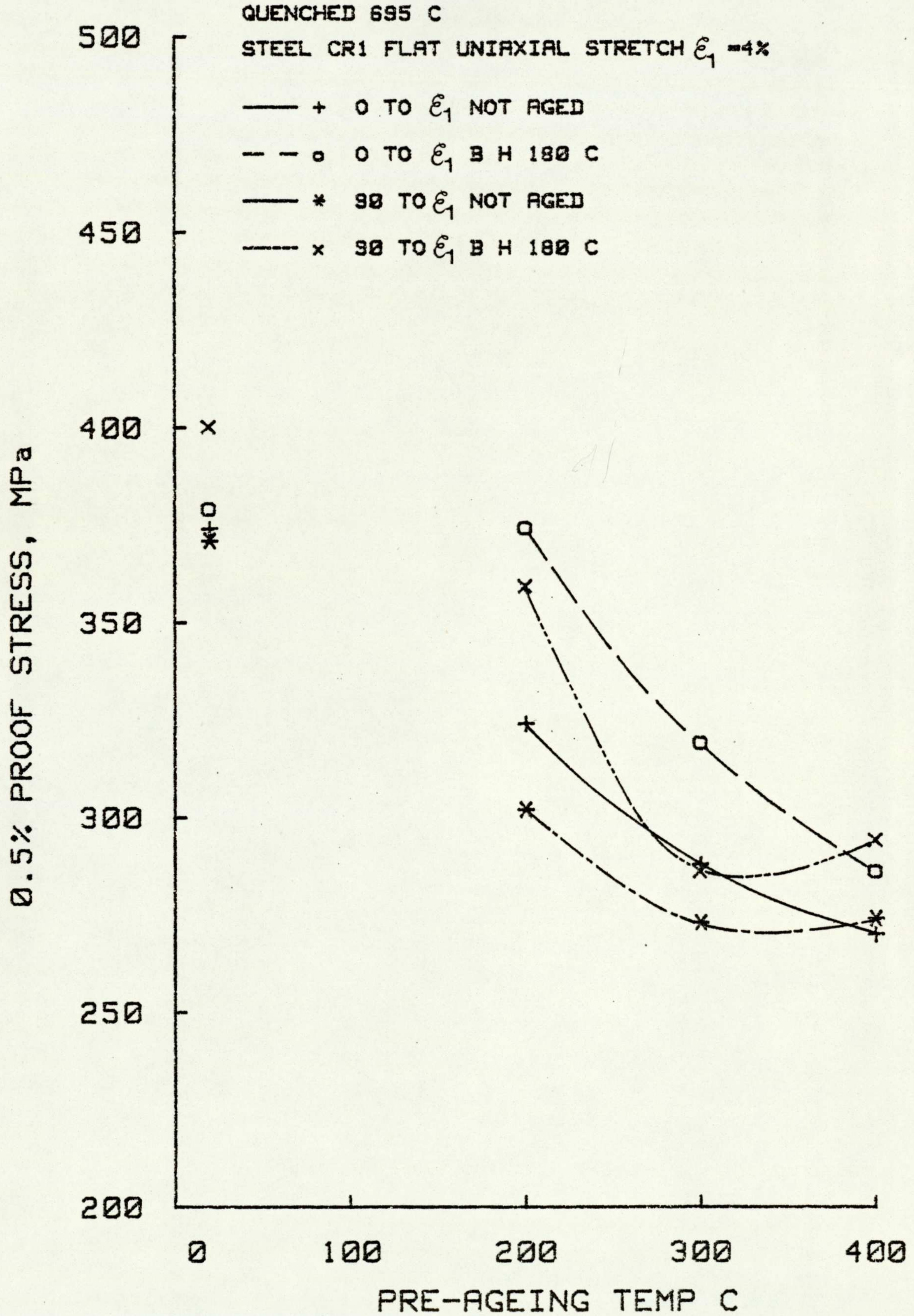


Fig 27. Effect of pre-ageing temperature on 0.5% proof stress values for re-heat treated steel CR1, uniaxially prestrained 4% and tested at 0° and 90° to the direction of extension in prestraining. The tests made on samples as-prestrained and after bake-hardening at 180°C for 30 minutes.

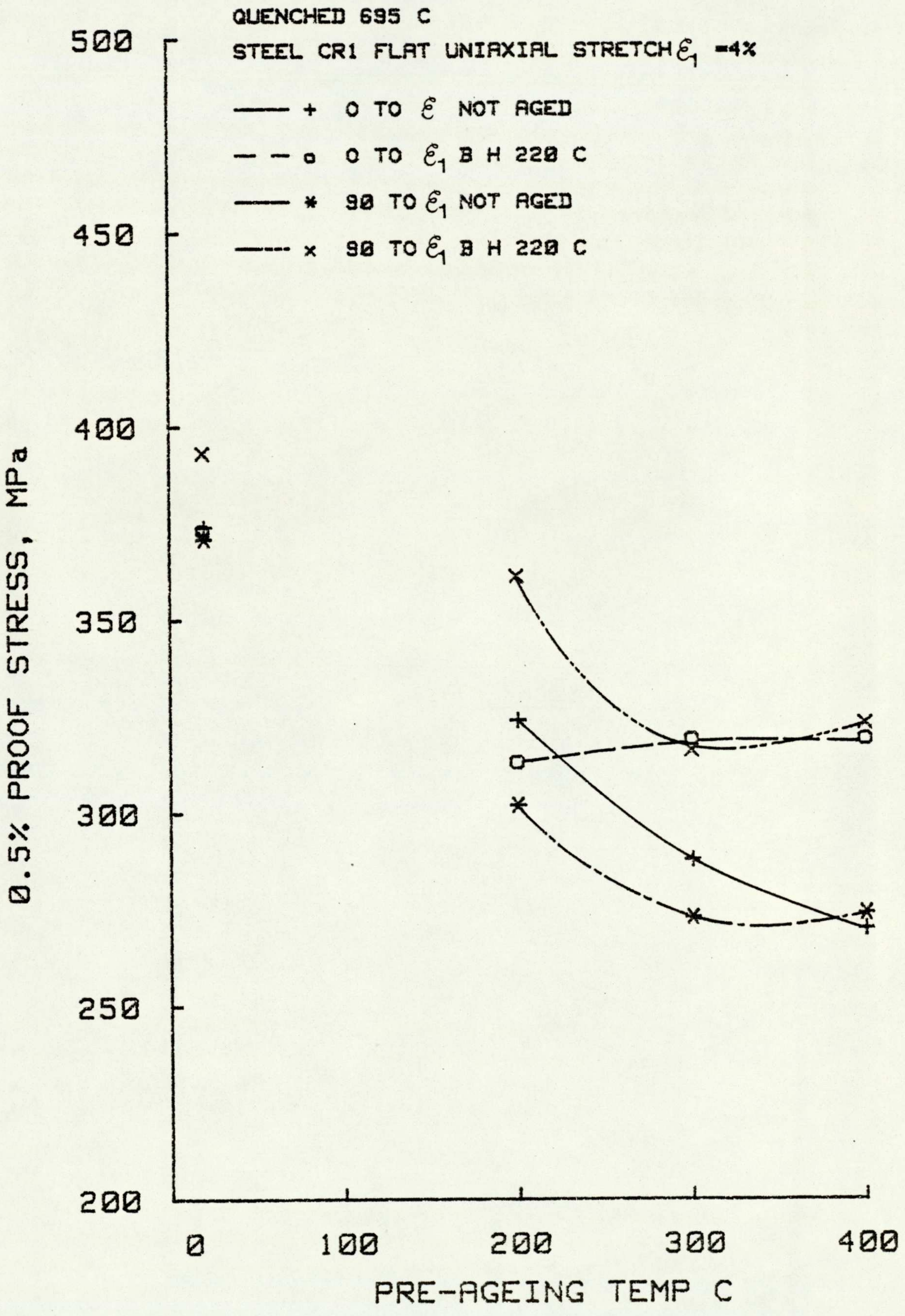


Fig 28. Effect of pre-ageing temperature on 0.5% proof stress values for re-heat treated steel CR1, uniaxially prestrained 4% and tested at 0° and 90° to the direction of extension in prestraining. The tests made on samples as-prestrained and after bake-hardening at 220°C for 30 minutes.

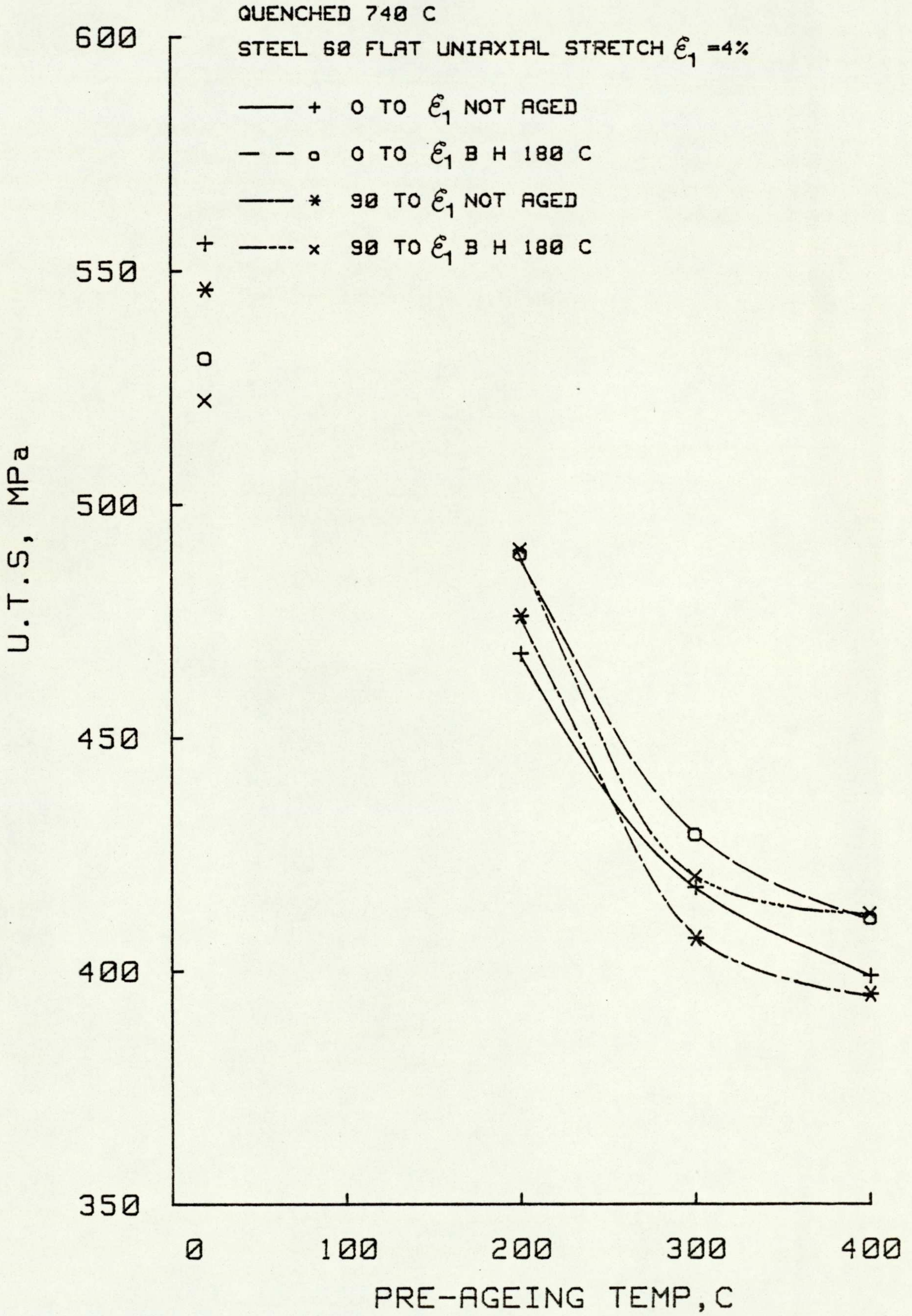


Fig 29. Effect of pre-ageing temperature on UTS values for re-heat treated steel 60, uniaxially prestrained 4% and tested at 0° and 90° to the direction of extension in prestraining. The tests made on samples as-prestrained and after bake-hardening at $180^{\circ}C$ for 30 minutes.

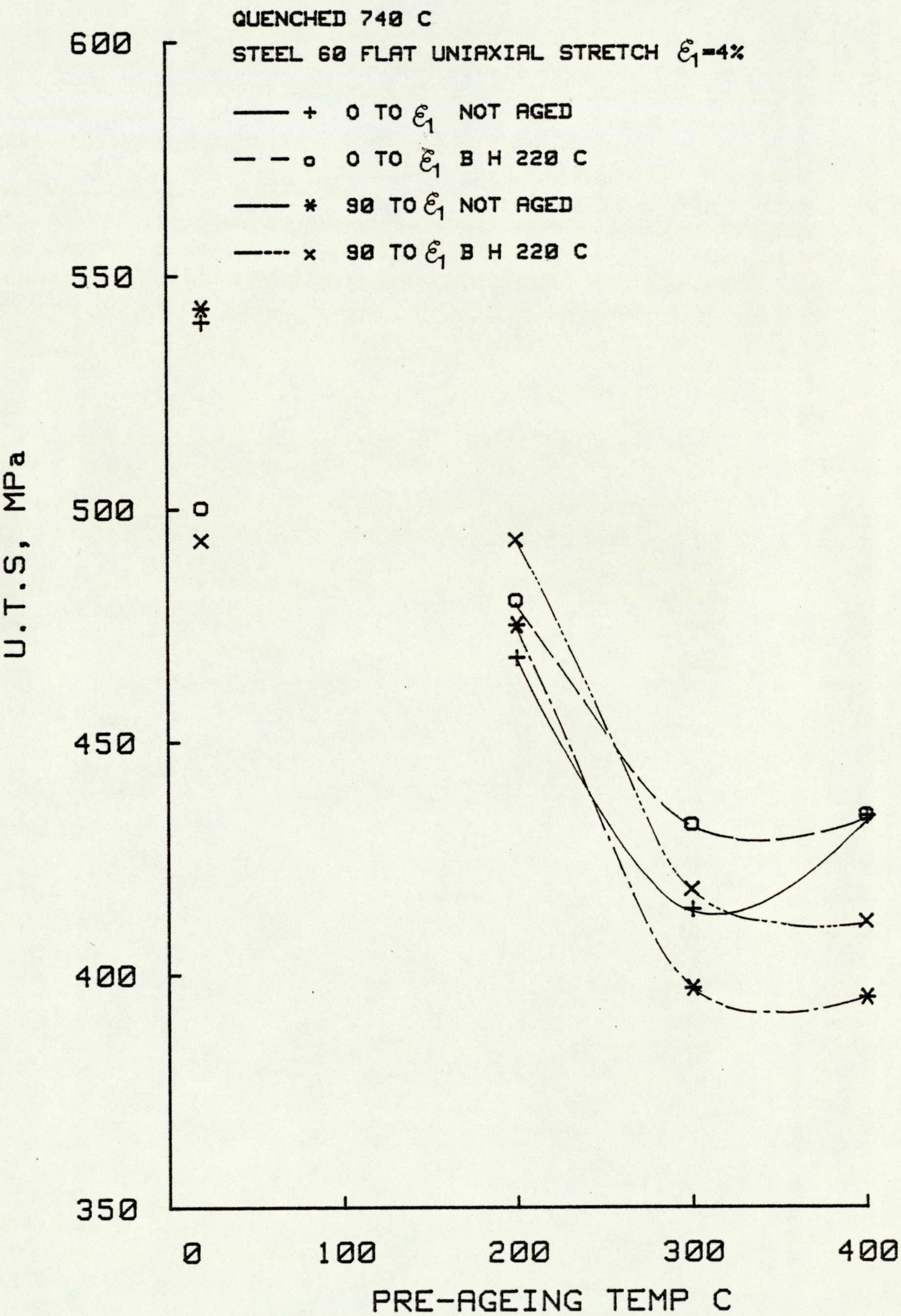


Fig 30. Effect of pre-ageing temperature on UTS values for re-heat treated steel 60, uniaxially prestrained 4% and tested at 0° and 90° to the direction of extension in prestraining. The tests made on samples as-prestrained and after bake-hardening at 220°C for 30 minutes.

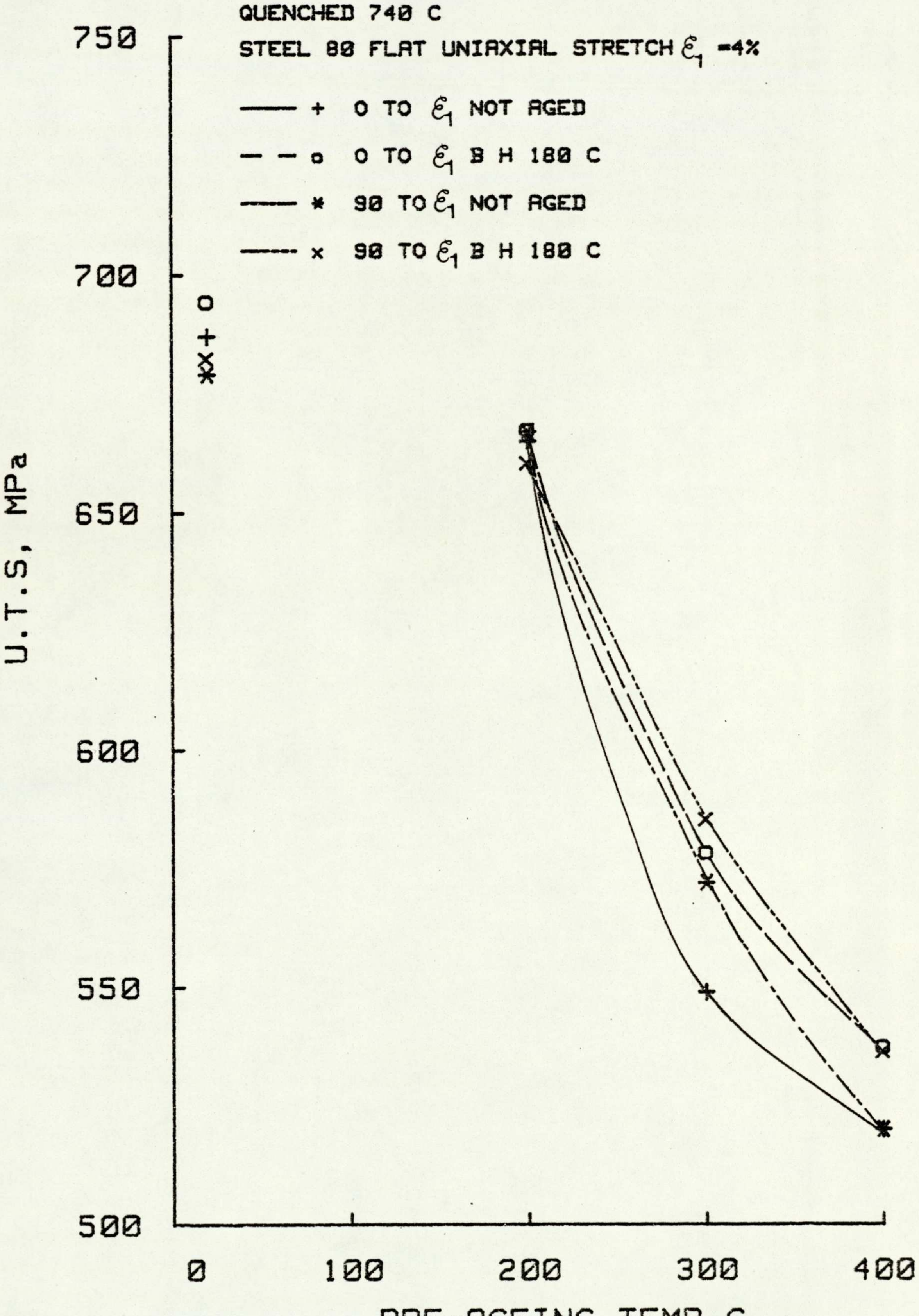


Fig 31. Effect of pre-ageing temperature on UTS values for re-heat treated steel 80, uniaxially prestrained 4% and tested at 0° and 90° to the direction of extension in prestraining. The tests made on samples as-prestrained and after bake-hardening at $180^{\circ}C$ for 30 minutes.

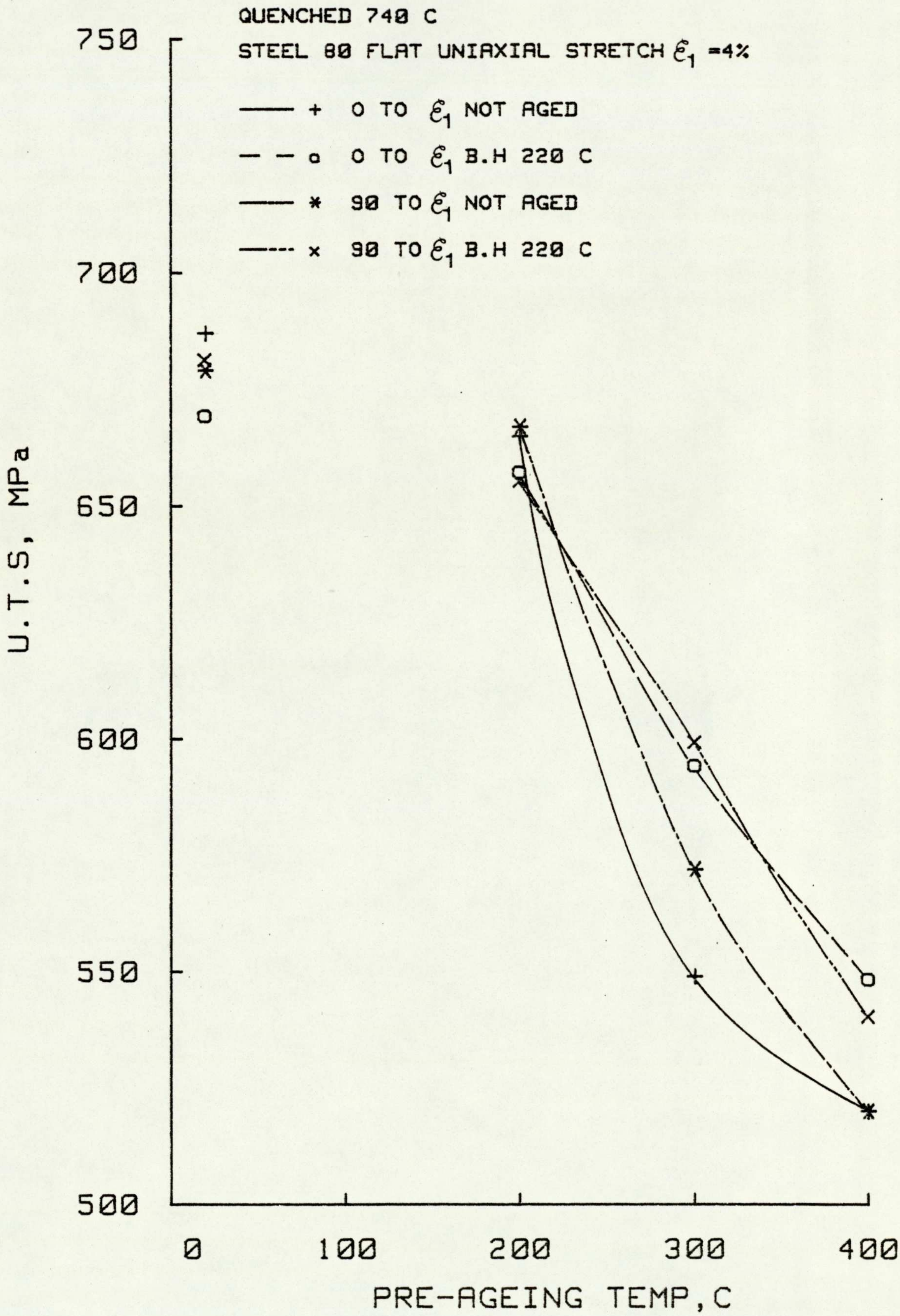


Fig 32. Effect of pre-ageing temperature on UTS values for re-heat treated steel 80, uniaxially pre-strained 4% and tested at 0° and 90° to the direction of extension in prestraining. The tests made on samples as-prestrained and after bake-hardening at $220^{\circ}C$ for 30 minutes.

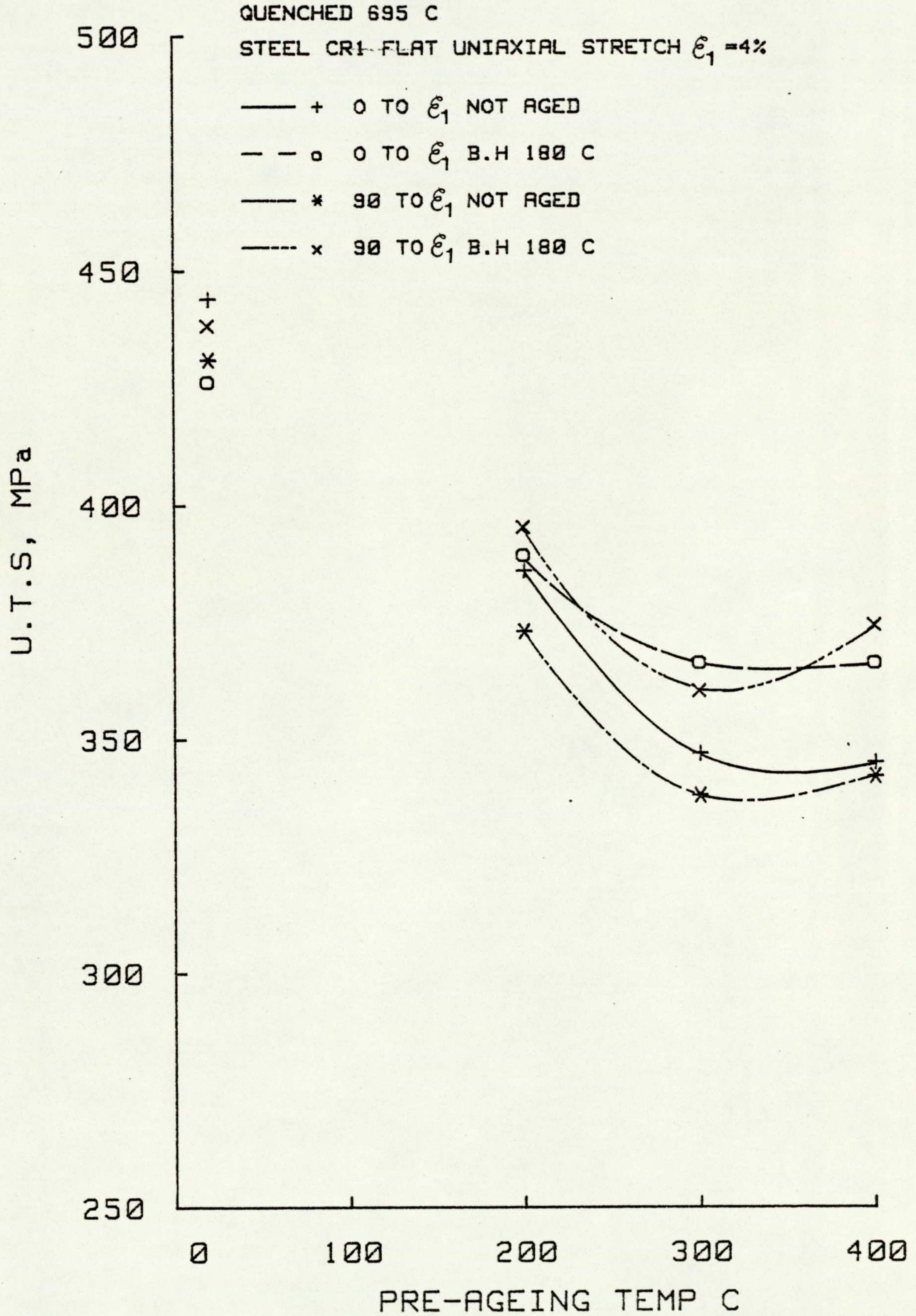


Fig 33. Effect of pre-ageing temperature on UTS values for re-heat treated steel CR1, uniaxially prestrained 4% and tested at 0° and 90° to the direction of extension in prestraining. The tests made on samples as-prestrained and after bake-hardening at $180^\circ C$ for 30 minutes.

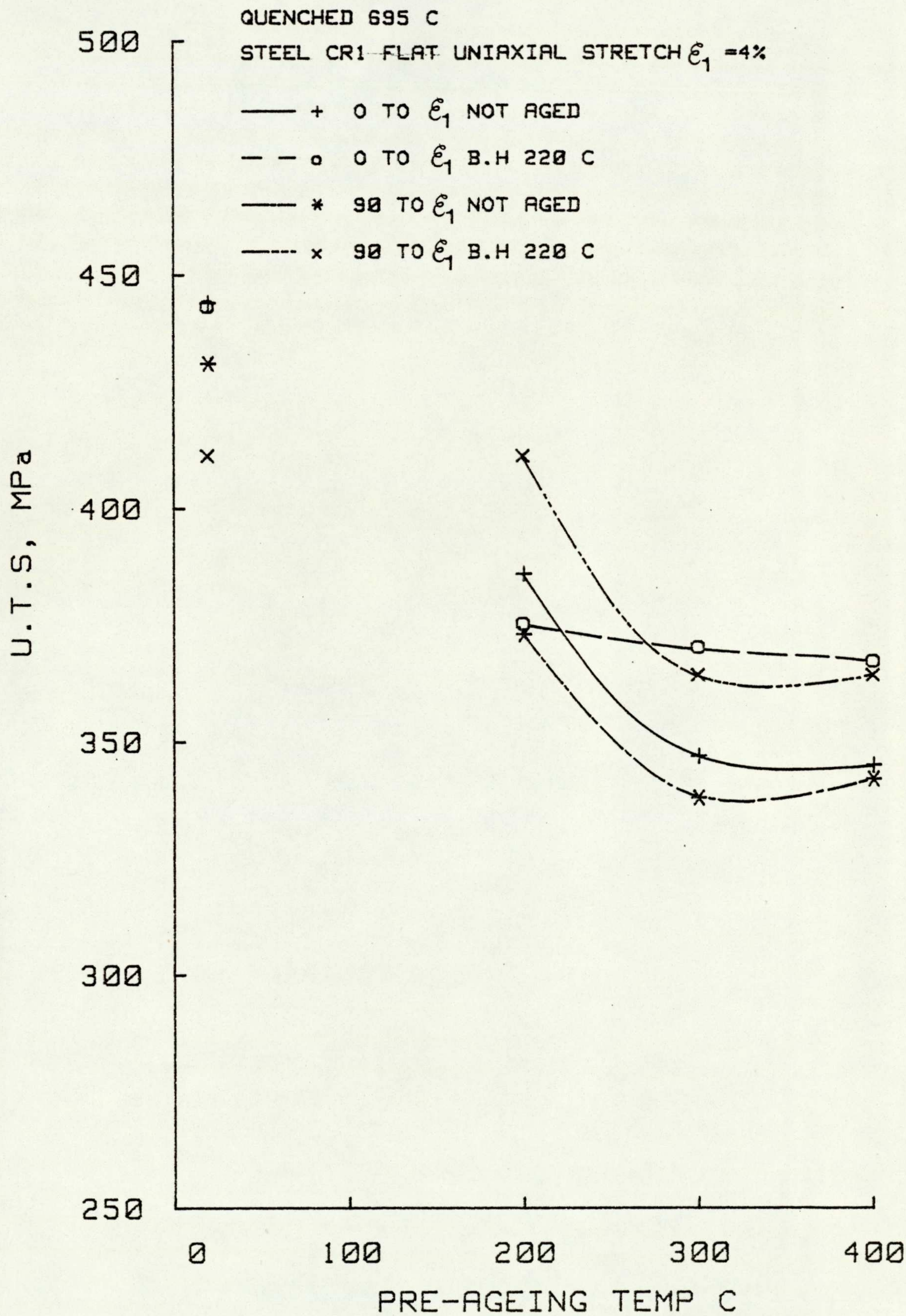


Fig 34. Effect of pre-ageing temperature on UTS values for re-heat treated steel CR1, uniaxially pre-strained 4% and tested at 0° and 90° to the direction of extension in prestraining. The tests made on samples as-prestrained and after bake-hardening at $220^{\circ}C$ for 30 minutes.

Fig 35 a-d

Effect of ageing time at 47°C on steel 60, temper-rolled ~ 1% after quenching from 740°C and preageing at 200°, 300° and 400°C (for 3 minutes) on 0.5% proof stress, UTS and total elongation.

(a) as quenched (b) preaged at 200°C

(c) preaged at 300°C (d) preaged at 400°C.

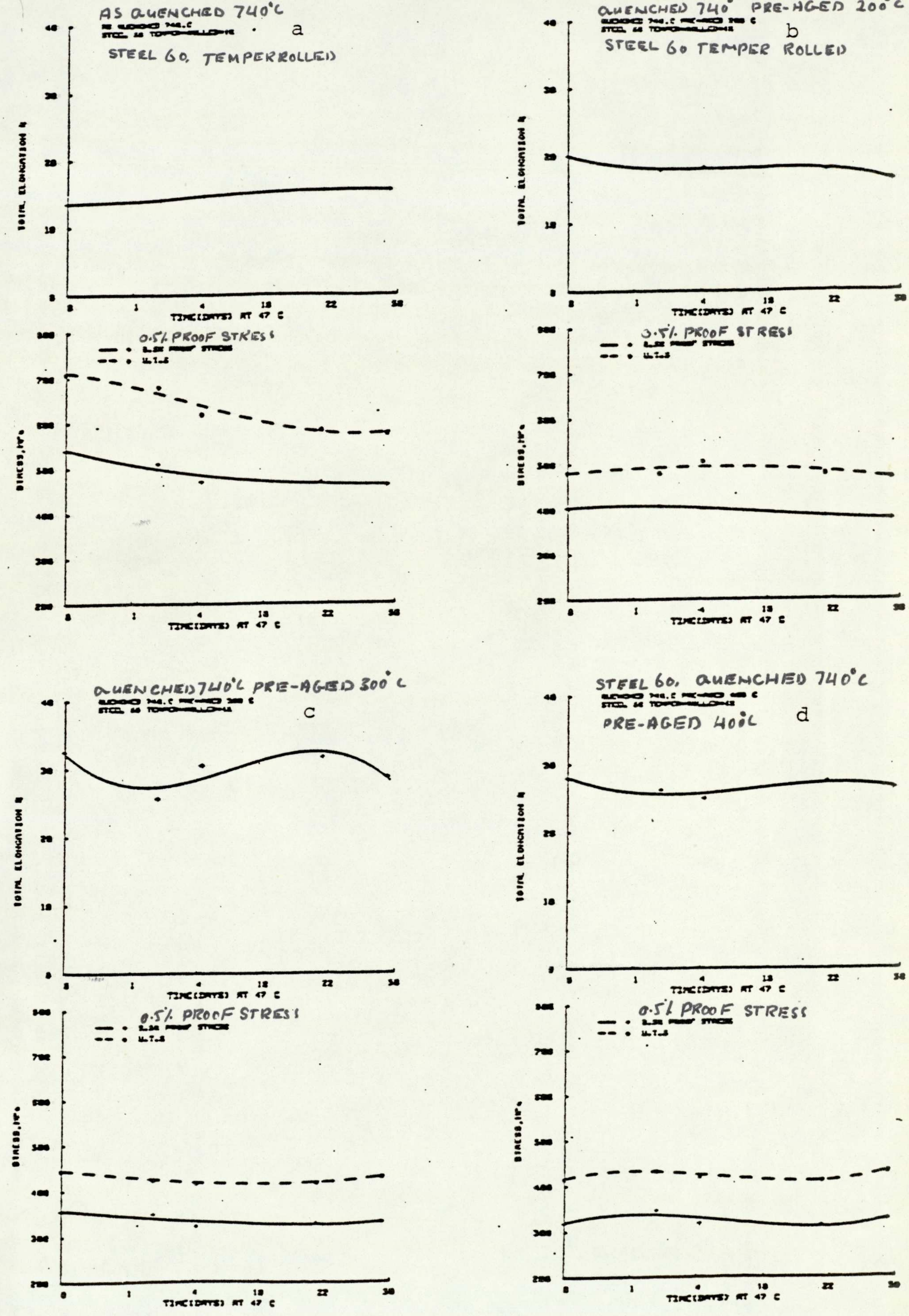

AS QUENCHED 740°C
a
STEEL 60, TEMPERROLLED
TOTAL ELONGATION %
TIME(DAYS) AT 47 C
0.5% PROOF STRESS
U.T.S
STRESS, MPa
QUENCHED 740° PRE-AGED 200°C
b
STEEL 60 TEMPER ROLLED
QUENCHED 740°C PRE-AGED 300°C
c
STEEL 60, QUENCHED 740°C
PRE-AGED 400°C
d

Fig 35 a-d

Fig 36 a-d

Effect of ageing time at 47°C on steel 80, temper-rolled ~ 1% after quenching from 740°C and preageing at 200°, 300° and 400°C (for 3 minutes) on 0.5% proof stress, UTS and total elongation.

(a) as quenched (b) preaged at 200°C

(c) preaged at 300°C (d) preaged at 400°C

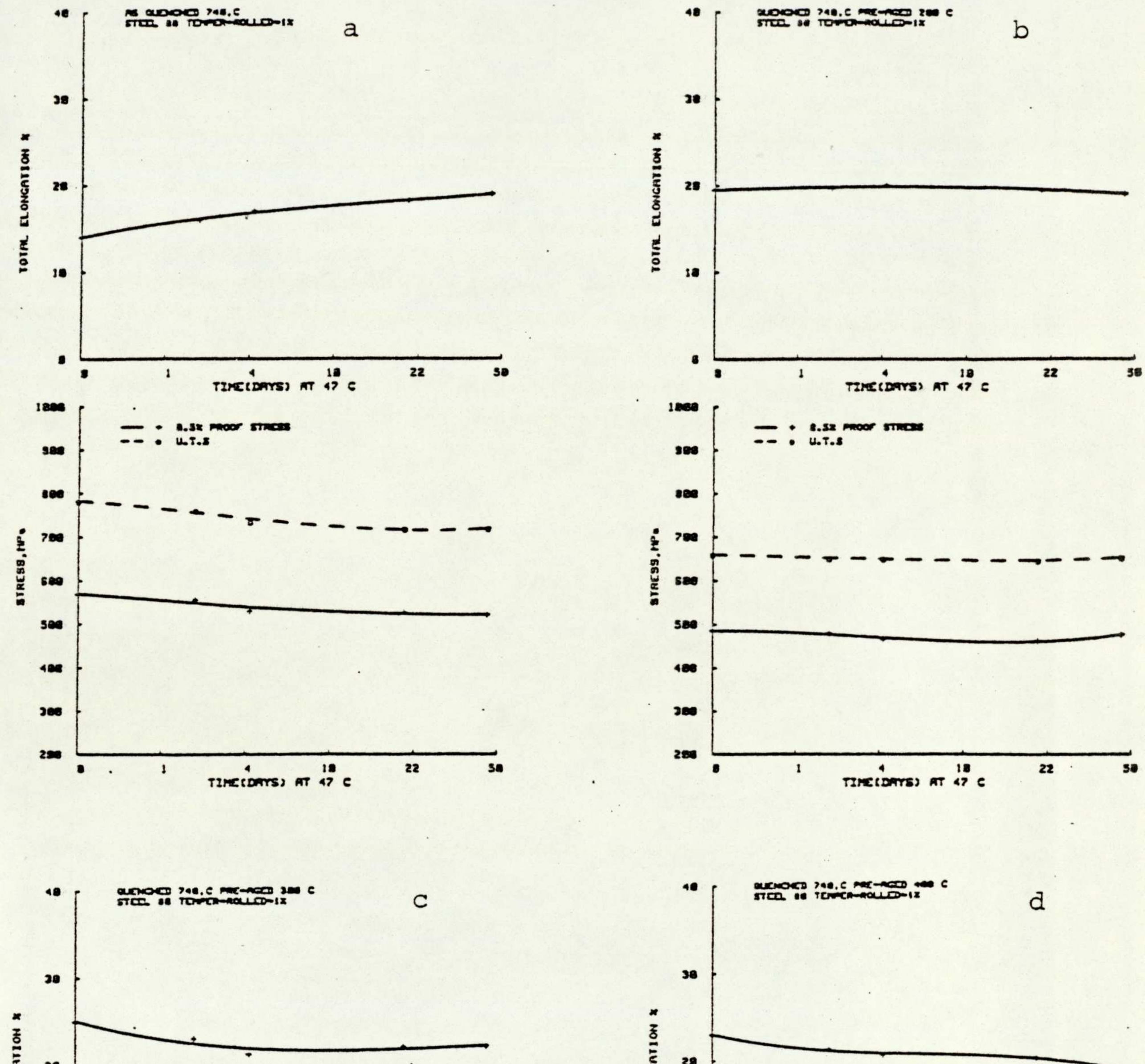

a
b
c
d
TOTAL ELONGATION %
TIME(DAYS) AT 47 C
STRESS, MPa
0.5% PROOF STRESS
U.T.S

Fig 36 a-d

Fig 37 a-d

Effect of ageing time at 47°C on steel CR1, temper-rolled ~ 1% after quenching from 695°C and preageing at 200°, 300° and 400°C (for 3 minutes) on 0.5% proof stress, UTS and total elongation.

(a) as quenched (b) preaged at 200°C

(c) preaged at 300°C (d) preaged at 400°C

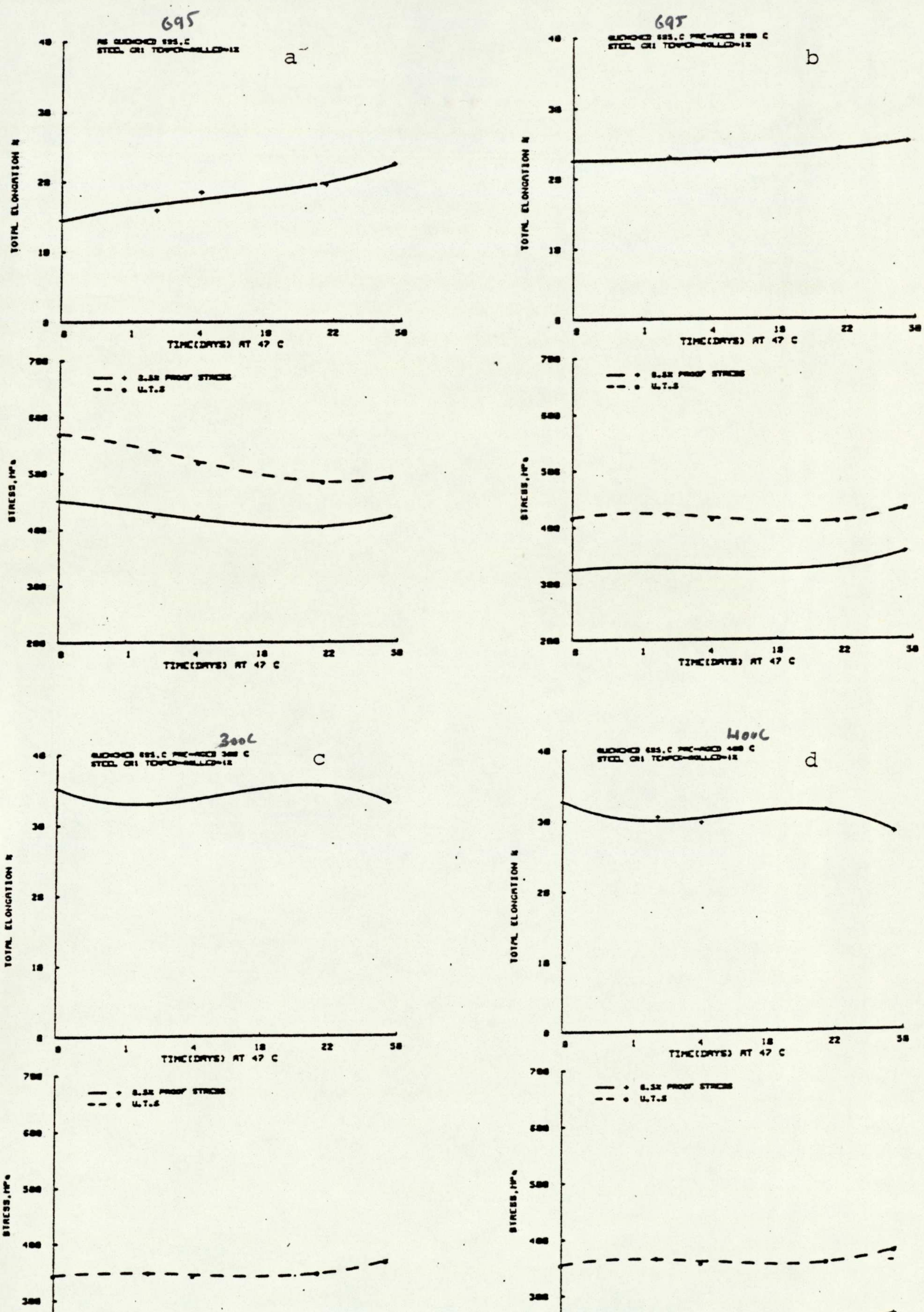


Fig 37 a-d

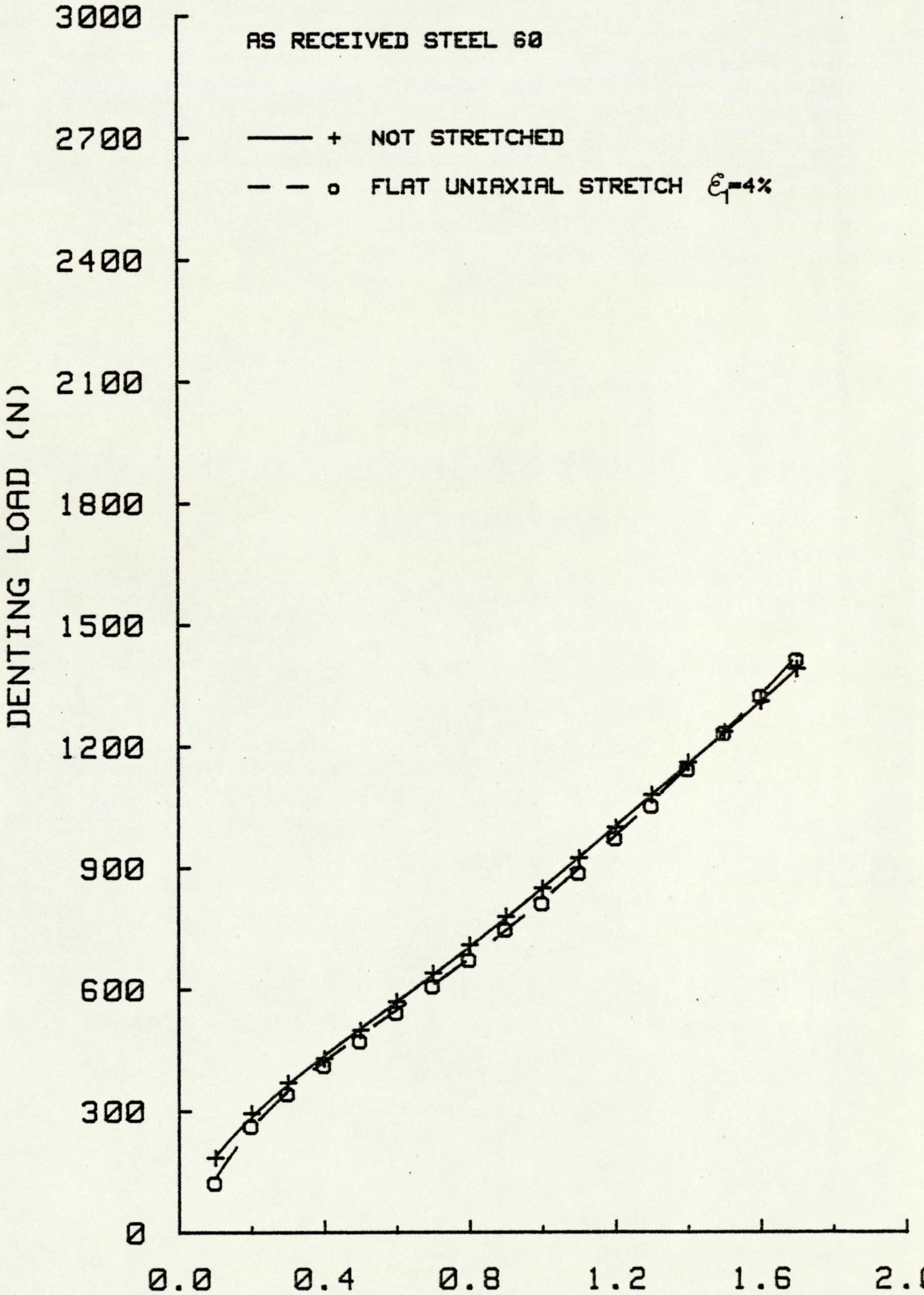


Fig 38. Dent resistance of flat panels of as-received dual-phase 60.

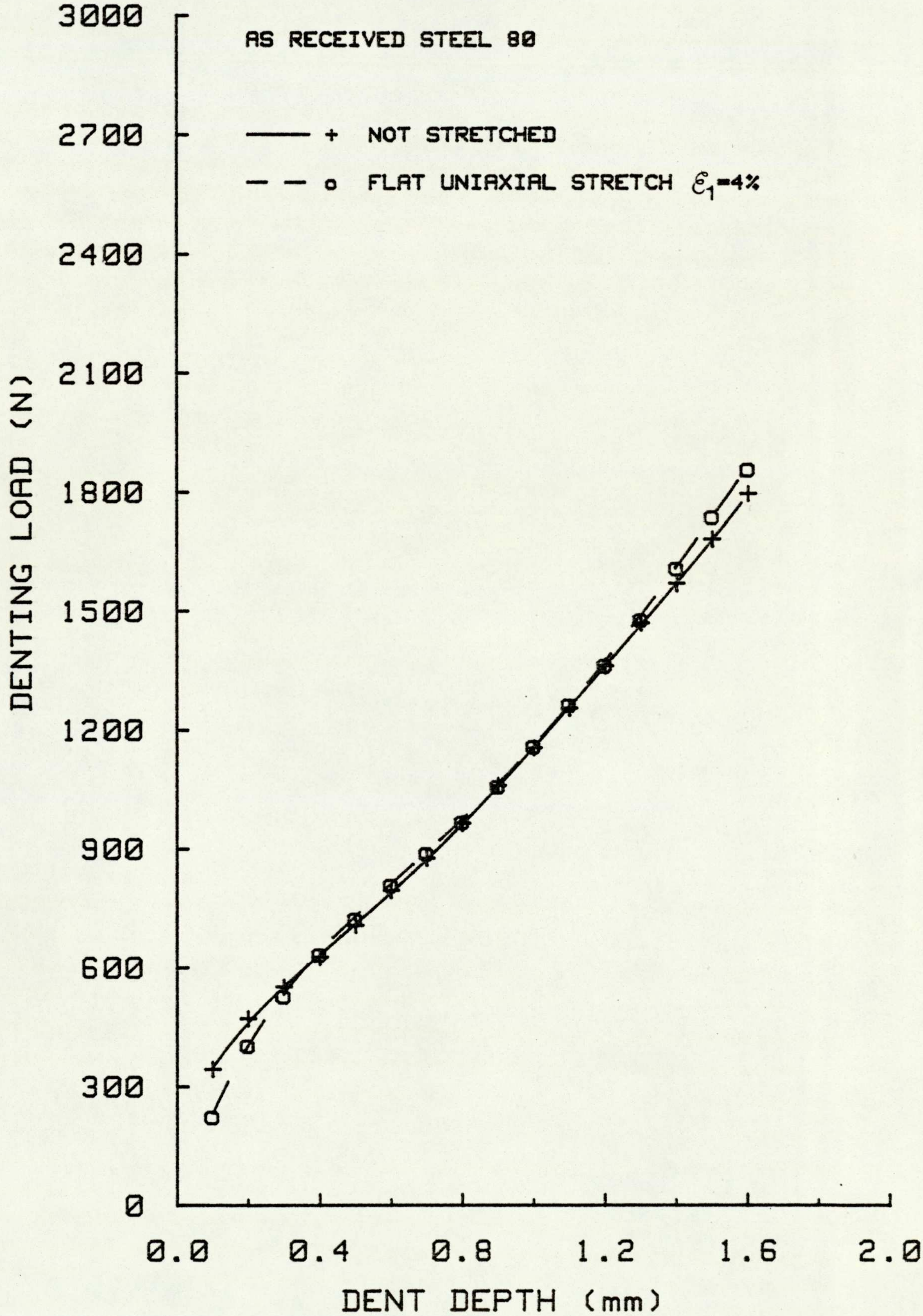


Fig 39. Dent resistance of flat panels of as-received dual-phase 80.

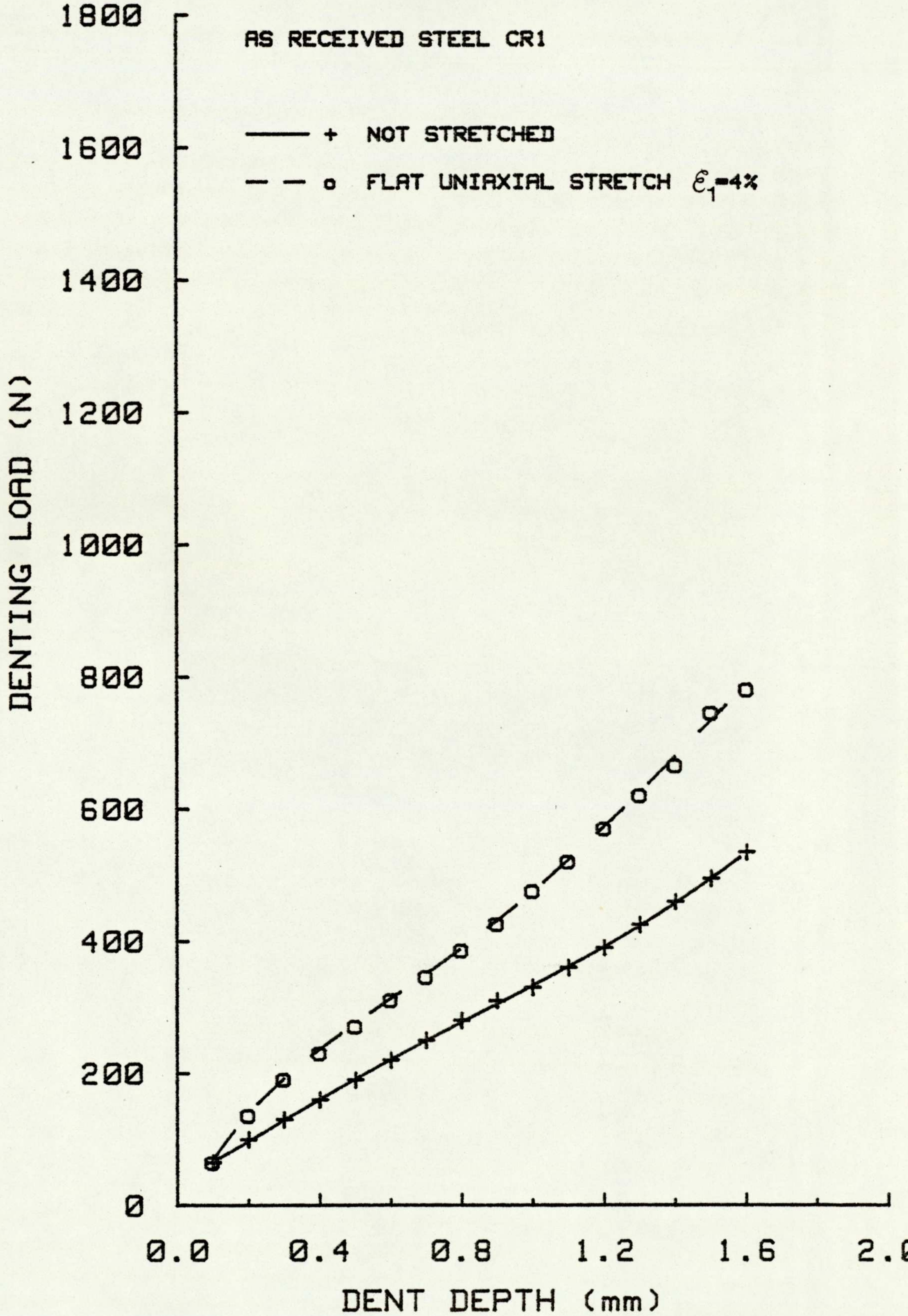


Fig 40. Dent resistance of flat panels of as-received CR1.

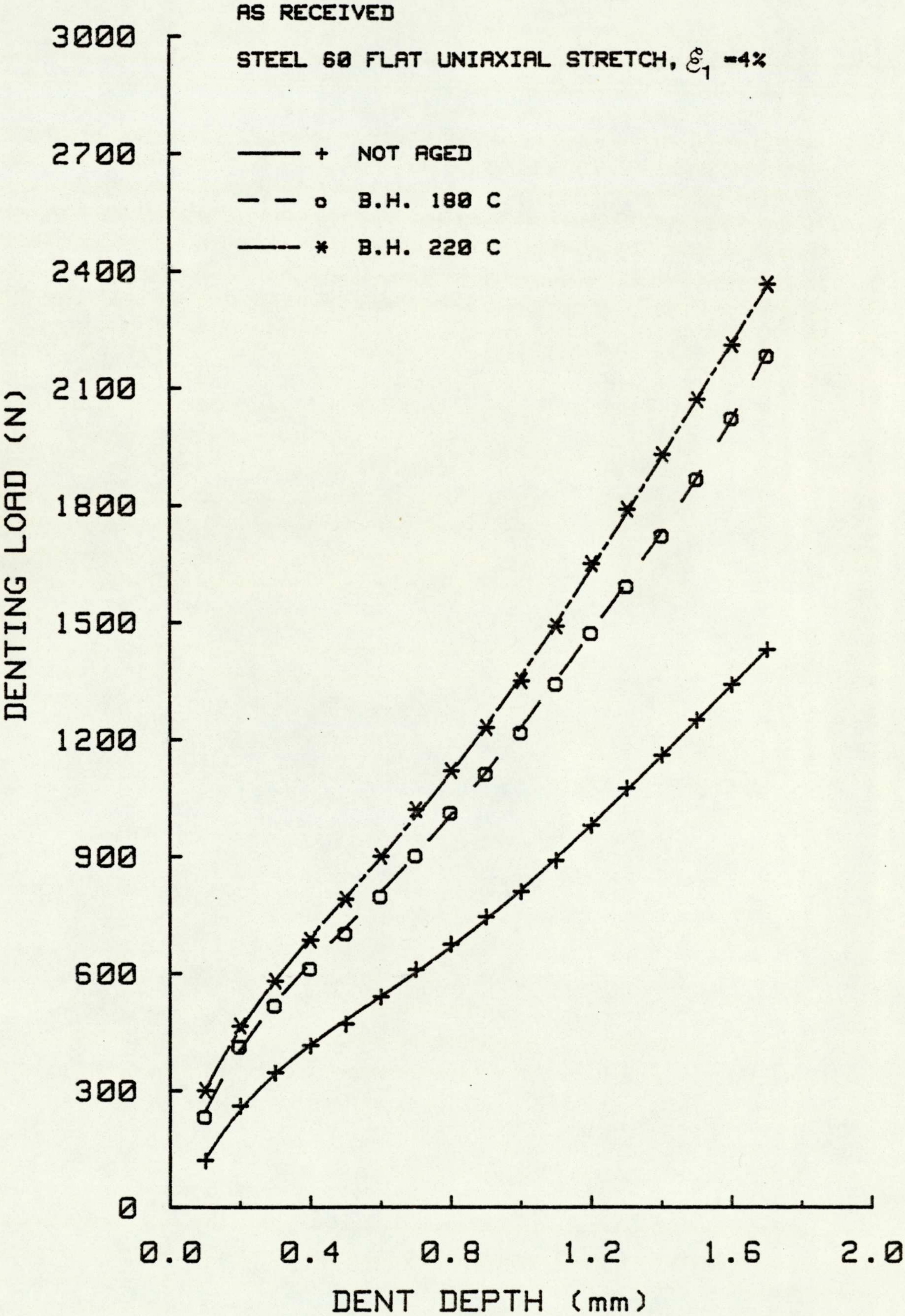


Fig 41. Effect of bake-hardening at 180° and 220°C for 30 minutes on dent-resistance of as-received steel 60, stretched 4% in uniaxial tension.

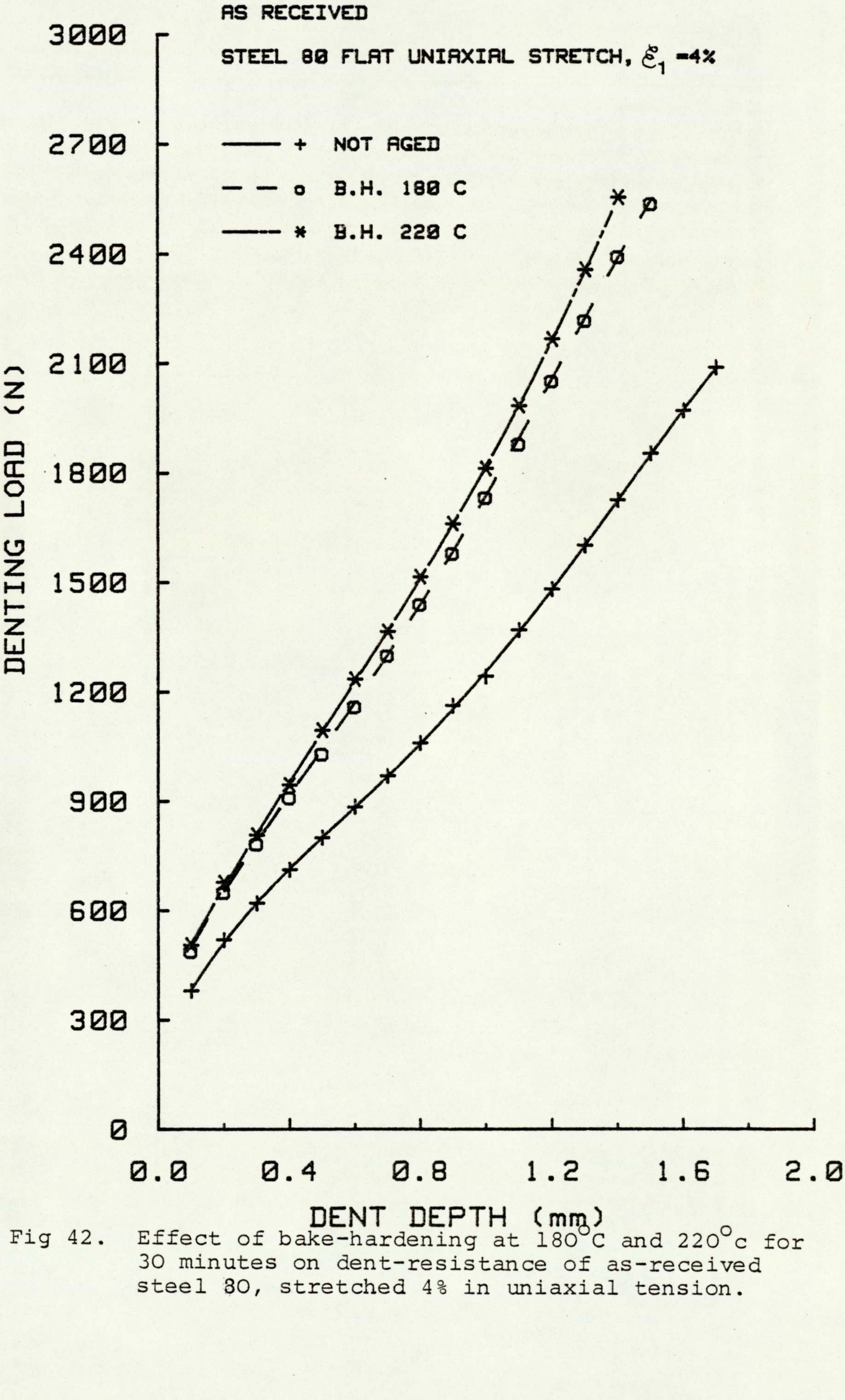


Fig 42. Effect of bake-hardening at 180°C and 220°c for 30 minutes on dent-resistance of as-received steel 80, stretched 4% in uniaxial tension.

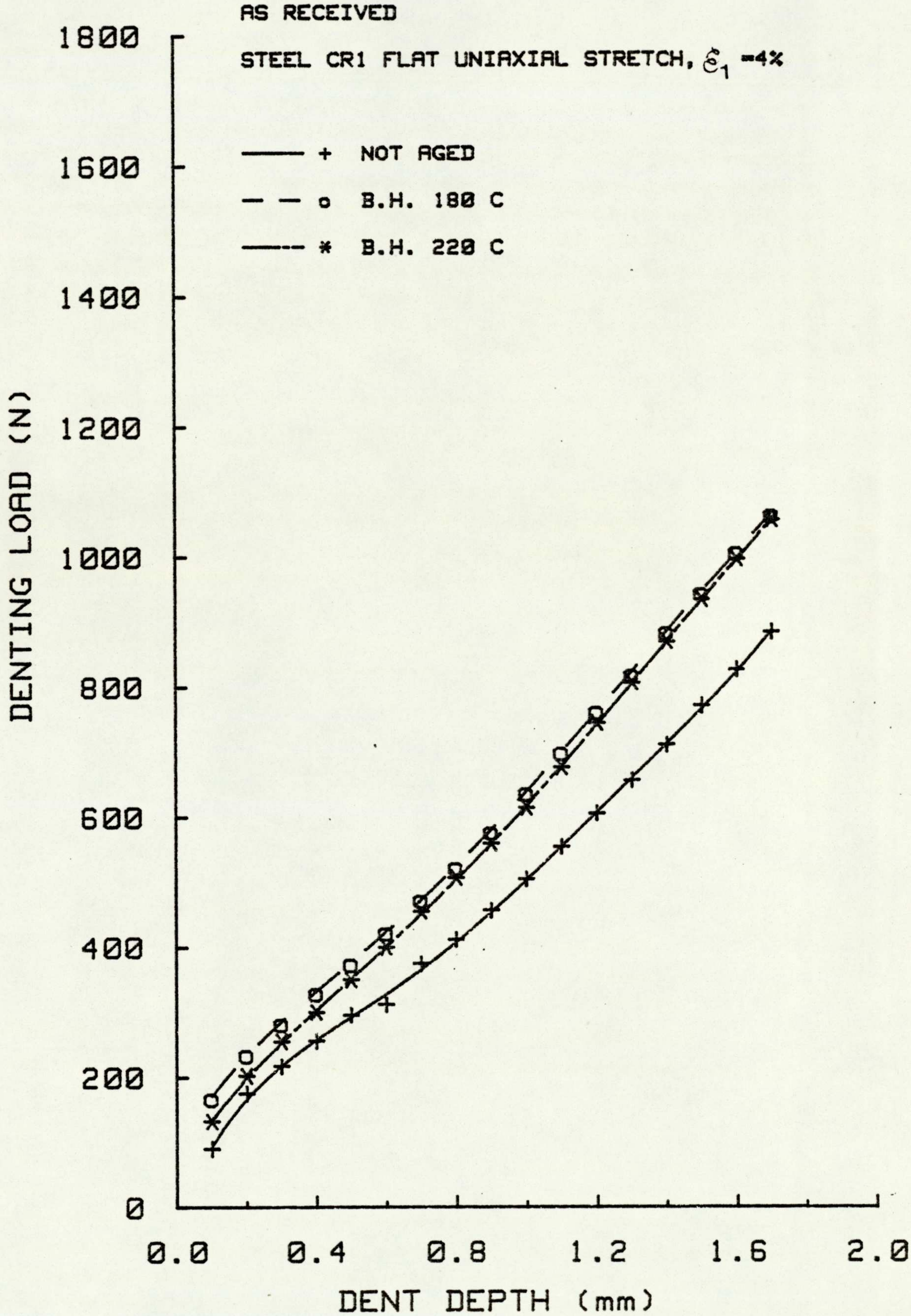


Fig 43. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent-resistance of as-received steel CR1 stretched 4% in uniaxial tension.

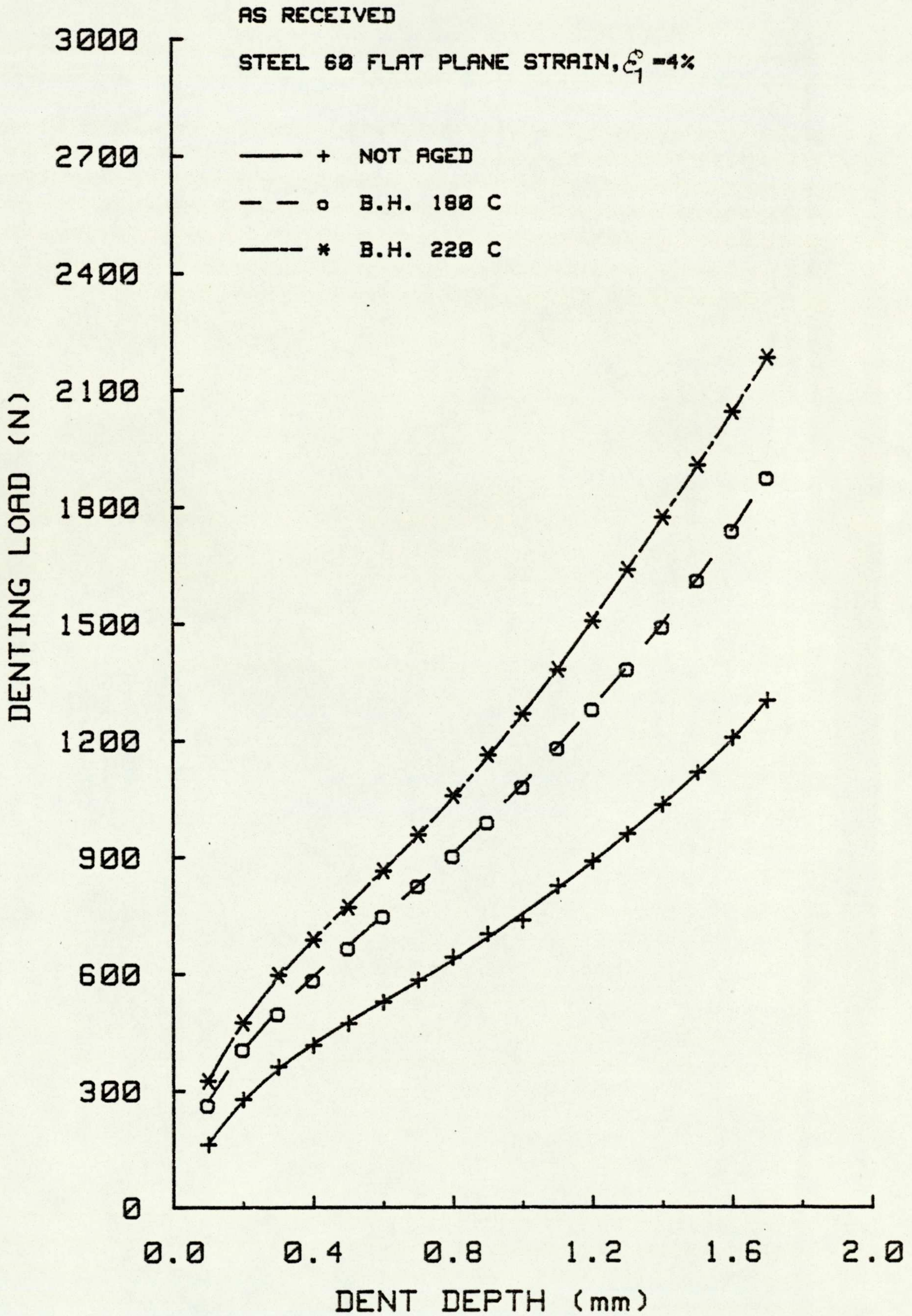


Fig 44. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent-resistance of as-received steel 60, stretched 4% with plane strain.

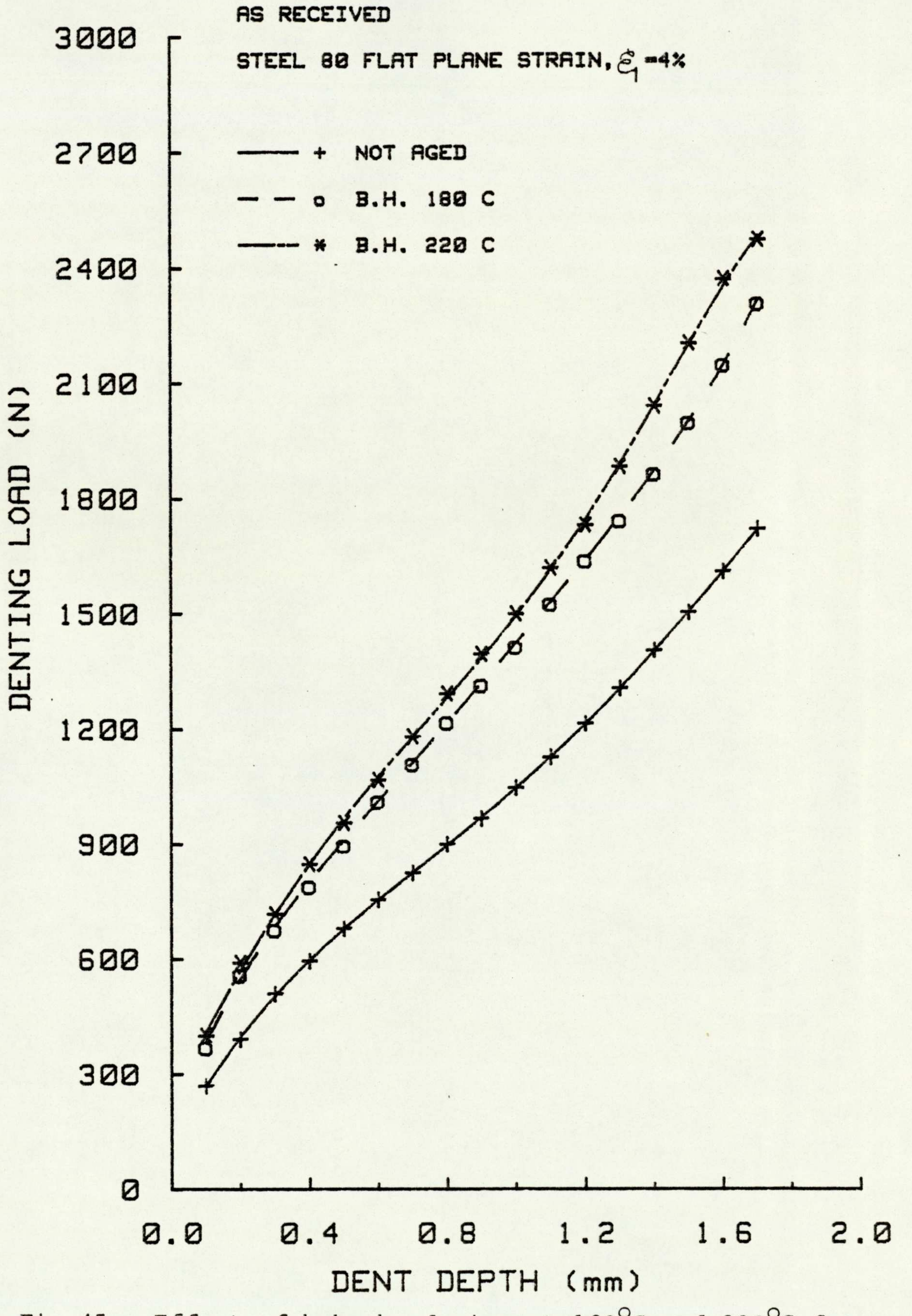


Fig 45. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent-resistance of as-received steel 80 stretched 4% with plane strain.

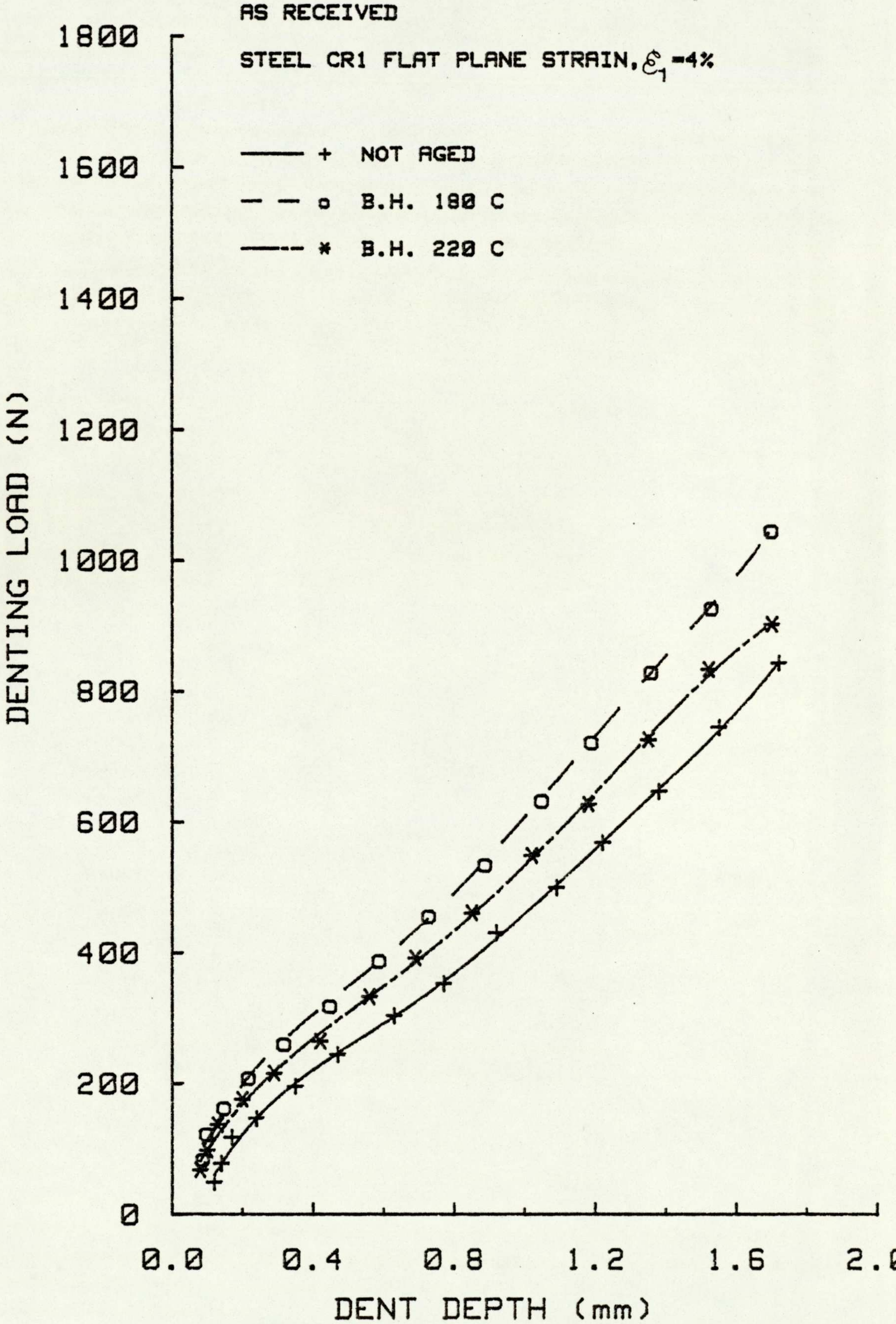


Fig 46. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent-resistance of as-received steel CR1 stretched 4% with plane strain.

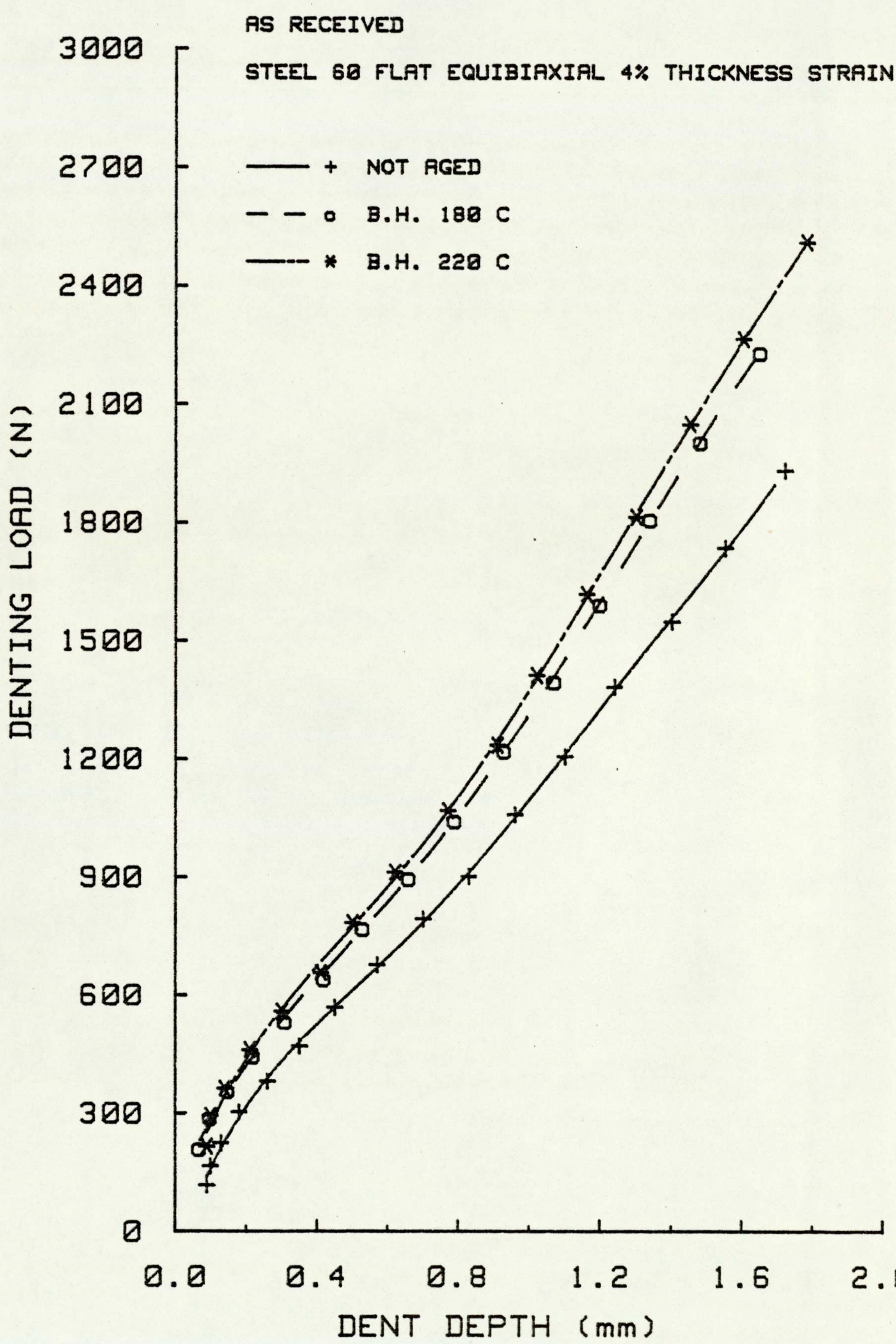


Fig 47. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent-resistance of flat panels as-received steel 60, stretched to 4% thickness strain in equibiaxial stretching.

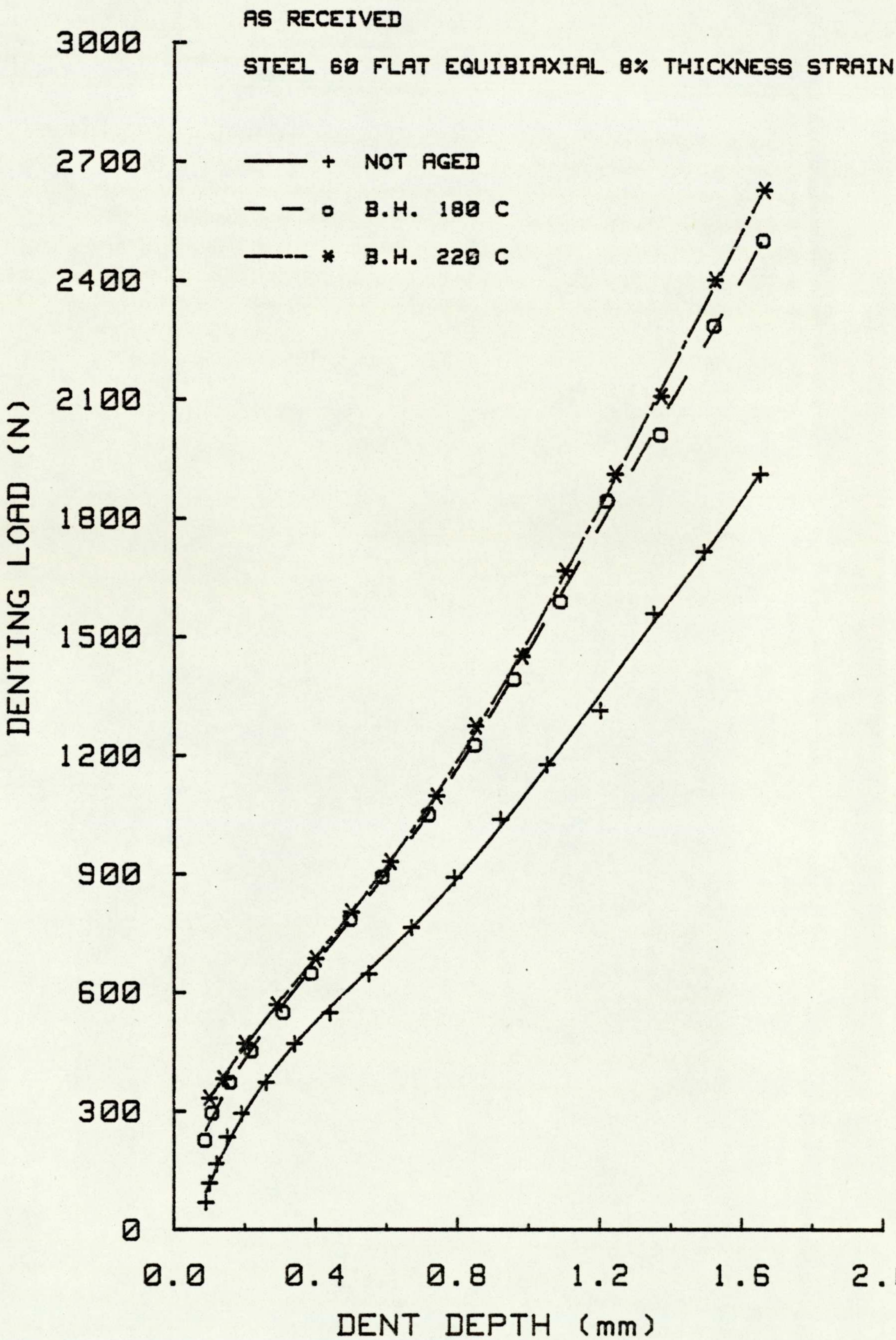


Fig 48. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent-resistance of flat panels as-received steel 60, stretched to 8% thickness strain in equibiaxial stretching.

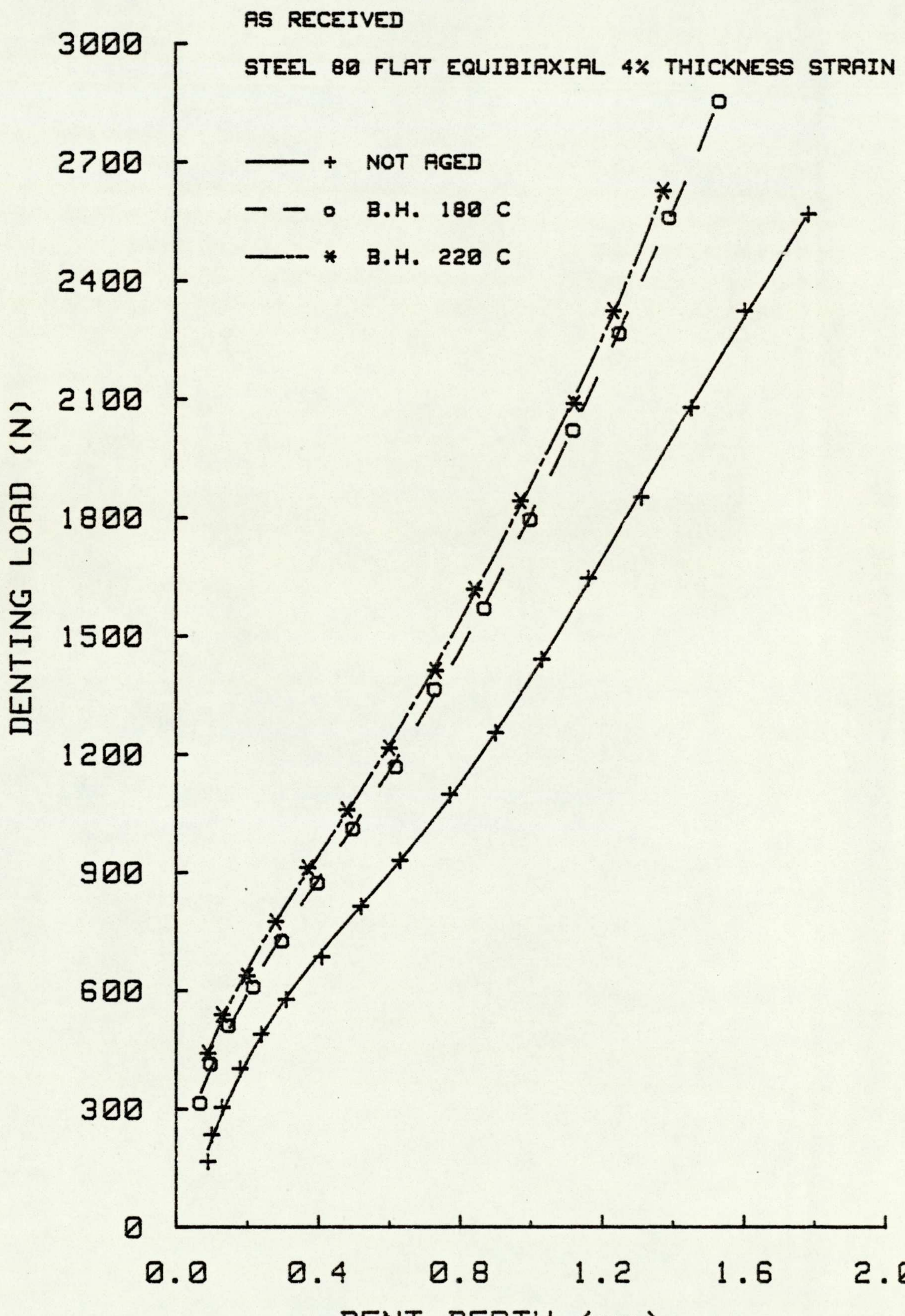


Fig 49. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent-resistance of flat panels as received steel 80, stretched to 4% thickness strain in equibiaxial stretching.

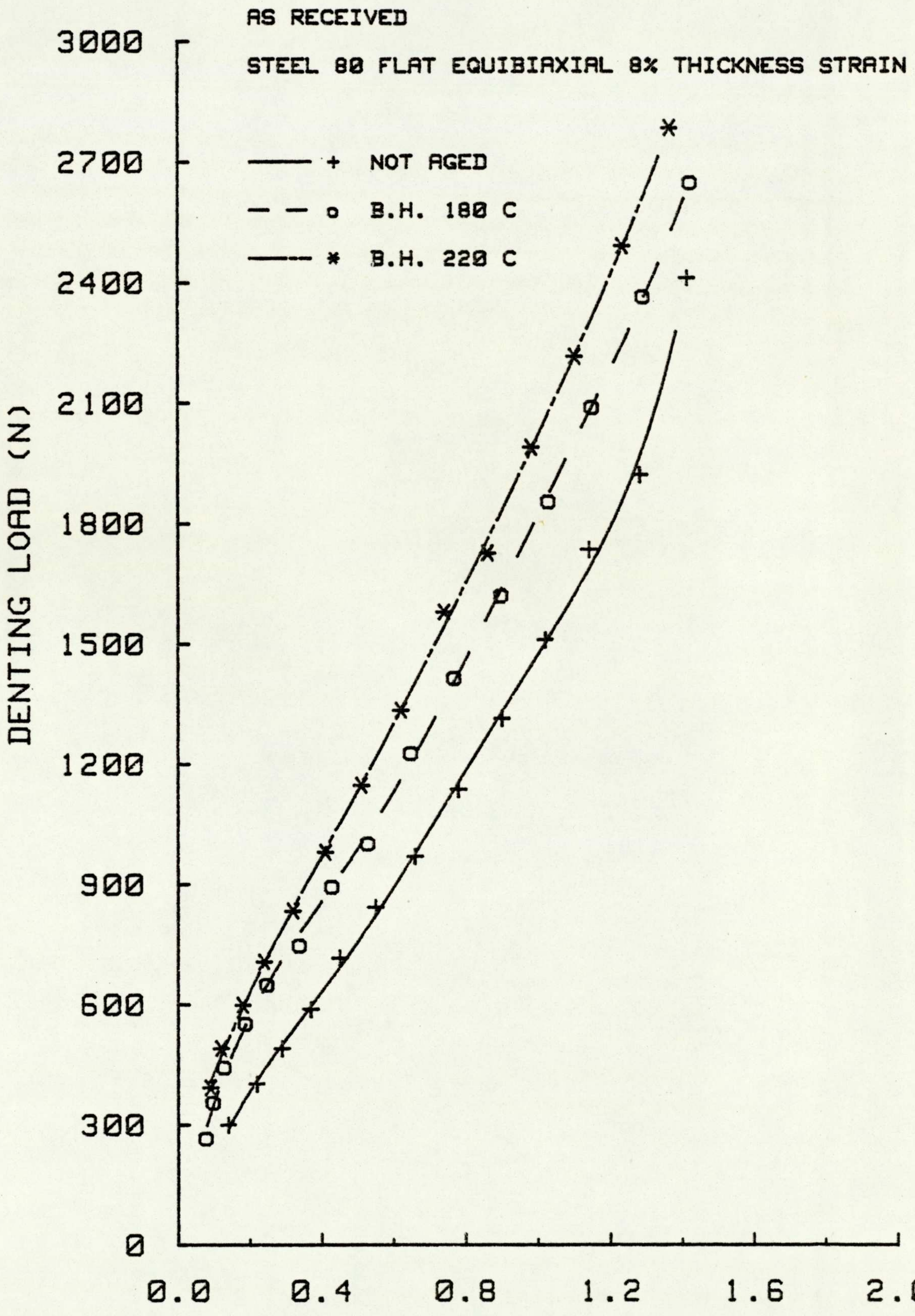


Fig 50. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent-resistance of flat panels as-received steel 80, stretched to 8% thickness strain in equibiaxial stretching.

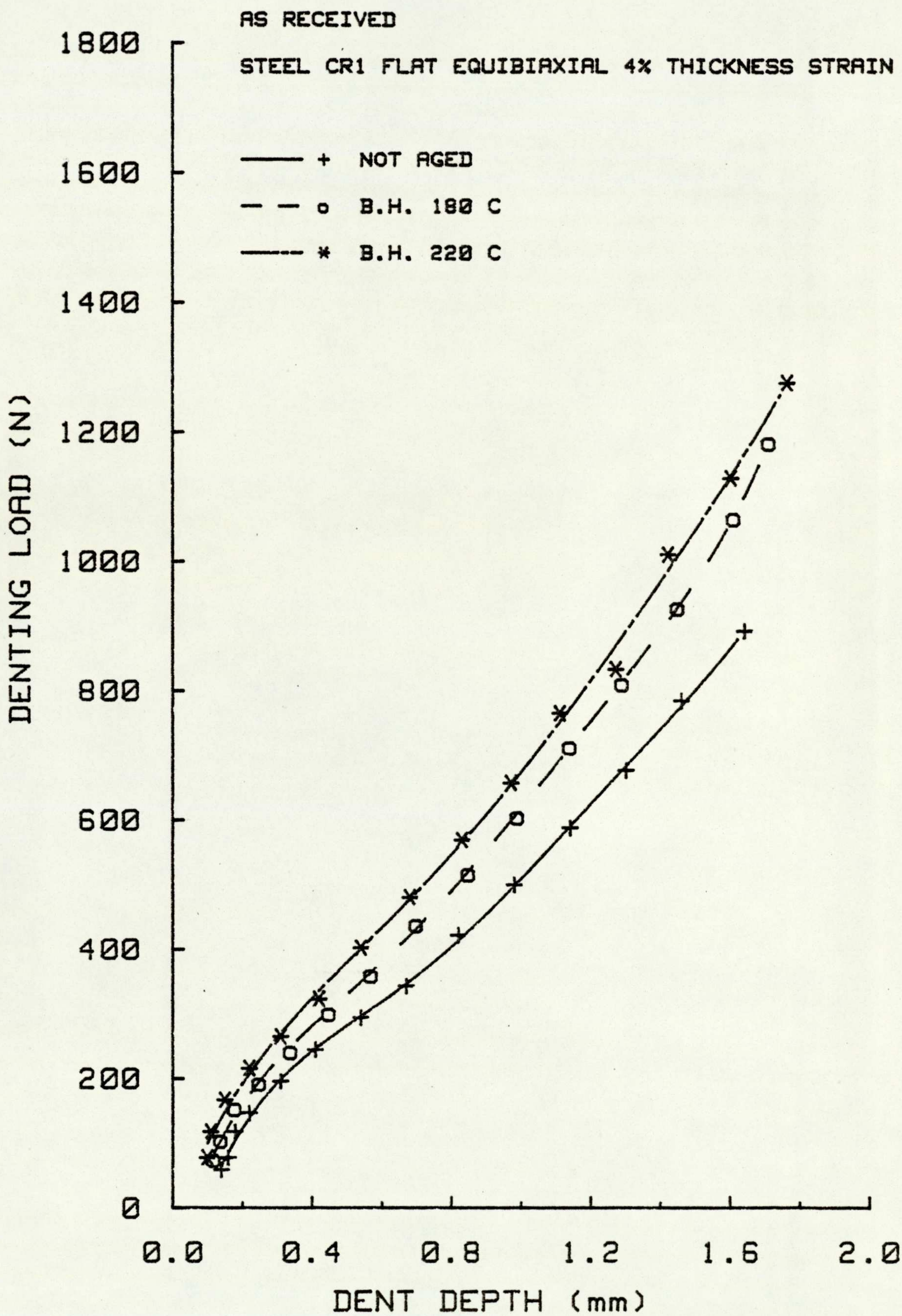


Fig 51. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent-resistance of flat panels as-received steel CR1, stretched to 4% thickness strain in equibiaxial stretching.

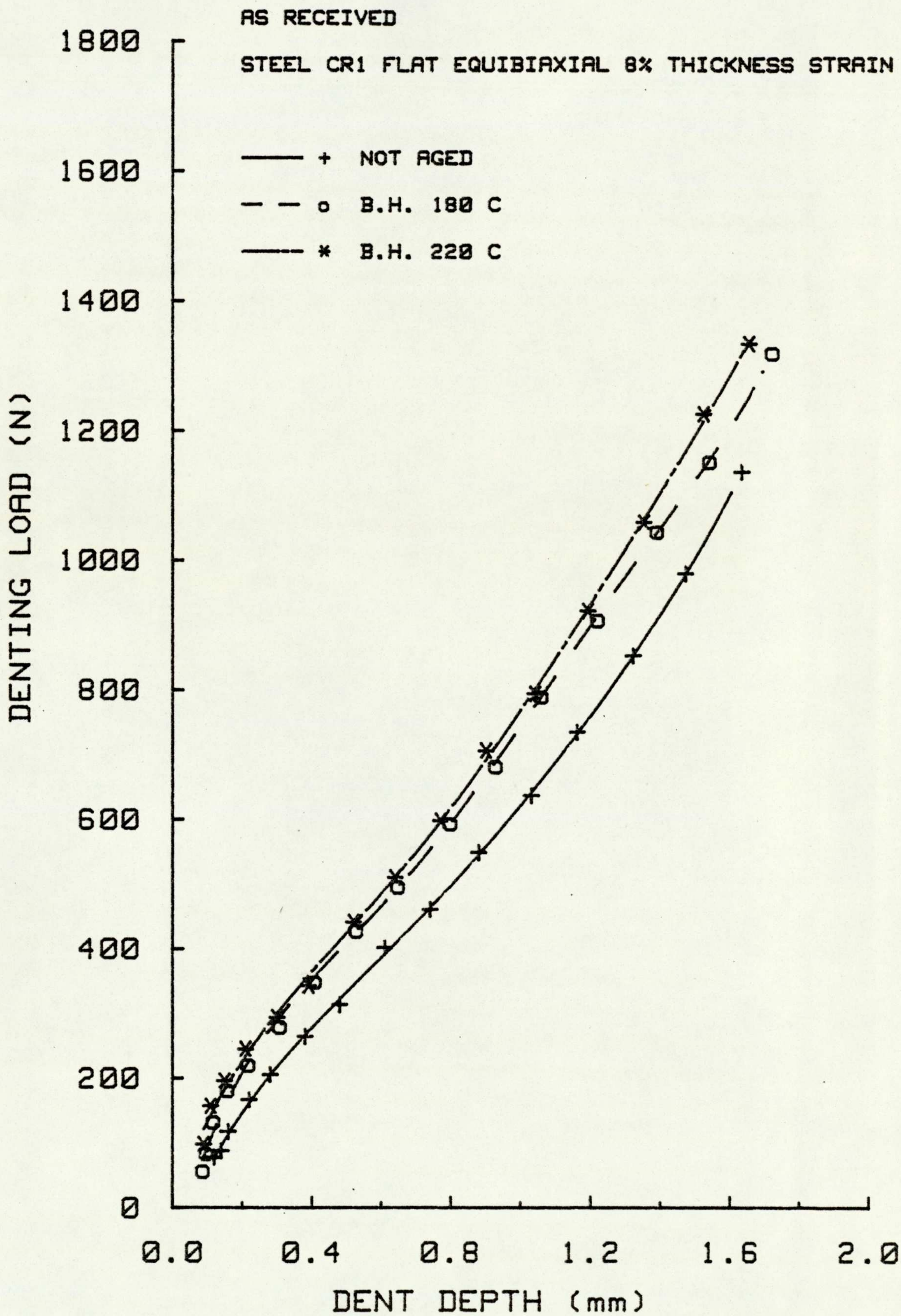


Fig 52. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent-resistance of flat panels as-received steel CR1, stretched to 8% thickness strain in equibiaxial stretching.

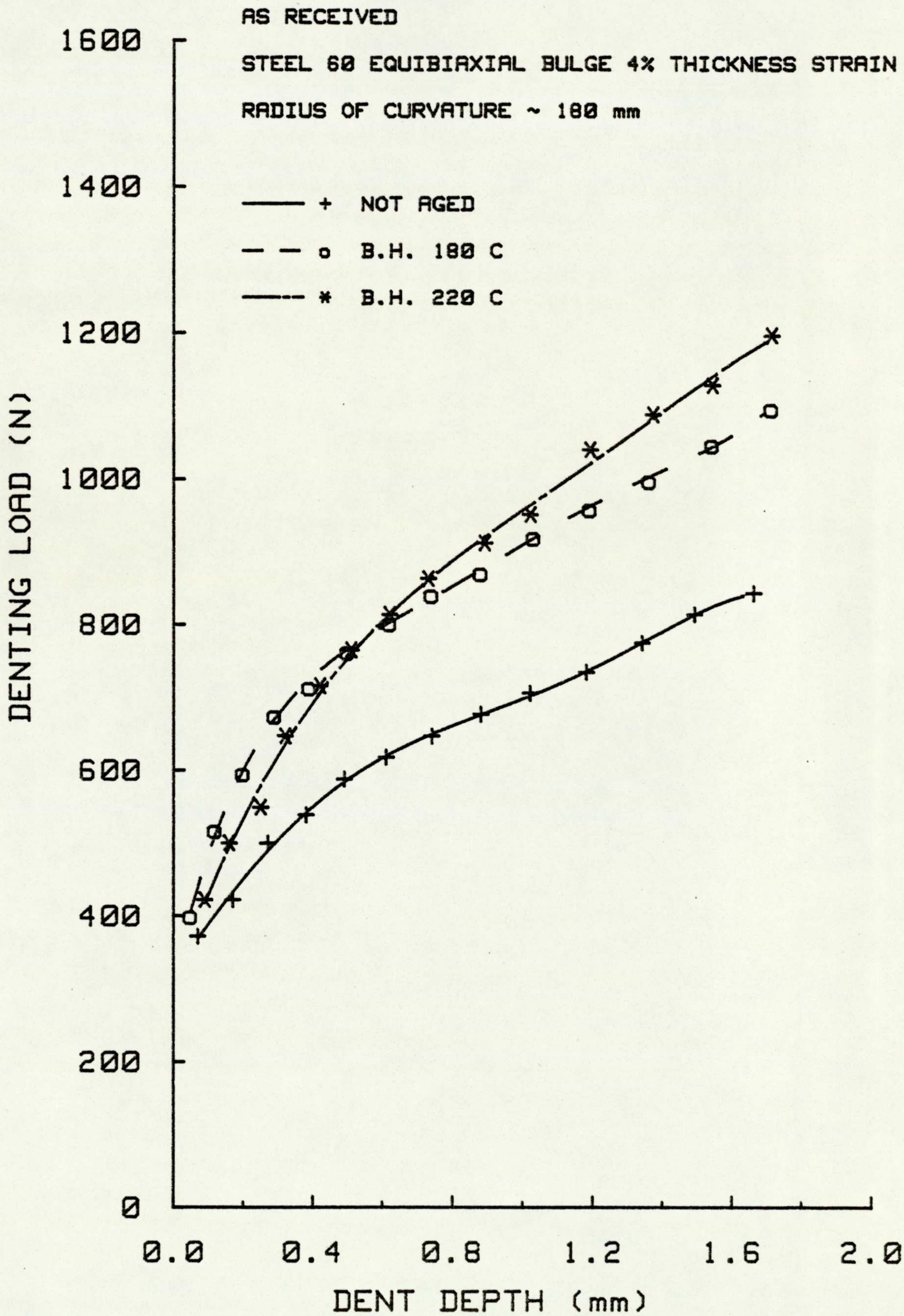


Fig 53. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent resistance of curved panels of as-received steel 60, stretched equibiaxially to 4% thickness strain.

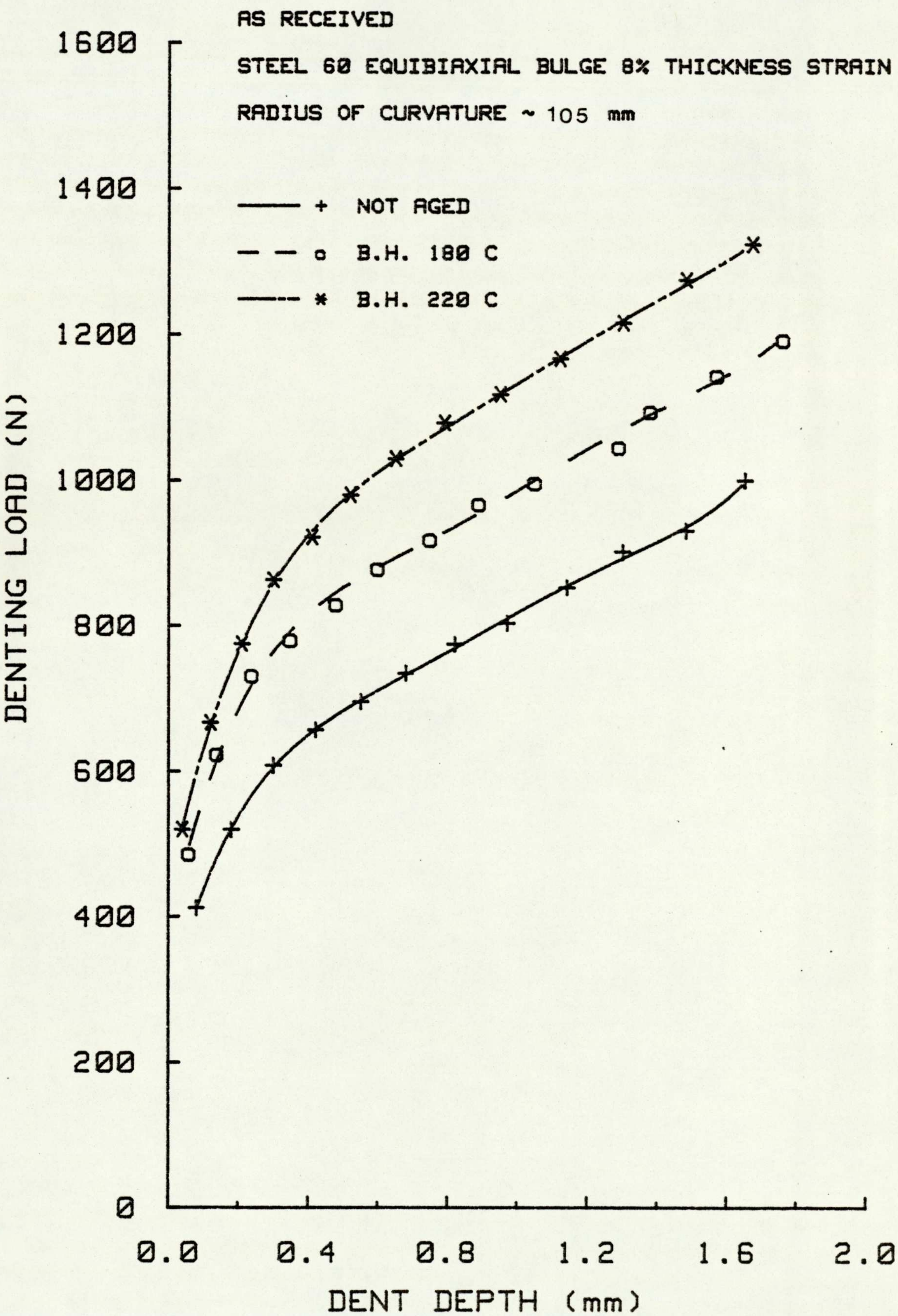


Fig 54. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent resistance of curved panels of as-received steel 60, stretched equibiaxially to 8% thickness strain.

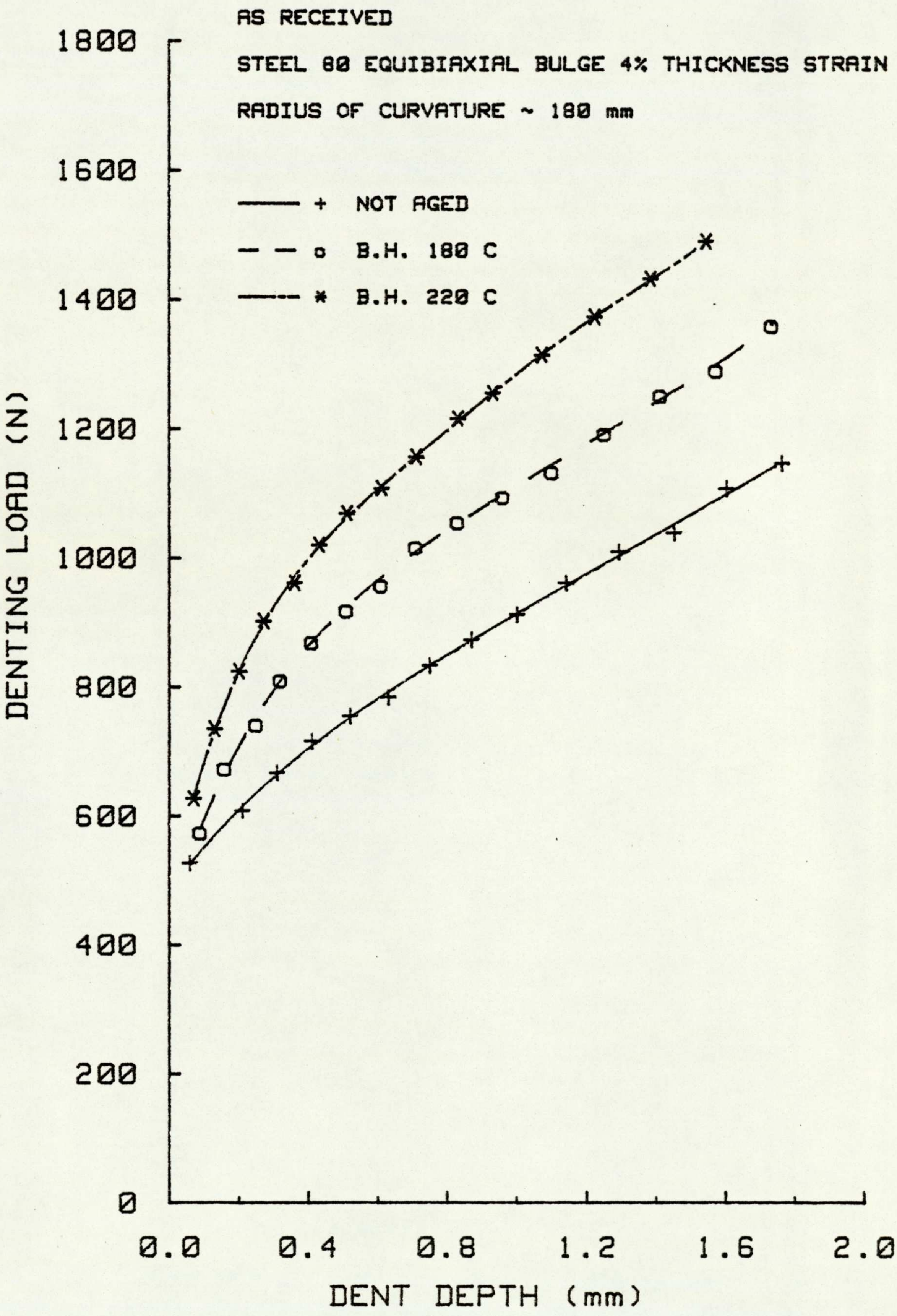


Fig 55. Effect of bake hardening at 180°C and 220°C for 30 minutes on dent resistance of curved panels of as-received steel 80, stretched equibiaxially to 4% thickness strain.

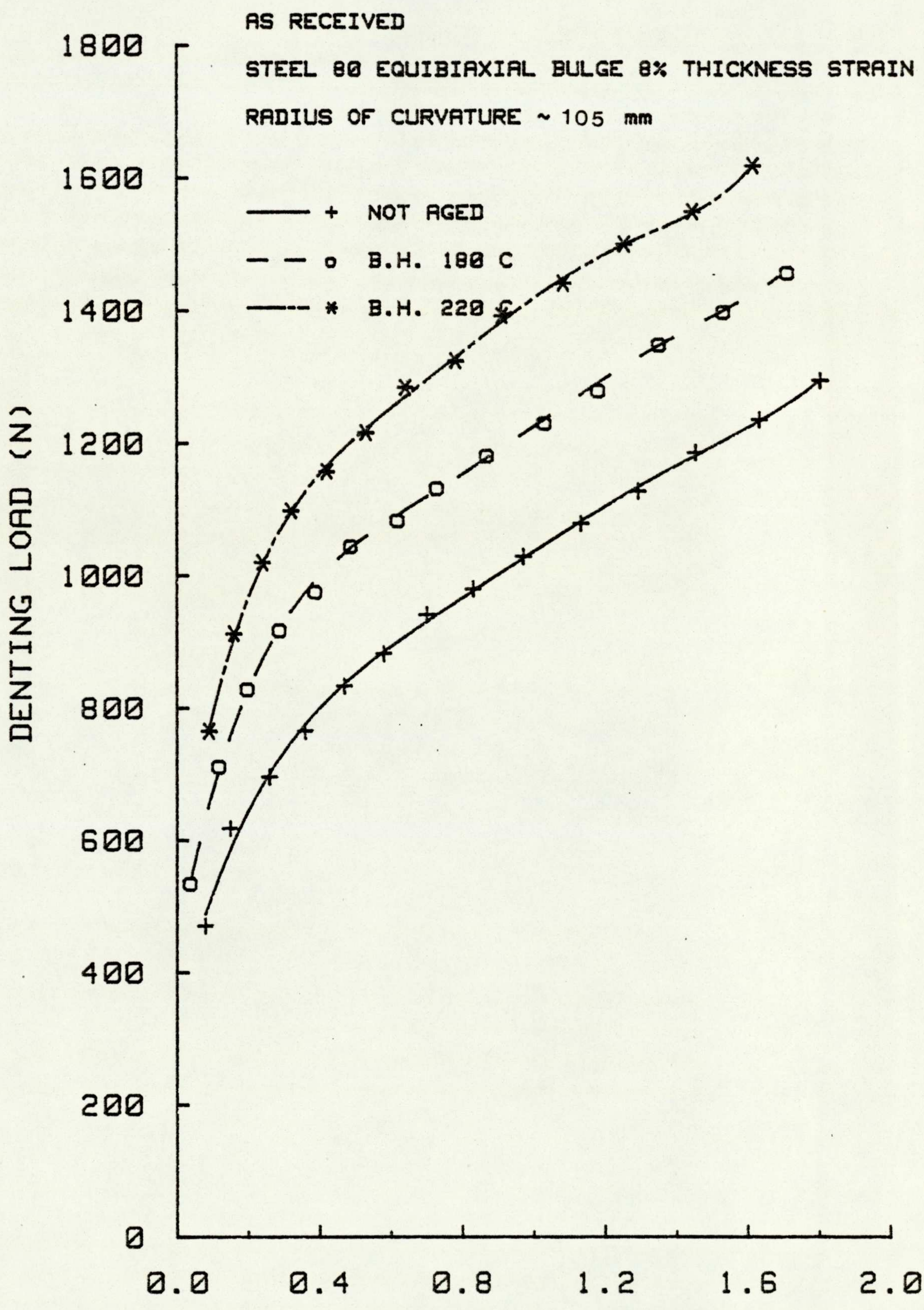


Fig 56. Effect of bake hardening at 180°C and 220°C for 30 minutes on dent resistance of curved panels of as-received steel 80, stretched equibiaxially to 8% thickness strain.

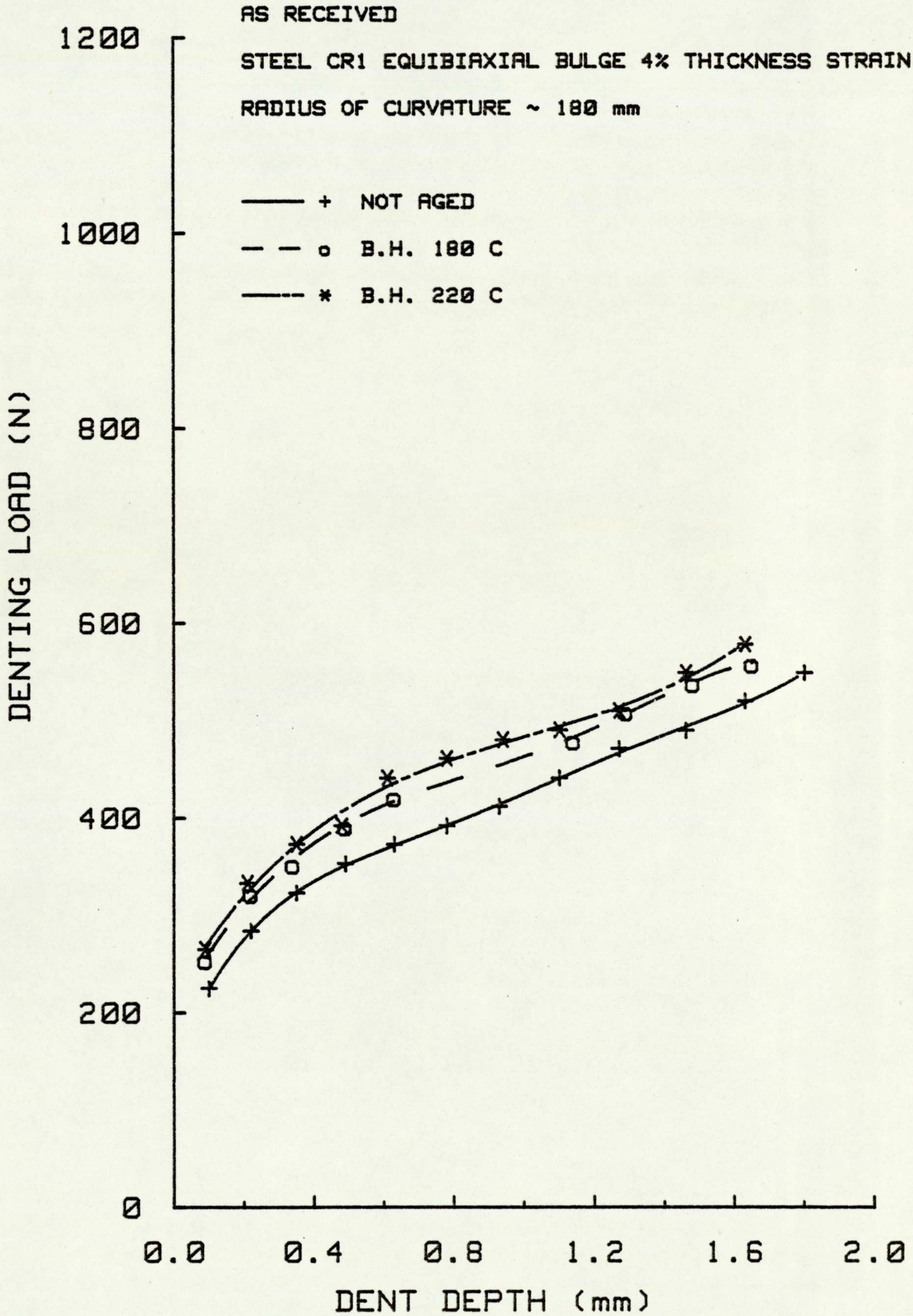


Fig 57. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent resistance of curved panels of as-received steel CR1, stretched equibiaxially to 4% thickness strain.

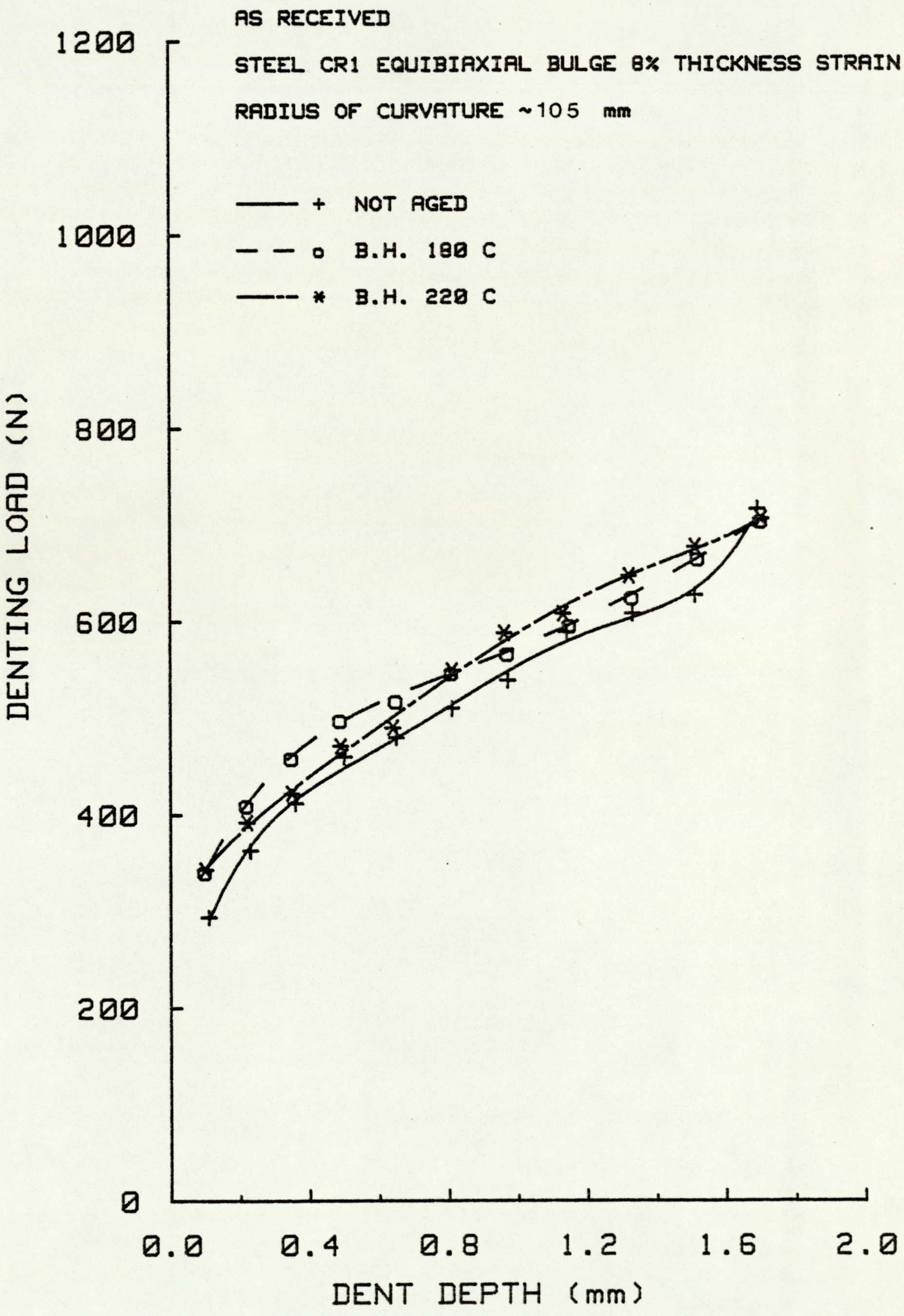


Fig 58. Effect of bake hardening at 180°C and 220°C for 30 minutes on dent resistance of curved panels of as-received steel CR1, stretched equibiaxially to 8% thickness strain.

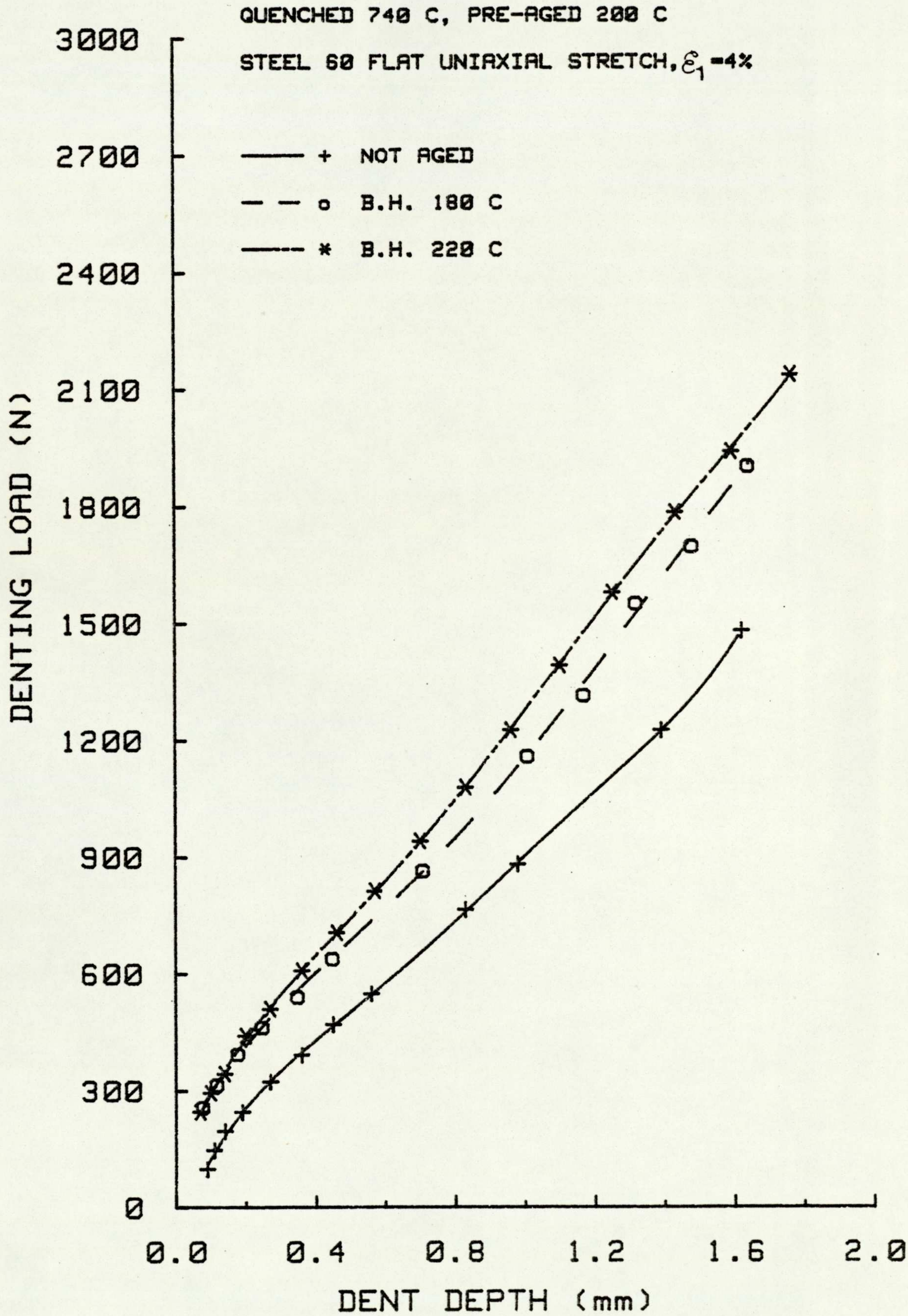


Fig 59. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent-resistance of steel 60, uniaxially stretched 4% after quenching from 740°C and pre-ageing at 200°C for 3 minutes.

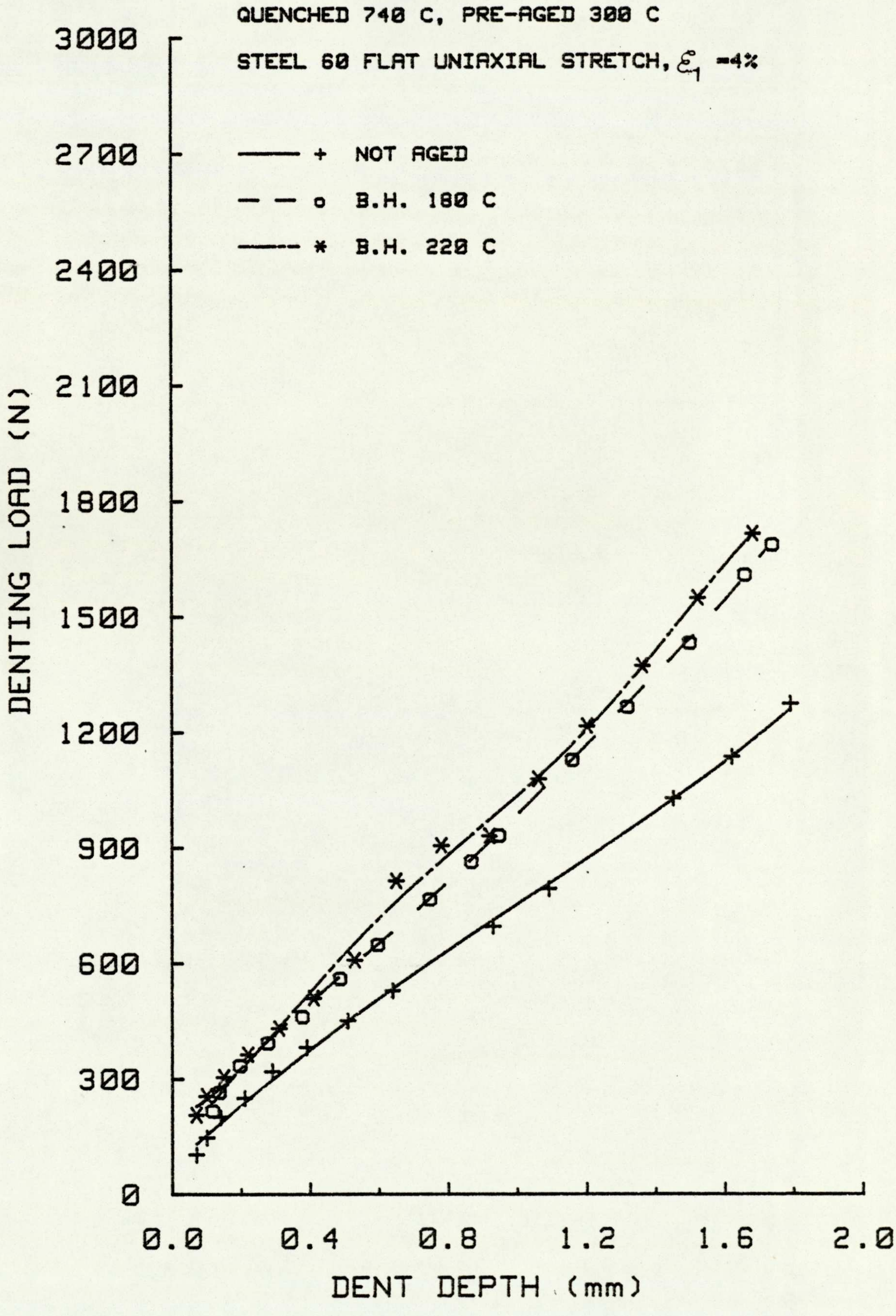


Fig 60. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent-resistance of steel 60, uniaxially stretched 4% after quenching from 740°C and pre-ageing at 300°C for 3 minutes.

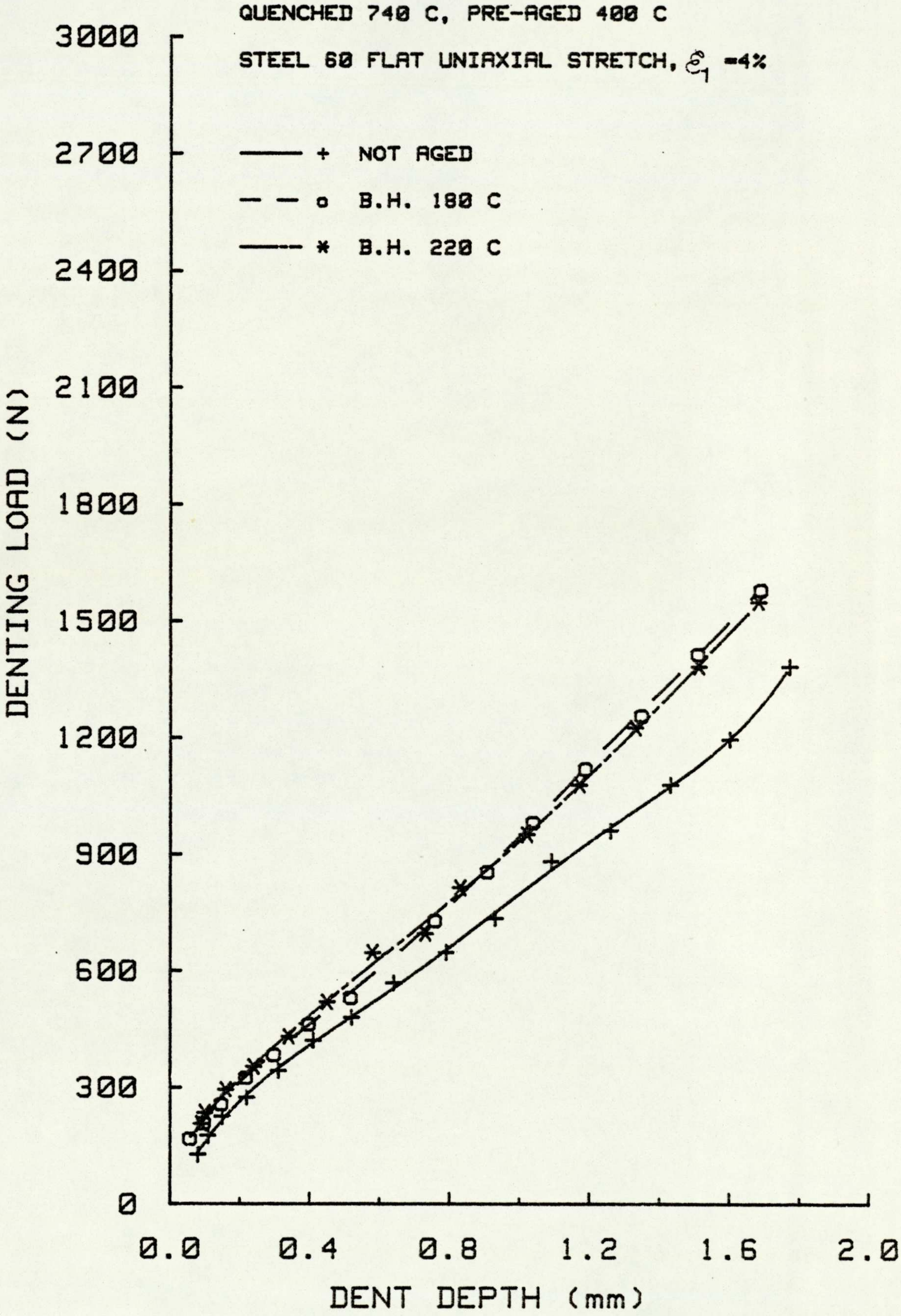


Fig 61. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent-resistance of steel 60, uniaxially stretched 4% after quenching from 740°C and pre-ageing at 400°C for 3 minutes.

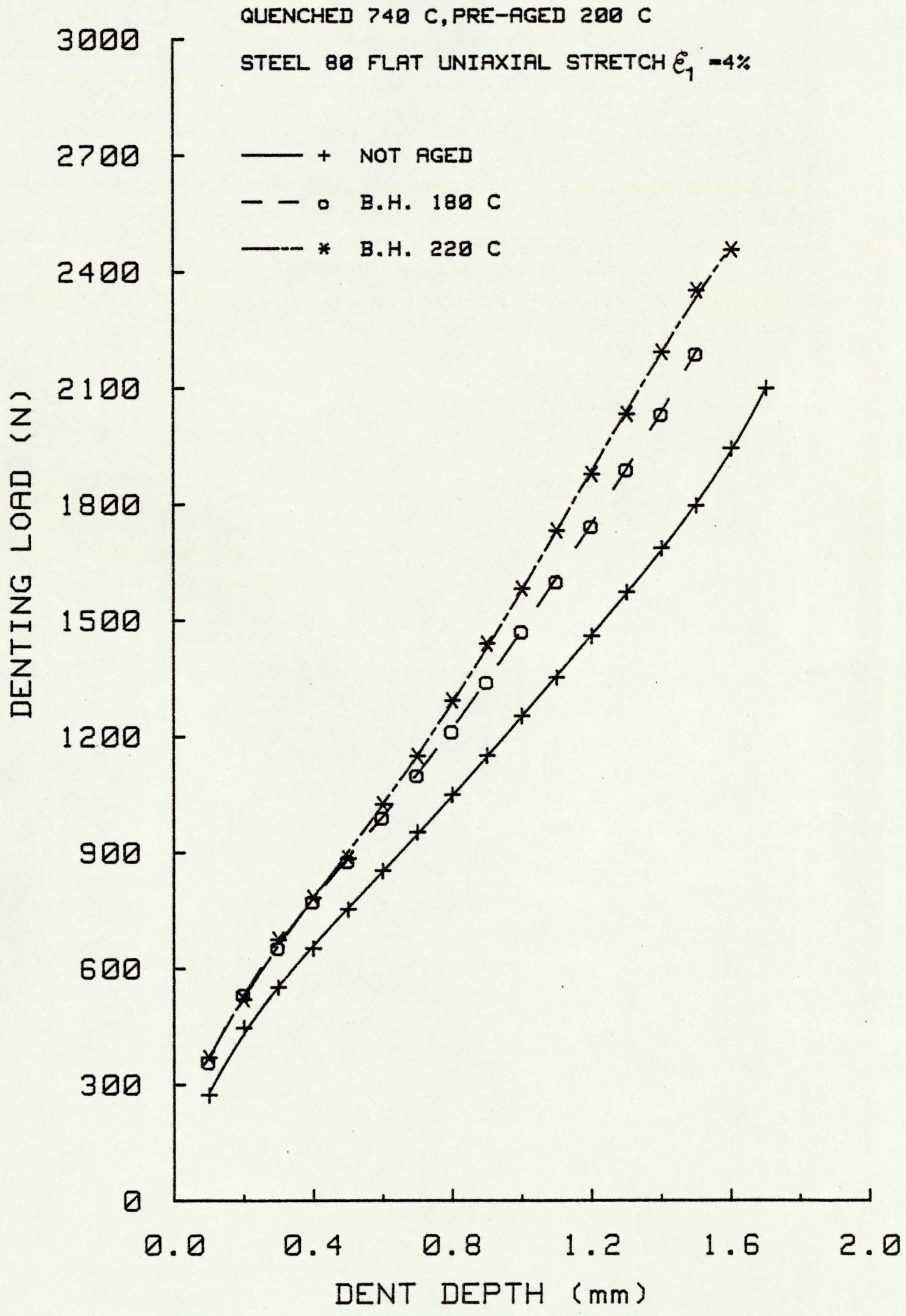


Fig 62. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent-resistance of steel 80, uniaxially stretched 4% after quenching from 740°C and pre-ageing at 200°C for 3 minutes.

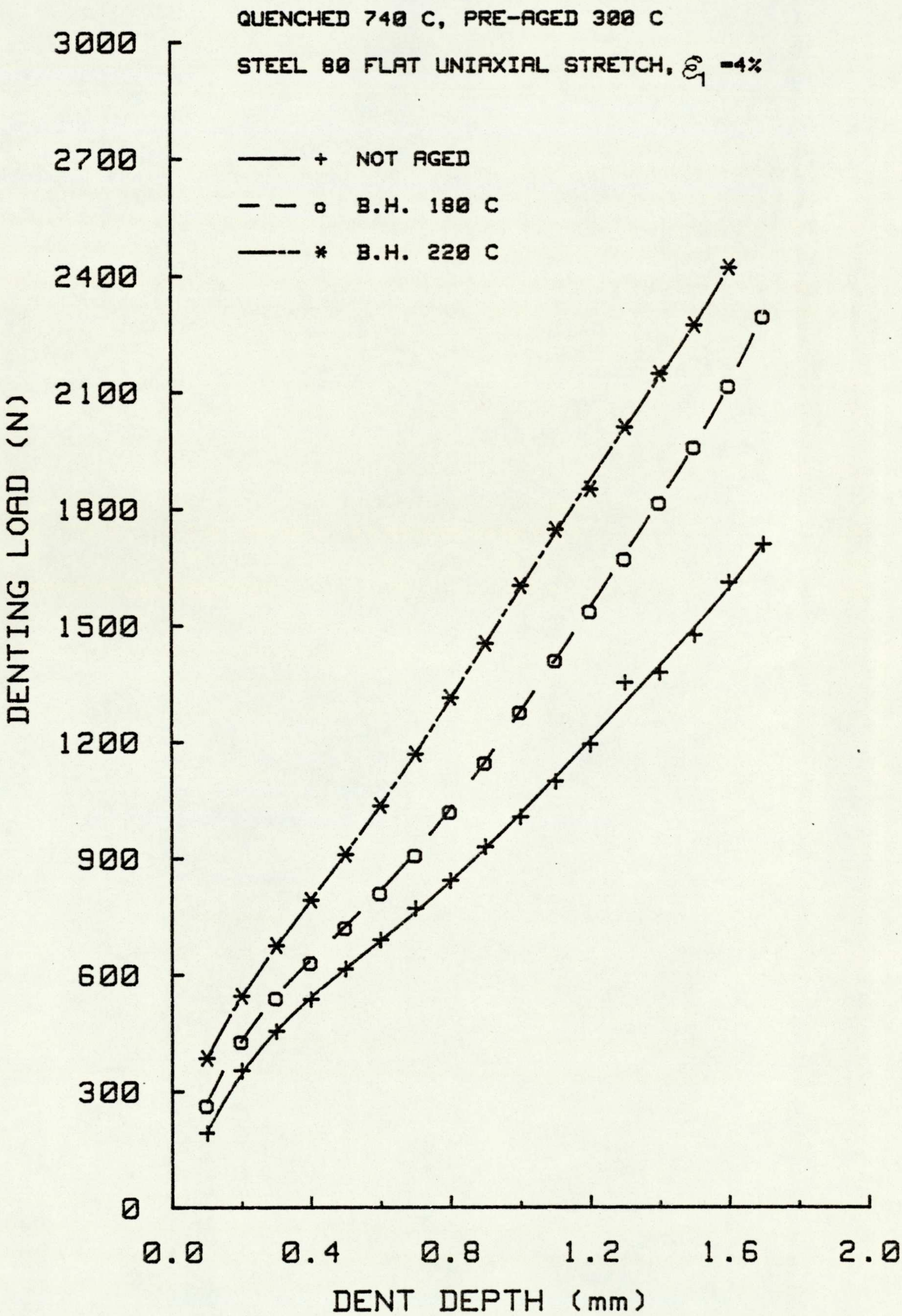


Fig 63. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent-resistance of steel 80, uniaxially stretched 4% after quenching from 740°C and pre-ageing at 300°C for 3 minutes.

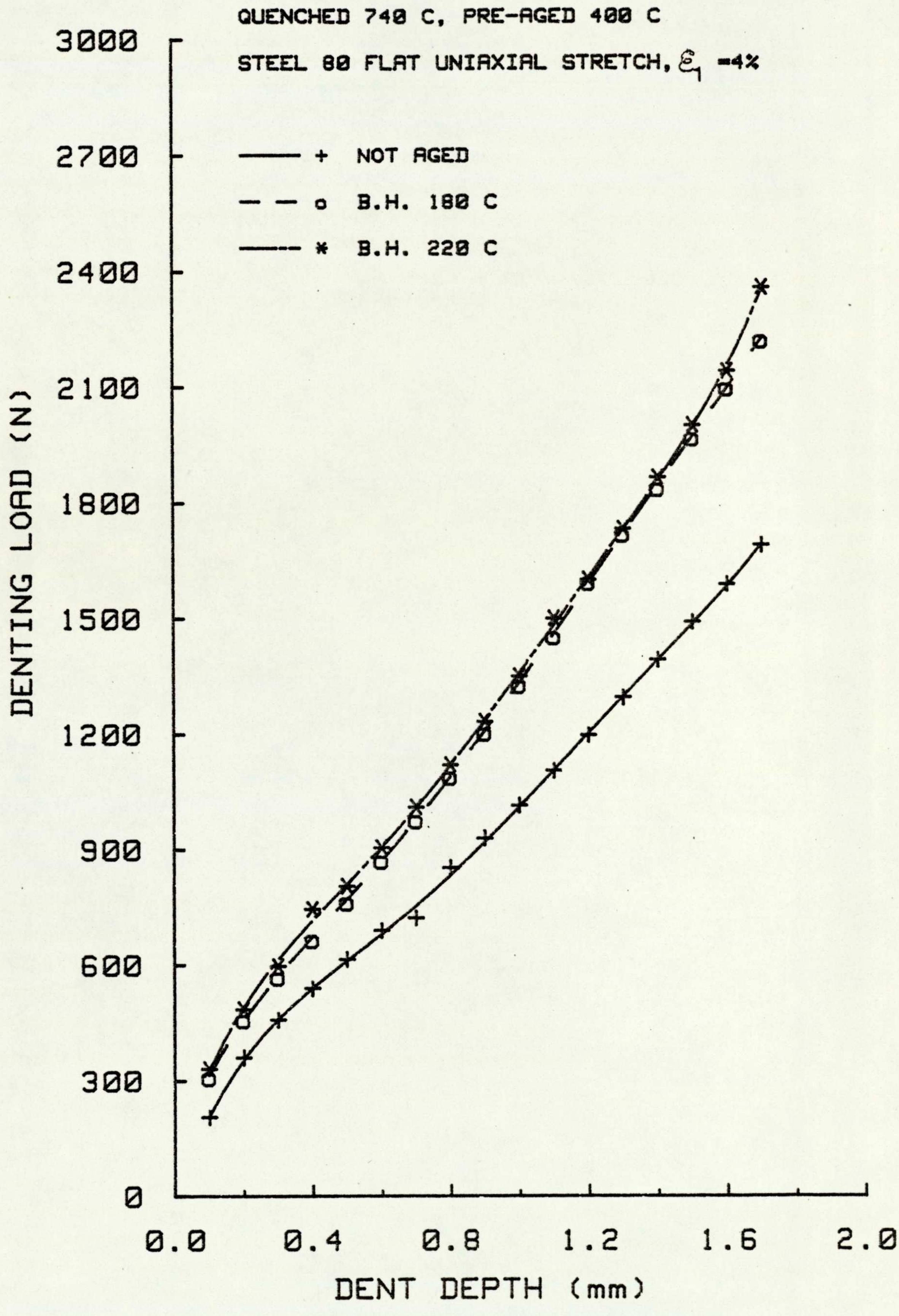


Fig 64. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent-resistance of steel 80, uniaxially stretched 4% after quenching from 740°C and pre-ageing at 400°C for 3 minutes.

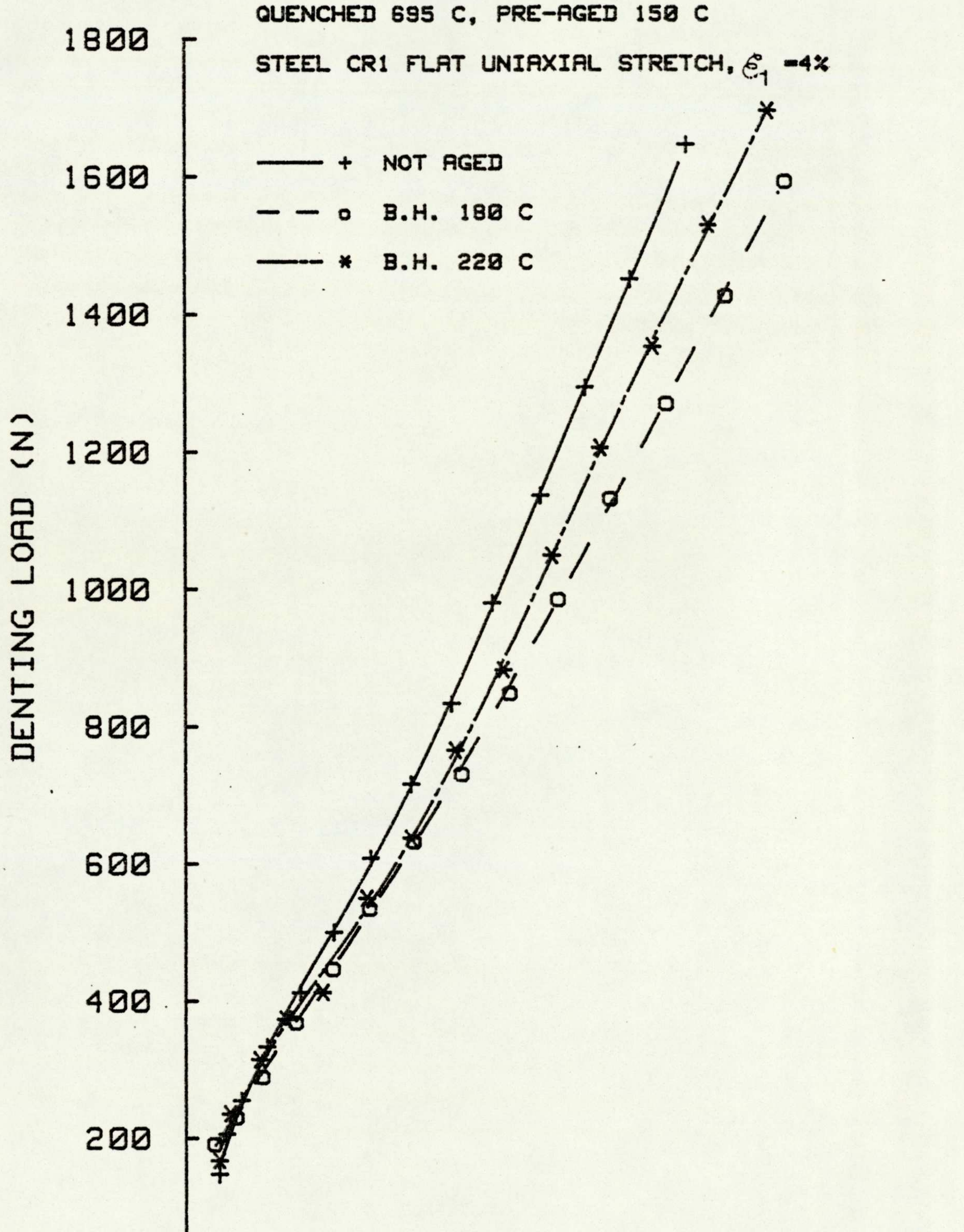


Fig 65. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent-resistance of steel CR1, uniaxially stretched 4% after quenching from 695°C and pre-ageing at 150°C for 3 minutes.

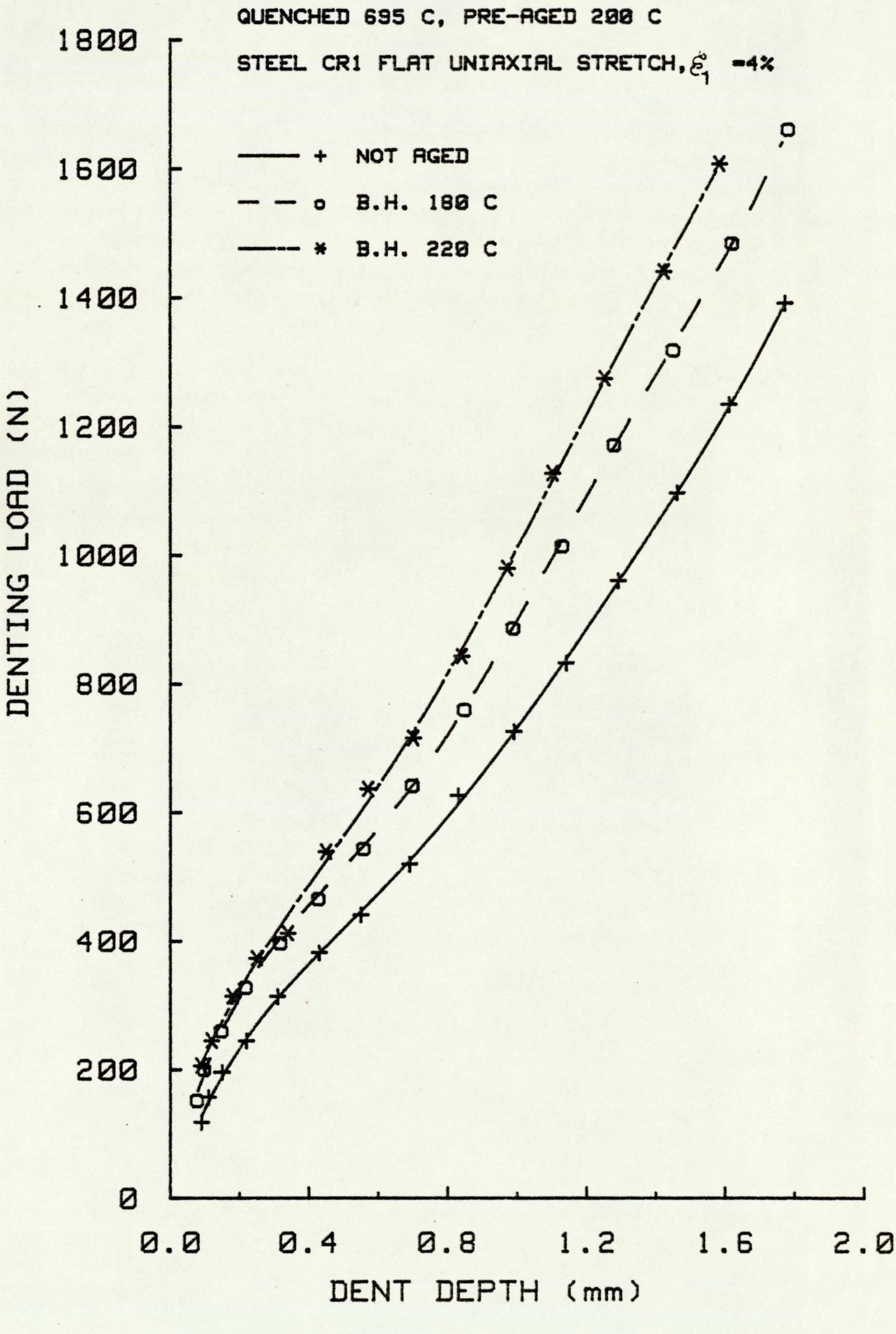


Fig 66. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent-resistance of steel CR1, uniaxially stretched 4% after quenching from 695°C and pre-ageing at 200°C for 3 minutes.

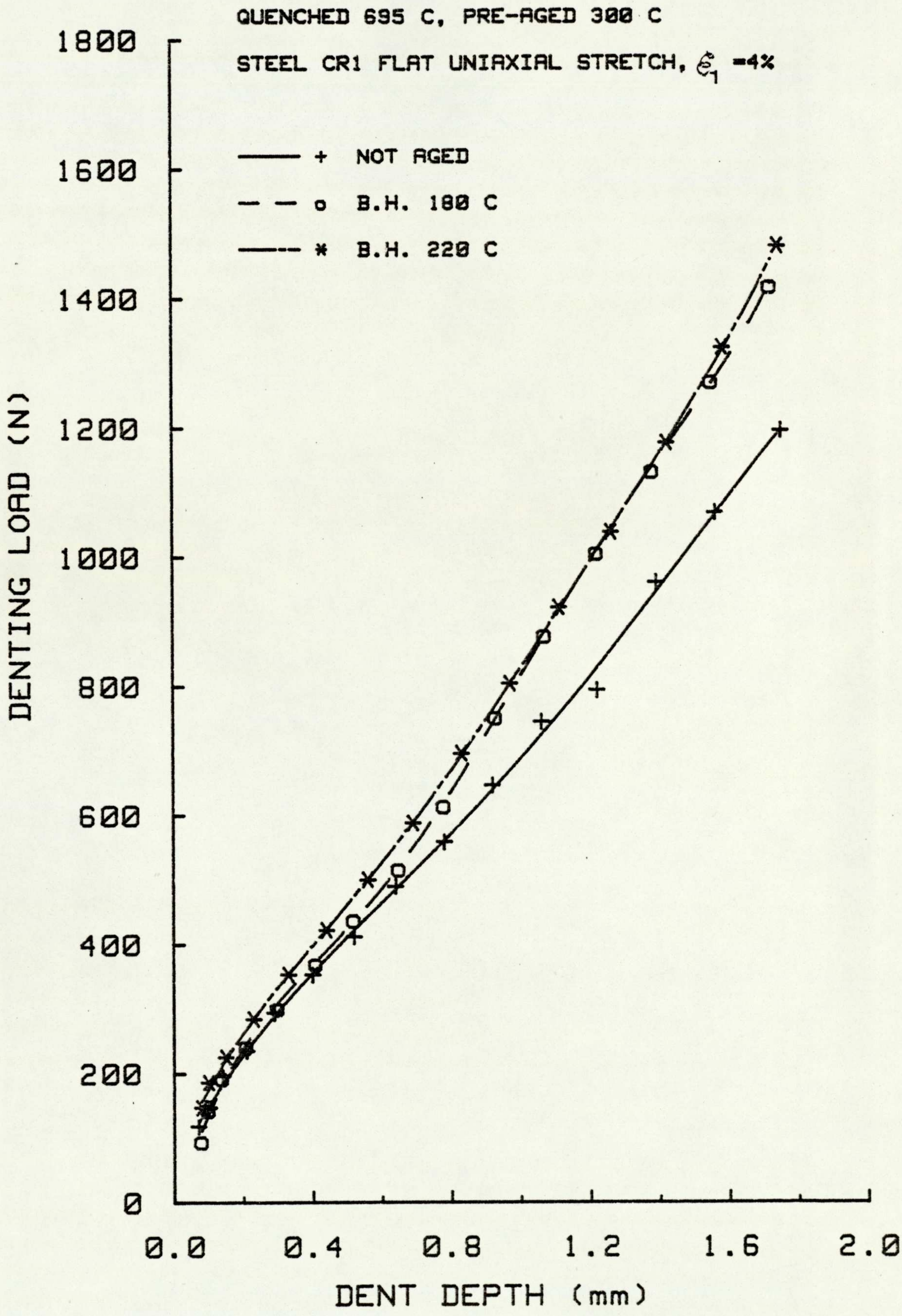


Fig 67. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent-resistance of steel CR1, uniaxially stretched 4% after quenching from 695 °C and pre-ageing at 300°C for 3 minutes.

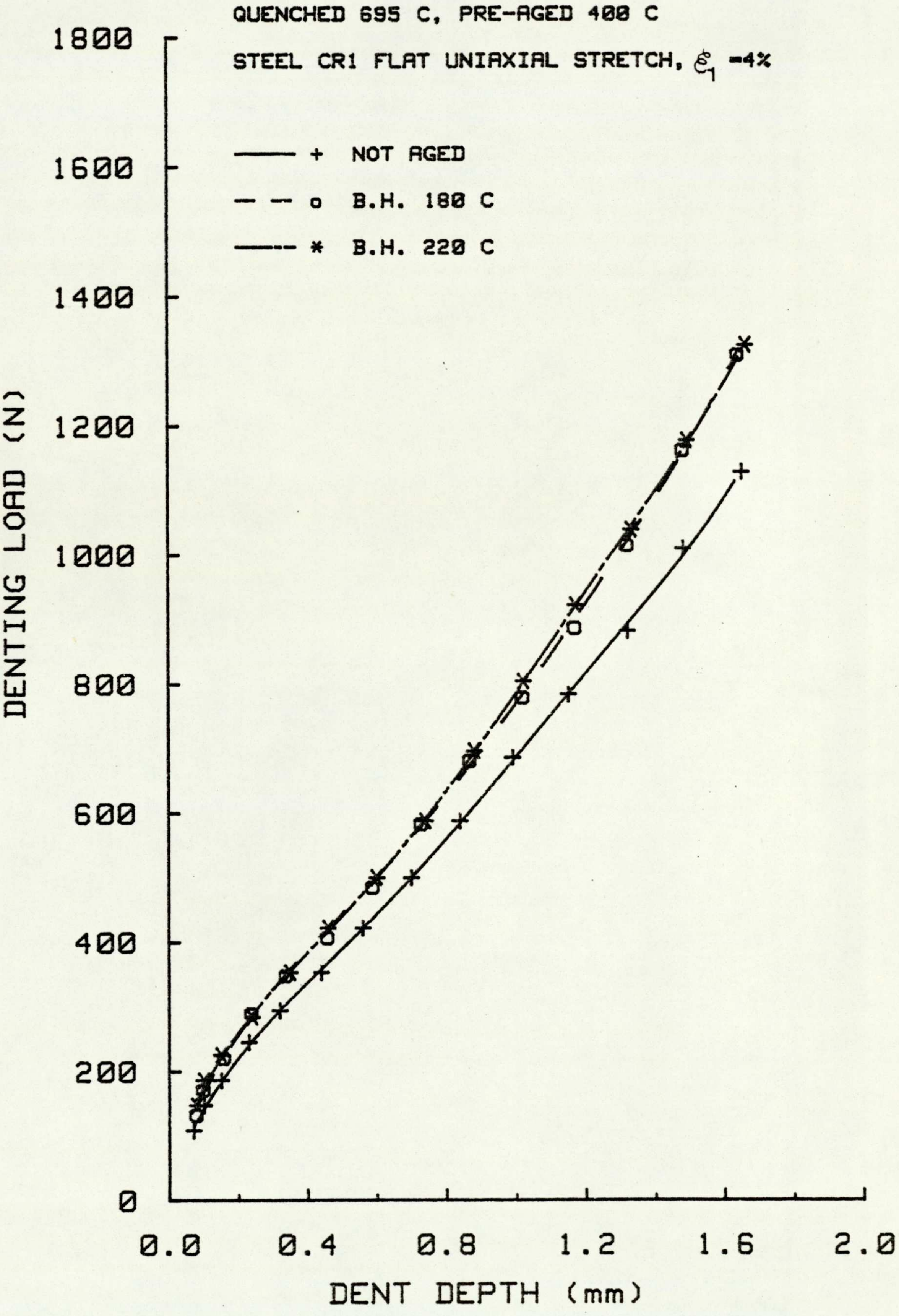


Fig 68. Effect of bake-hardening at 180°C and 220°C for 30 minutes on dent-resistance of steel CR1, uniaxially stretched 4% after quenching from 695°C and pre-ageing at 400°C for 3 minutes.

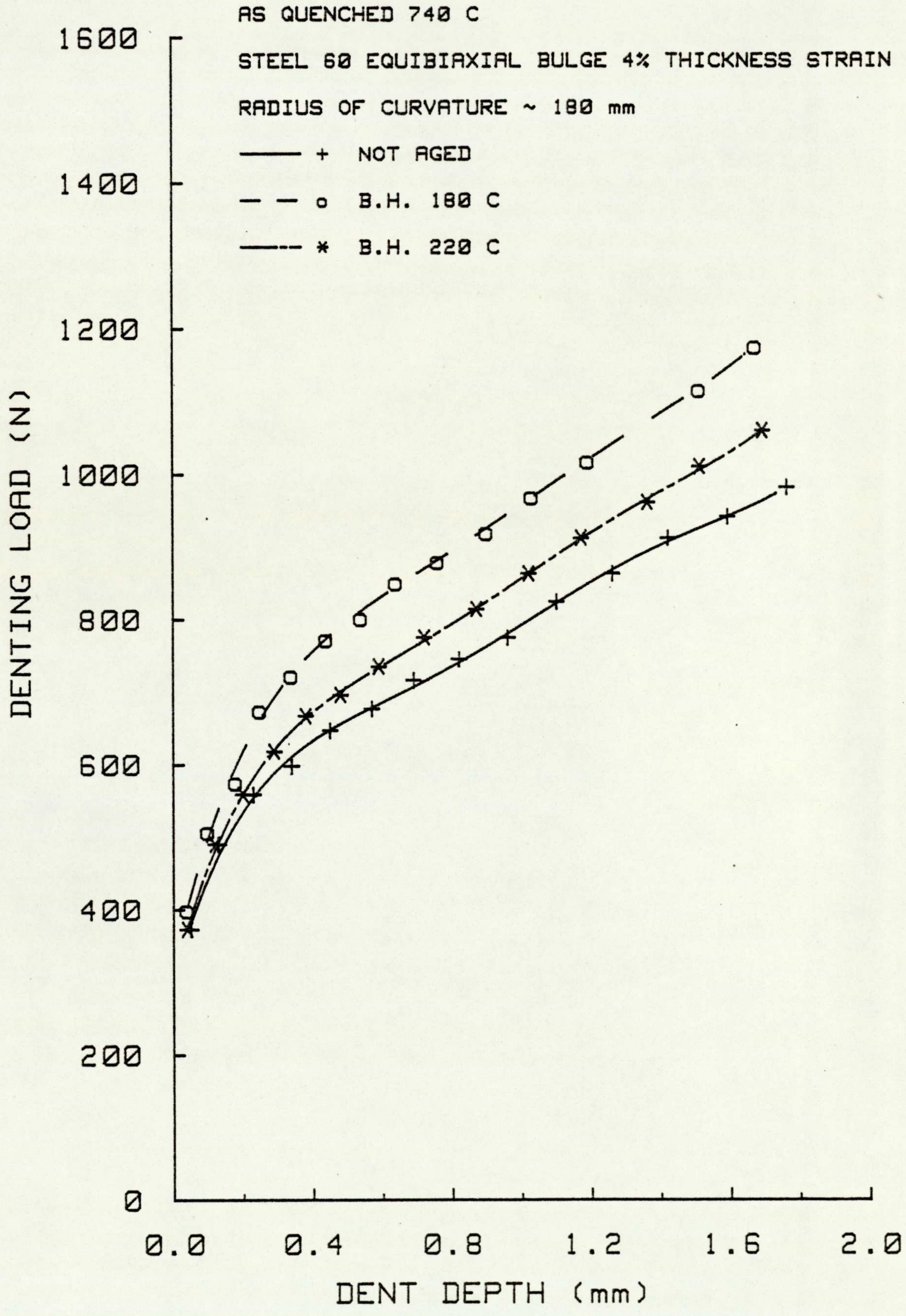


Fig 69. Effect of bake-hardening at 180°C and 220°C on dent-resistance of curved panels of steel 60, quenched from 740°C and then bulged equibiaxially to 4% thickness strain before bake-hardening.

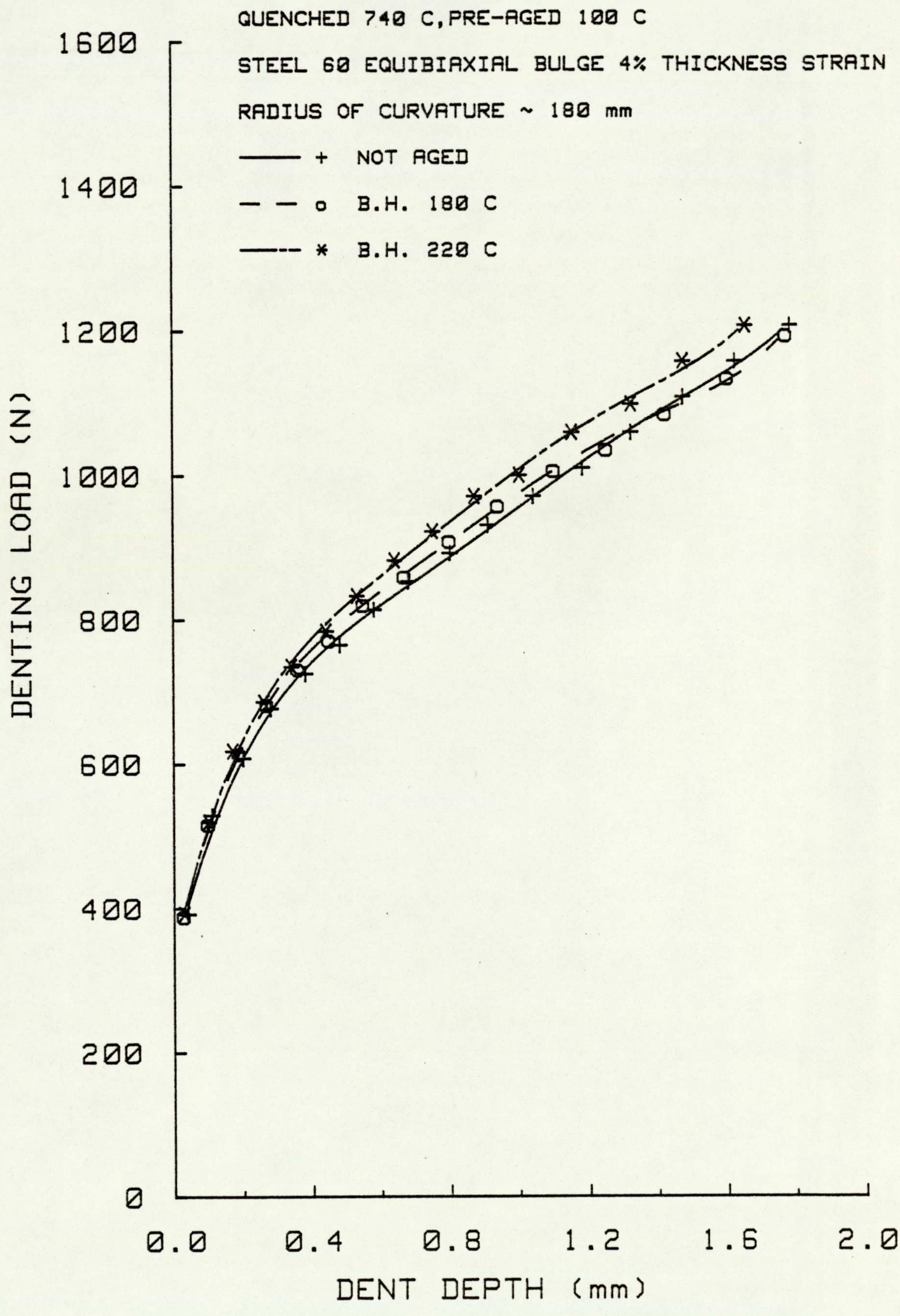


Fig 70. Effect of bake-hardening at 180°C and 220°C on dent-resistance of curved panels of steel 60, quenched from 740°C and pre-aged at 100°C before bulging equibiaxially to 4% thickness strain and then bake-hardening.

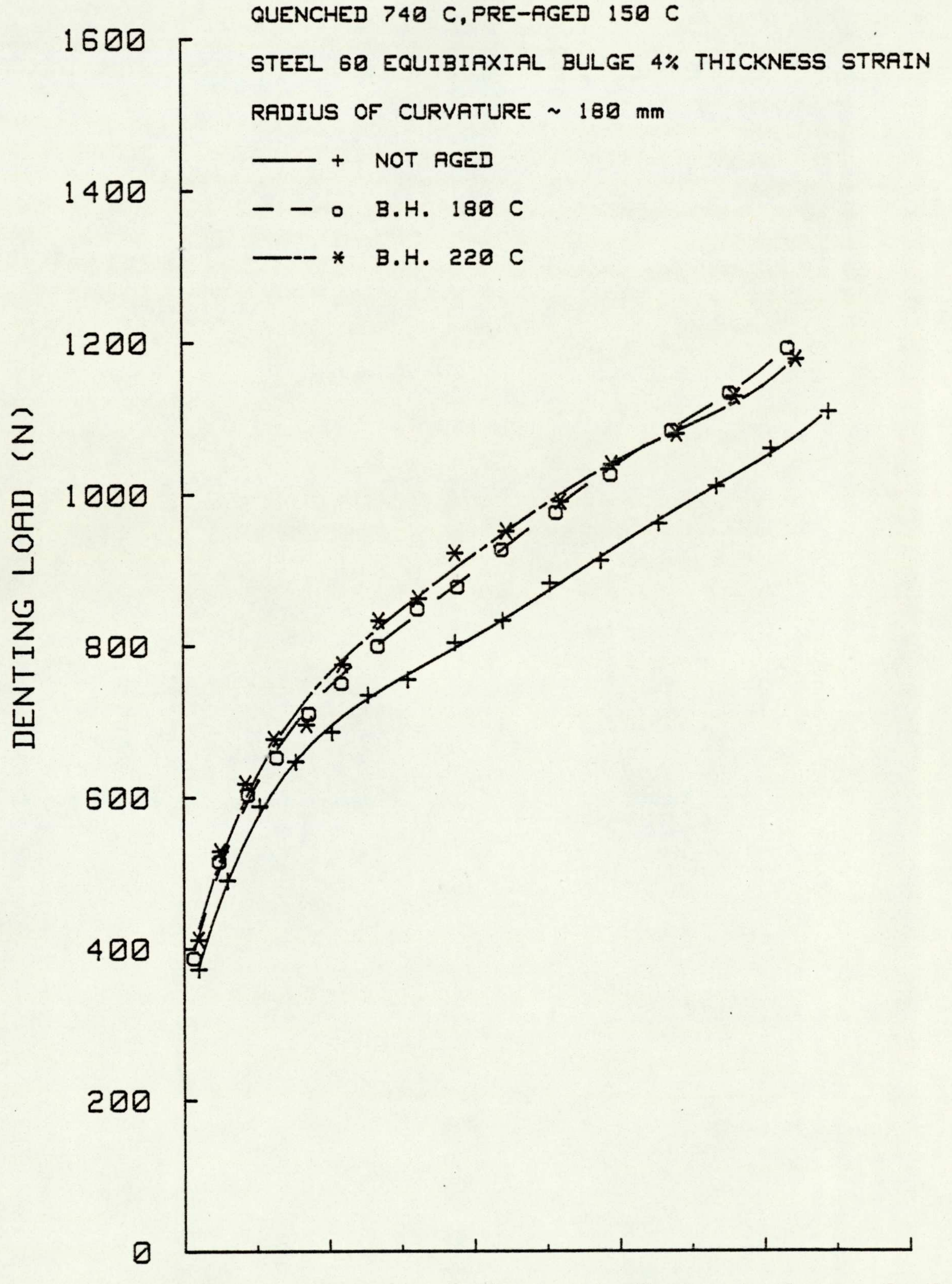


Fig 71. Effect of bake-hardening at 180°C and 220° C on dent-resistance of curved panels of steel 60, quenced from 740°C and preaged at 150°C bulging equibiaxially to 4% thickness strain and then bake-hardening.

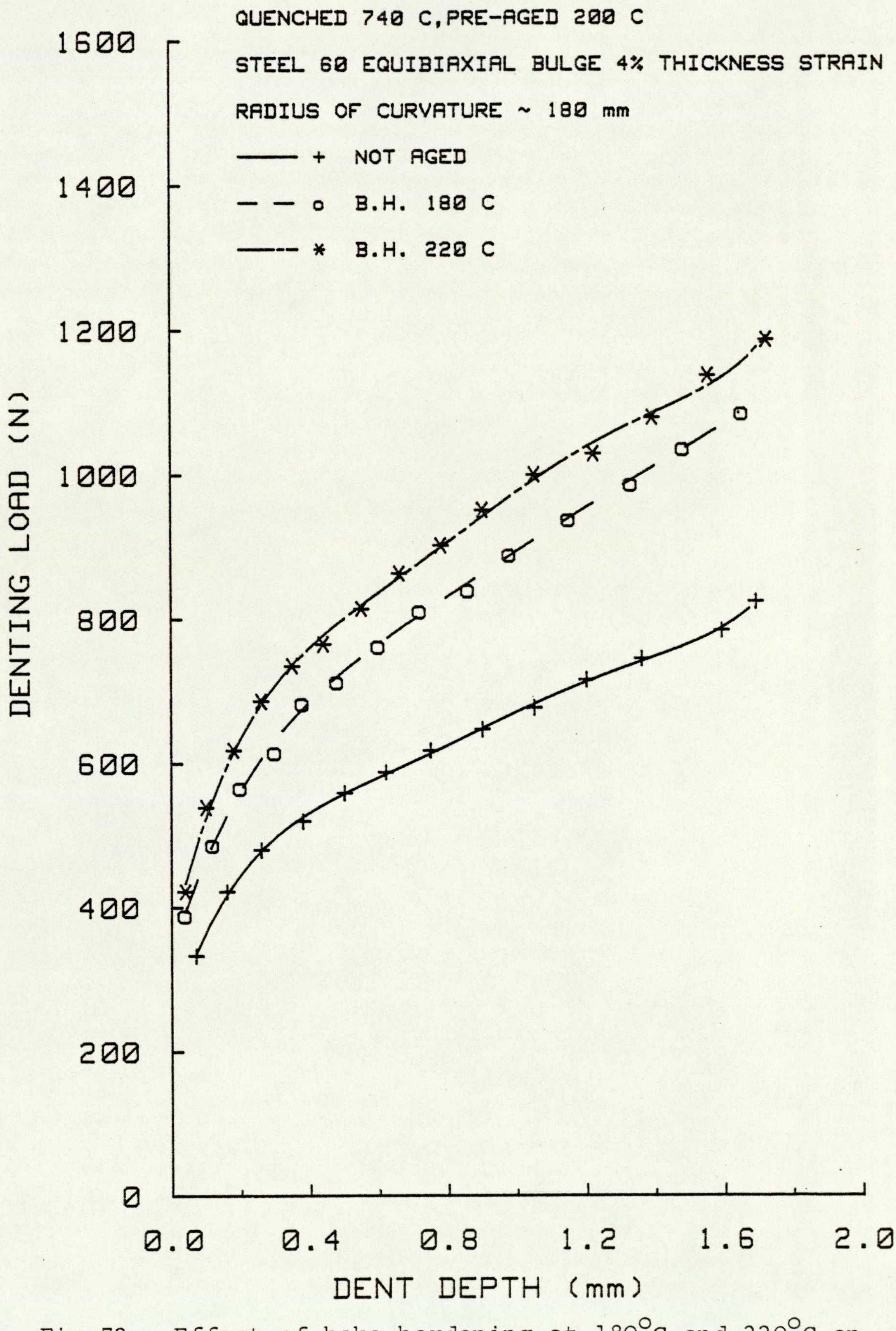


Fig 72. Effect of bake-hardening at 180°C and 220°C on dent-resistance of curved panels of steel 60, quenched from 740°C and preaged at 200°C before bulging equibiaxially to 4% thichness strain and then bake-hardening.

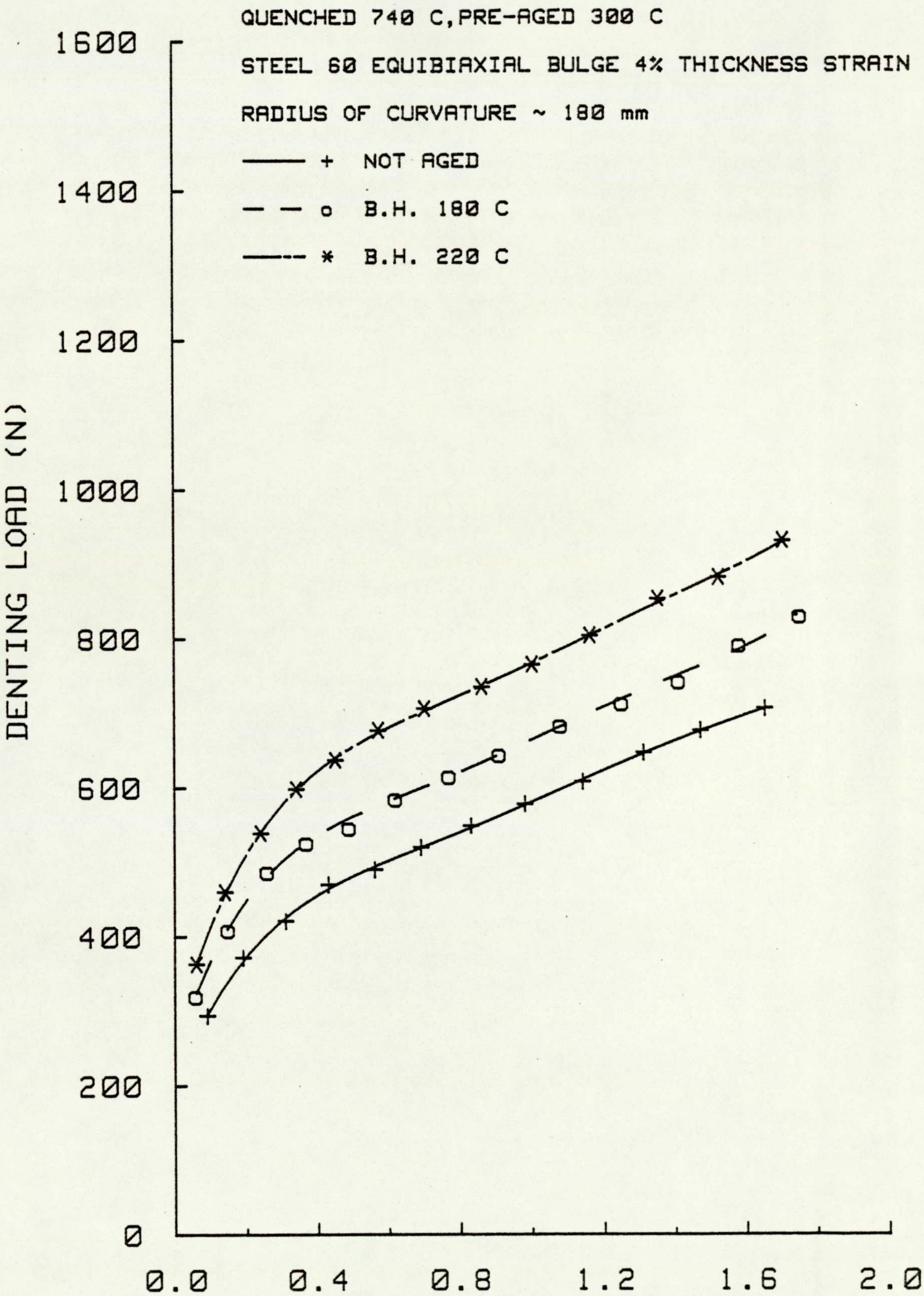


Fig 73. Effect of bake-hardening at 180°C and 220°C on dent-resistance of curved panels of steel 60, quenched from 740°C and preaged at 300°C before bulging equibiaxially to 4% thickness strain and then bake-hardening.

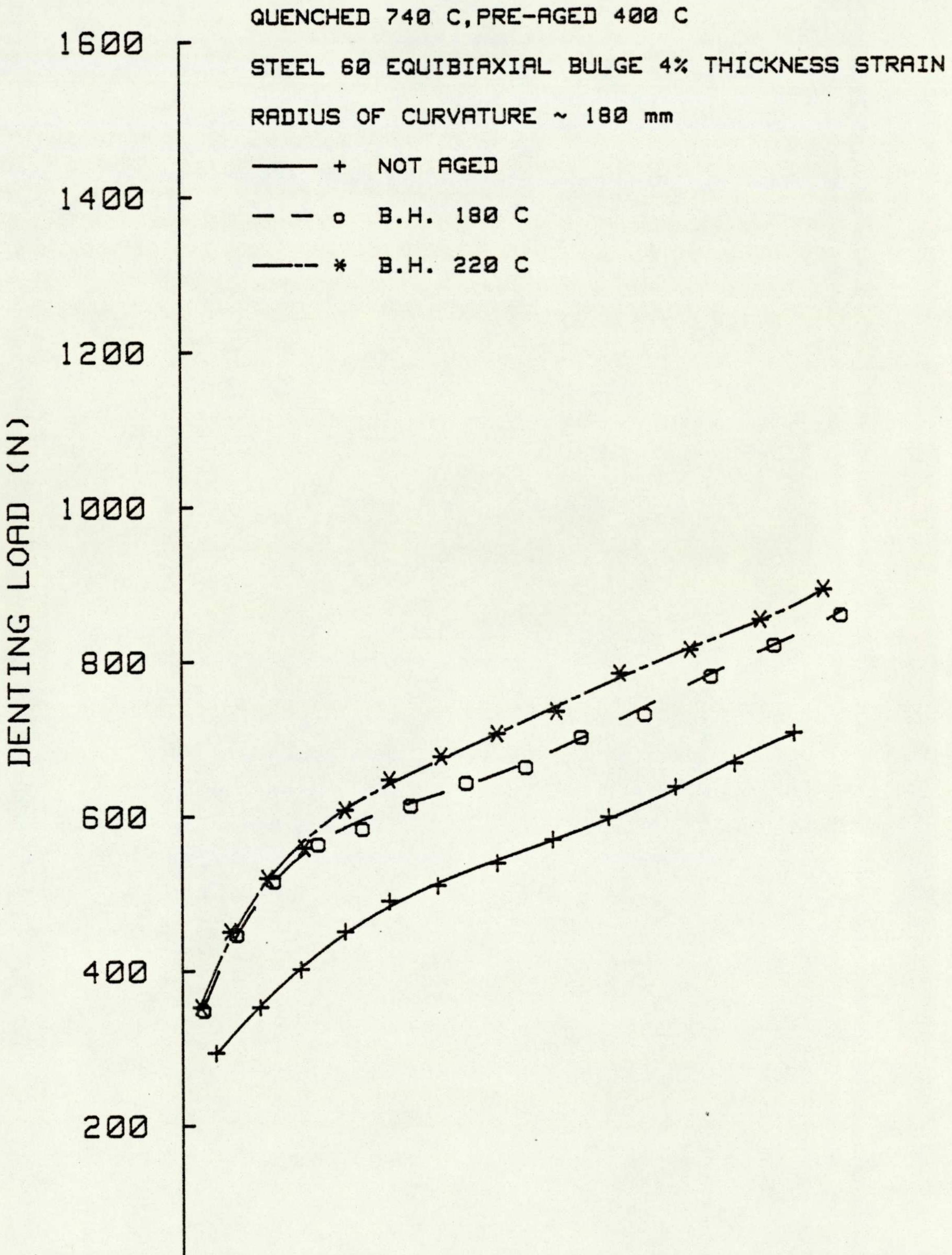


Fig 74. Effect of bake-hardening at 180°C and 220°C on dent-resistance of curved panels of steel 60, quenched from 740°C and pre-aged at 400°C before bulging equibiaxially to 4% thickness strain and then bake-hardening.

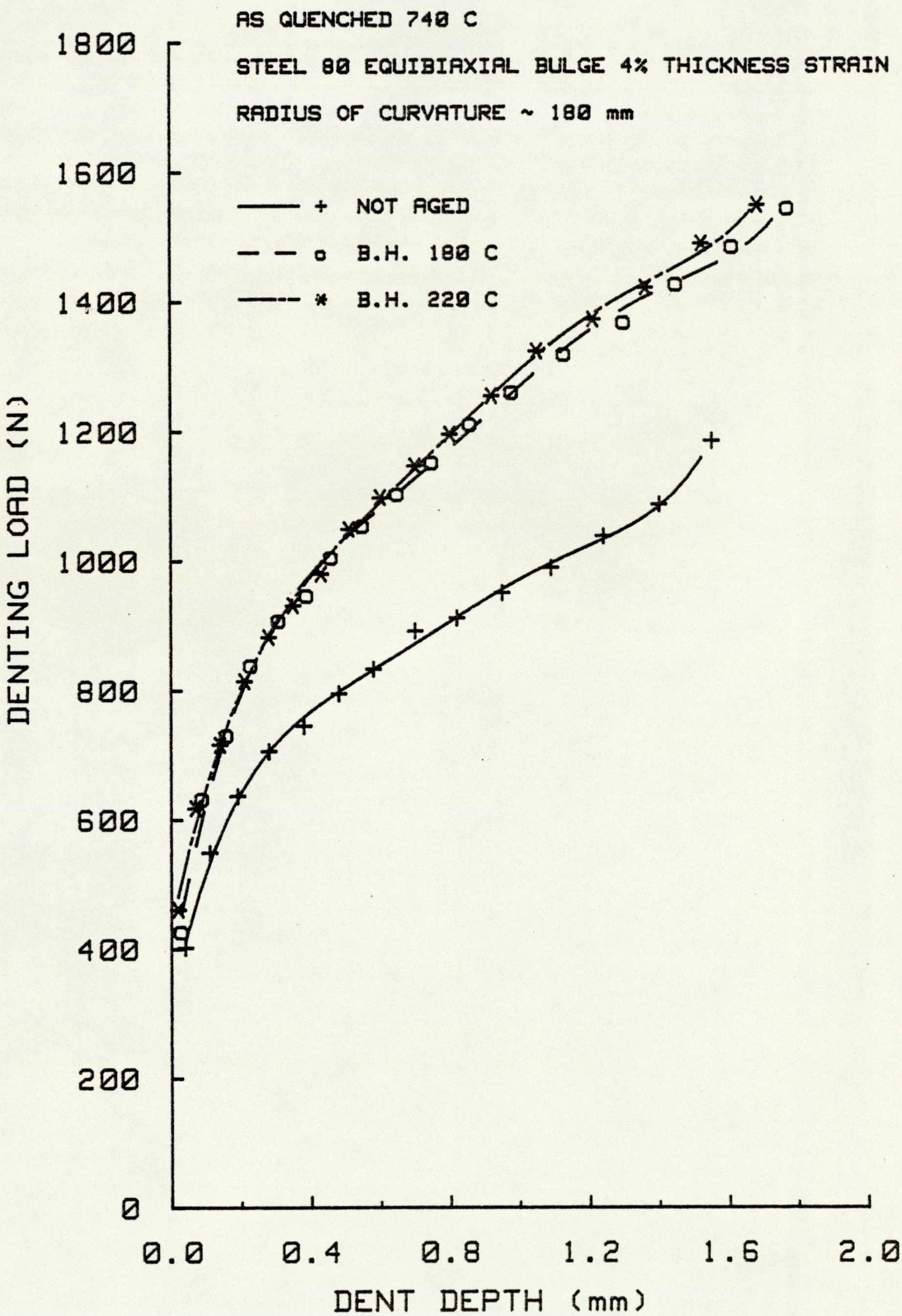


Fig 75. Effect of bake-hardening at 180°C and 220°C on dent-resistance of curved panels of steel 80, quenched from 740°C and then bulged equibiaxially to 4% thickness strain before bake-hardening.

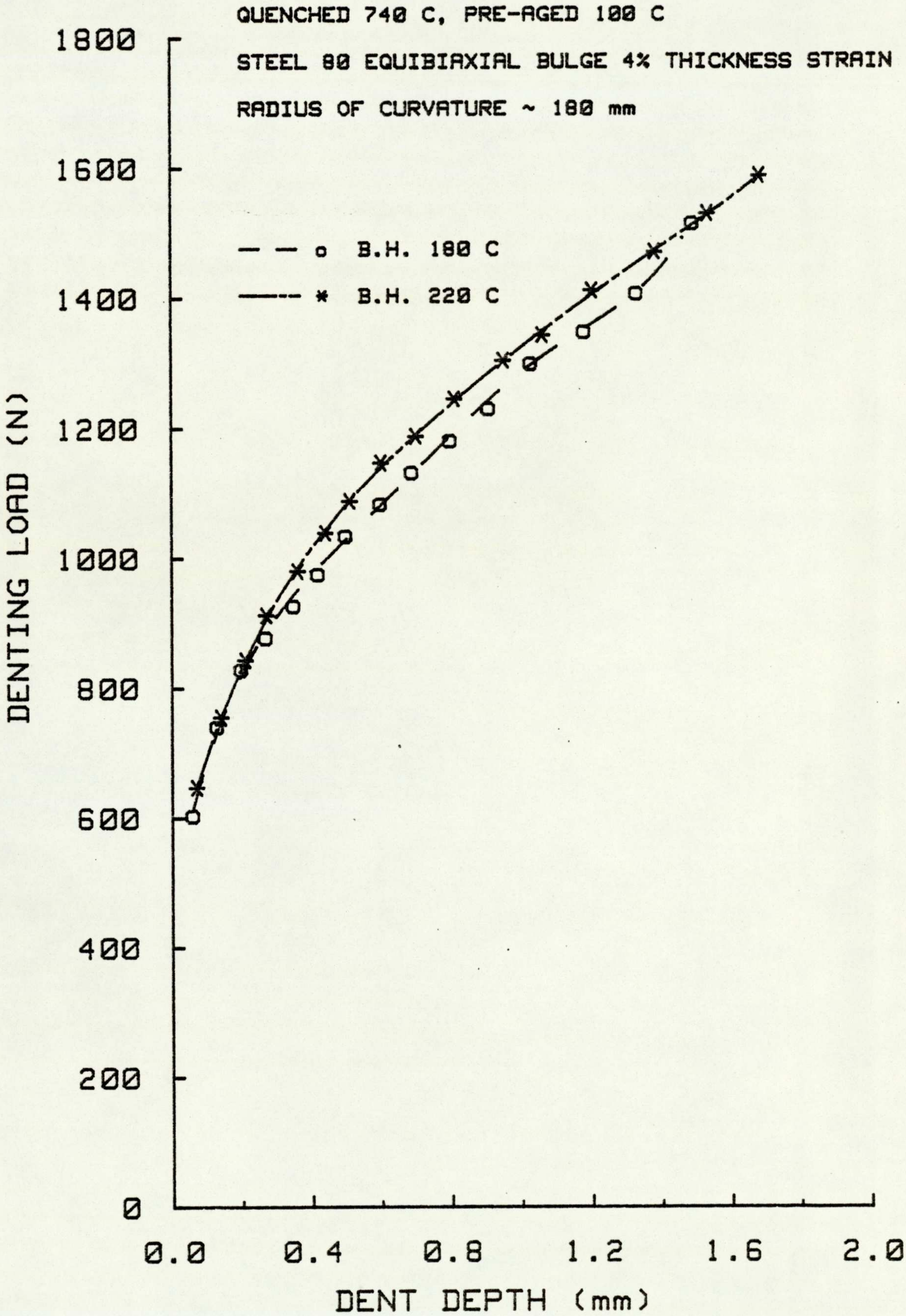


Fig 76. Effect of bake-hardening at 180°C and 220°C on dent-resistance of curved panels of steel 80, quenched from 740°C and preaged at 100°C before bulging equibiaxially to 4% thickness strain and then bake-hardening.

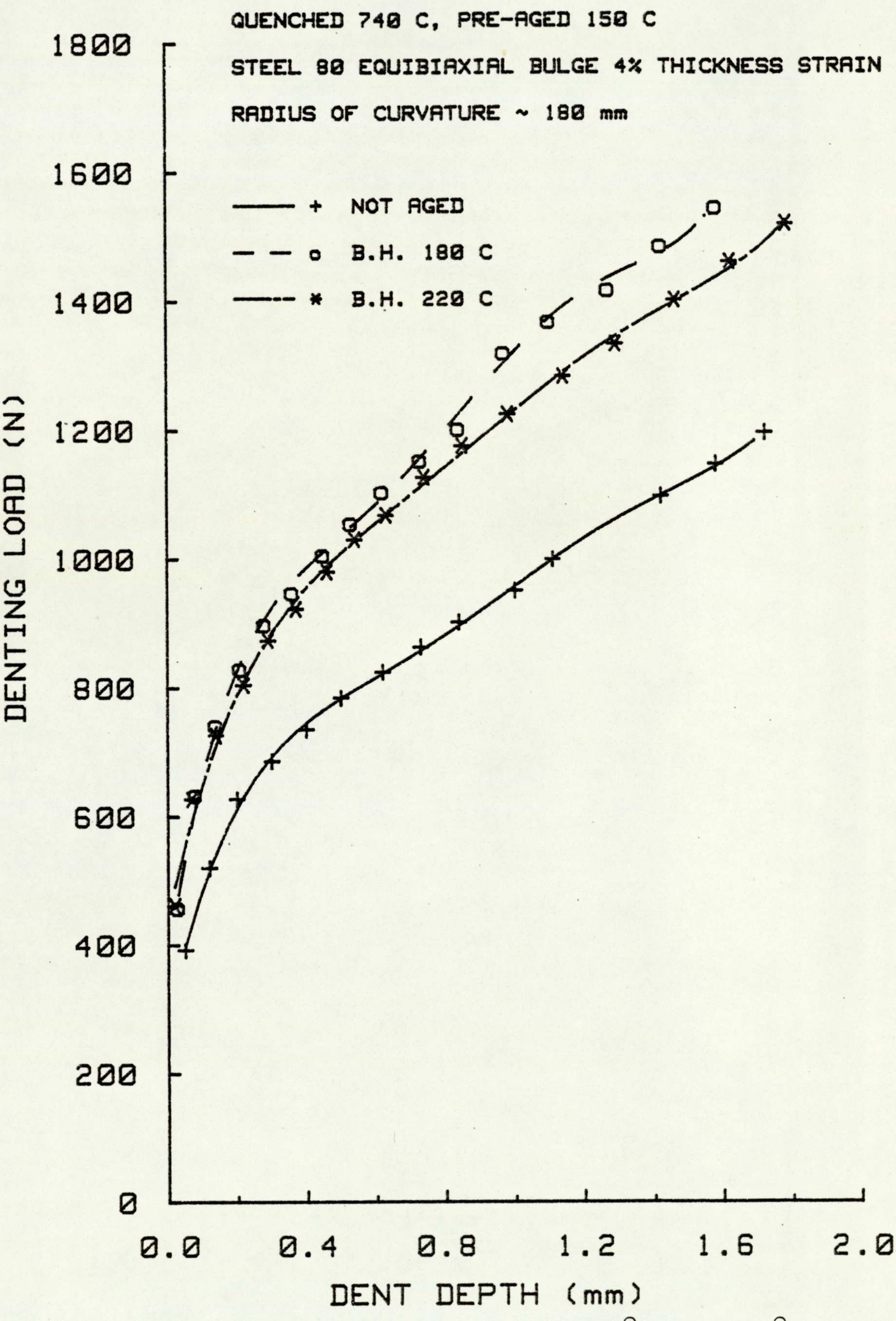


Fig 77. Effect of bake-hardening at 180°C and 220°C on dent-resistance of curved panels of steel 80, quenched from 740°C and pre-aged at 150°C before bulging equibiaxially to 4% thickness strain and then bake-hardening.

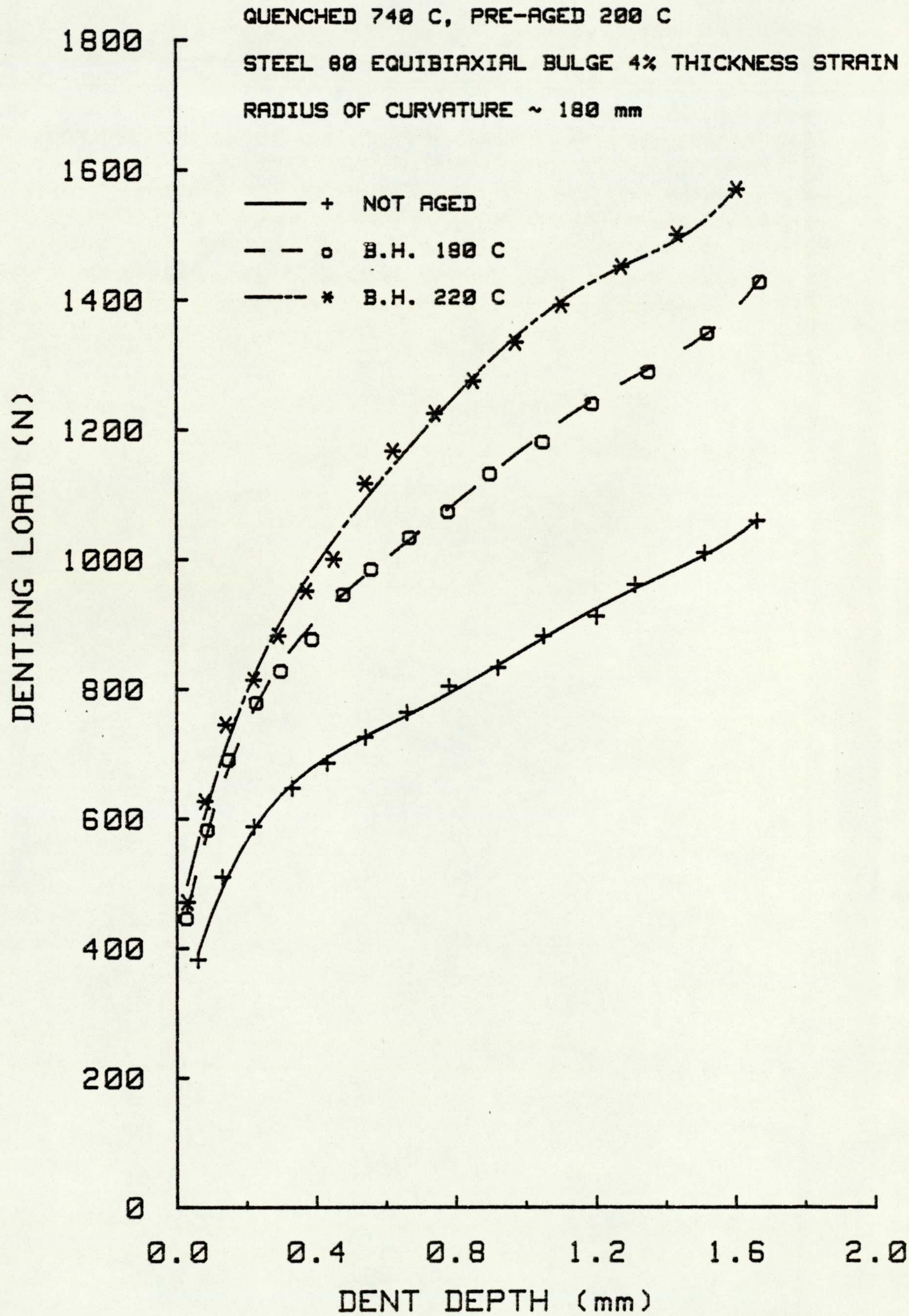


Fig 78. Effect of bake-hardening at 180°C and 220°C on dent-resistance of curved panels of steel 80, quenched from 740°C and pre-aged at 200°C before bulging equibiaxially to 4% thickness strain and then bake-hardening.

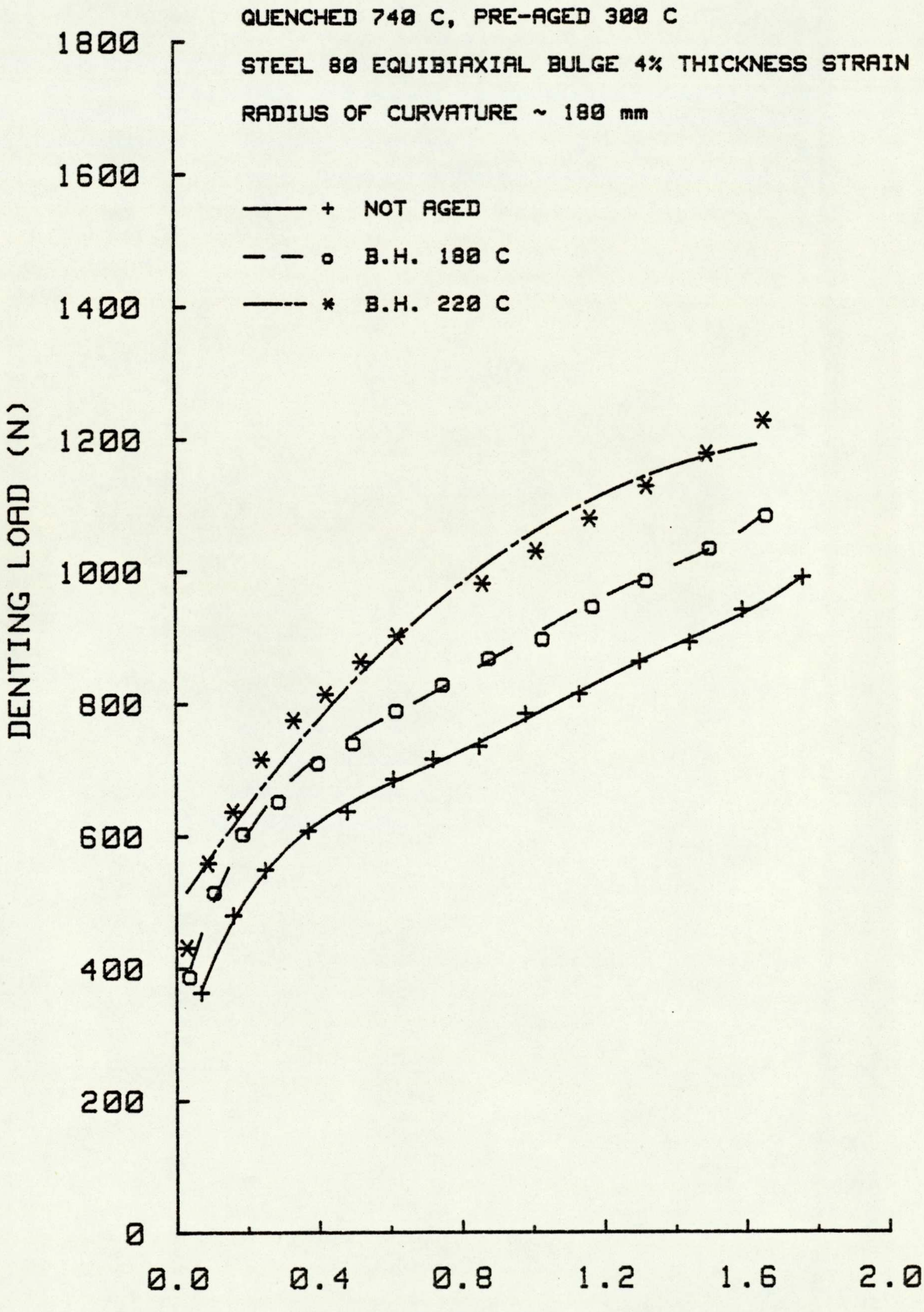


Fig 79. Effect of bake-hardening at 180°C and 220°C on dent-resistance of curved panels of steel 80, quenched from 740°C and pre-aged at 300°C before bulging equibiaxially to 4% thickness strain and then bake-hardening.

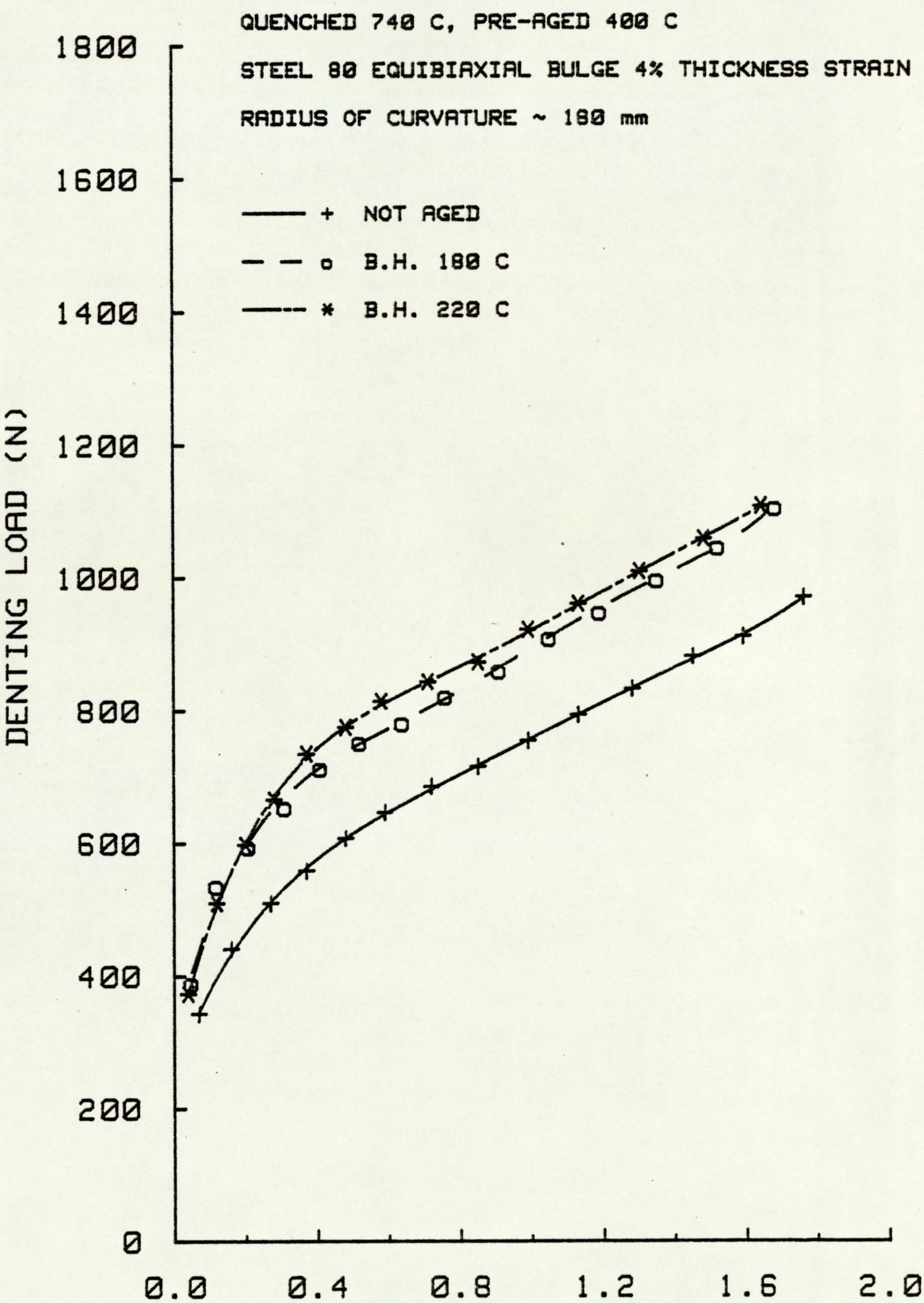


Fig 80. Effect of bake-hardening at 180°C and 220°C on dent-resistance of curved panels of steel 80, quenched from 740°C and pre-aged at 400°C before bulging equibiaxially to 4% thickness strain and then bake-hardening.

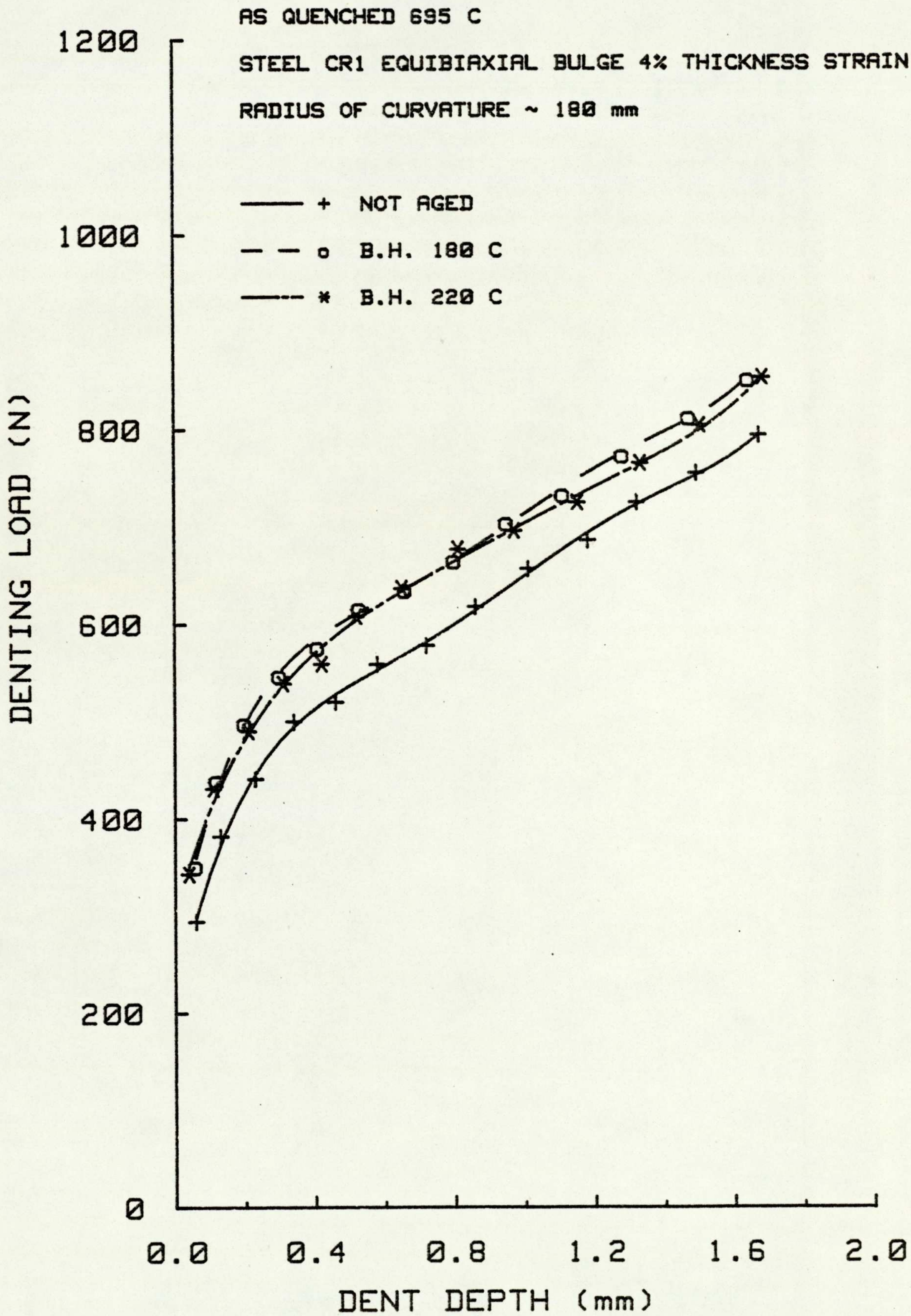


Fig 81. Effect of bake-hardening at 180°C and 220°C on dent-resistance of curved panels of steel CR1, quenched from 695°C and then bulged equibiaxially to 4% thickness strain before bake-hardening.

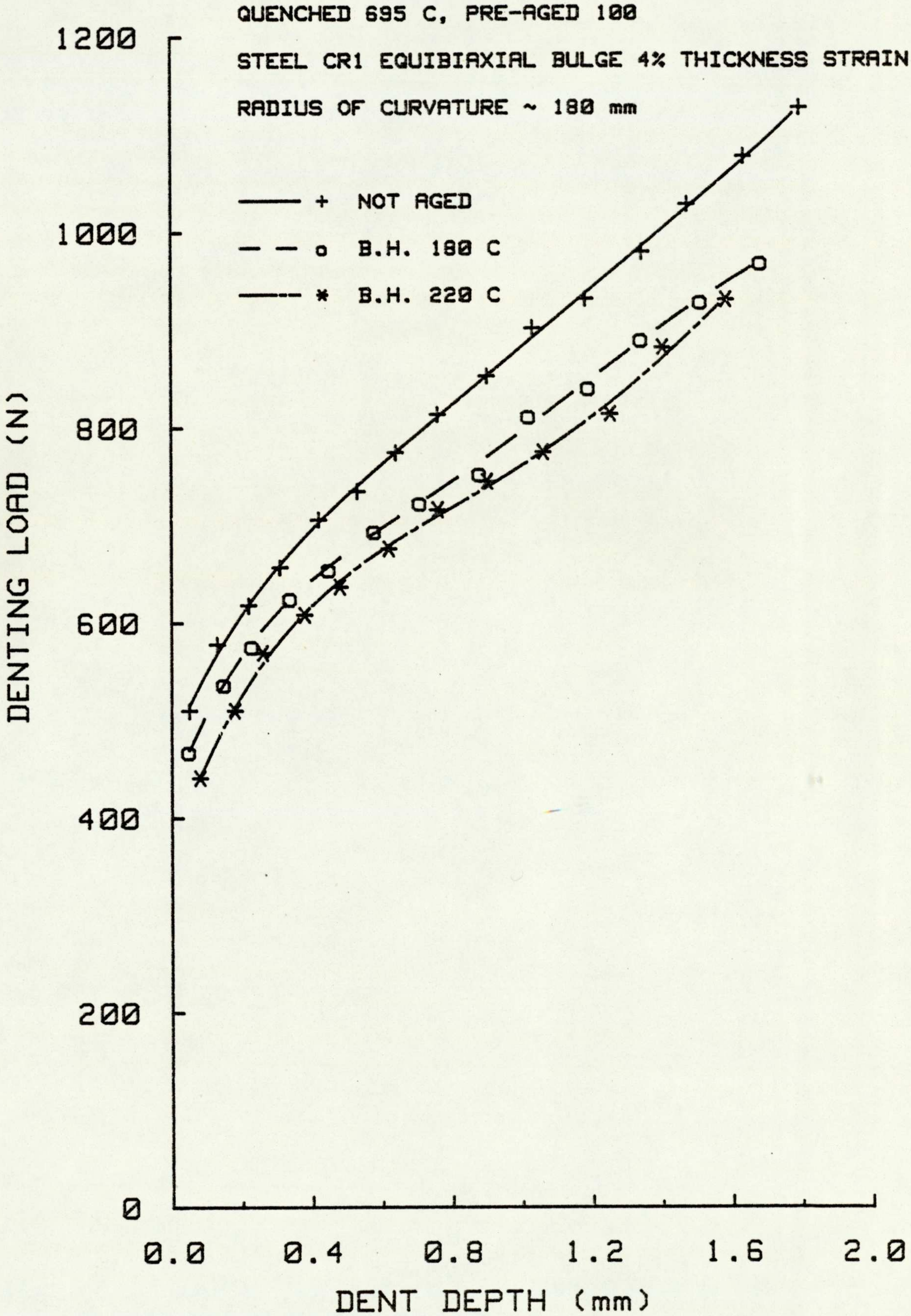


Fig 82. Effect of bake-hardening at 180°C and 220°C on dent-resistance of curved panels of steel CR1, quenched from 695°C and pre-aged at 100°C before bulging equibiaxially to 4% thickness strain and then bake-hardening.

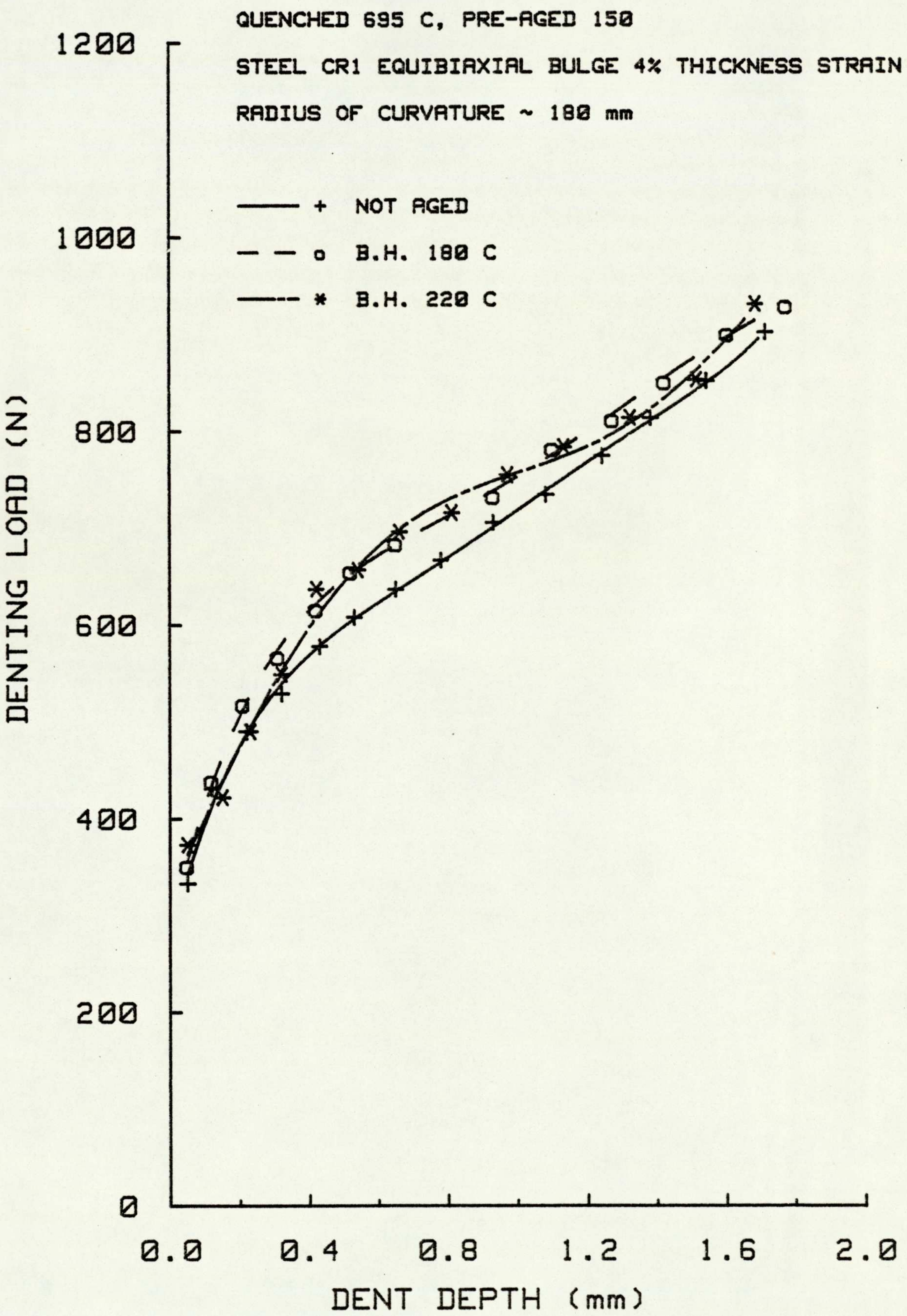


Fig 83. Effect of bake-hardening at 180°C and 220°C on dent-resistance of curved panels of steel CR1, quenched from 695°C and pre-aged at 150°C before bulging equibiaxially to 4% thickness strain and then bake-hardening.

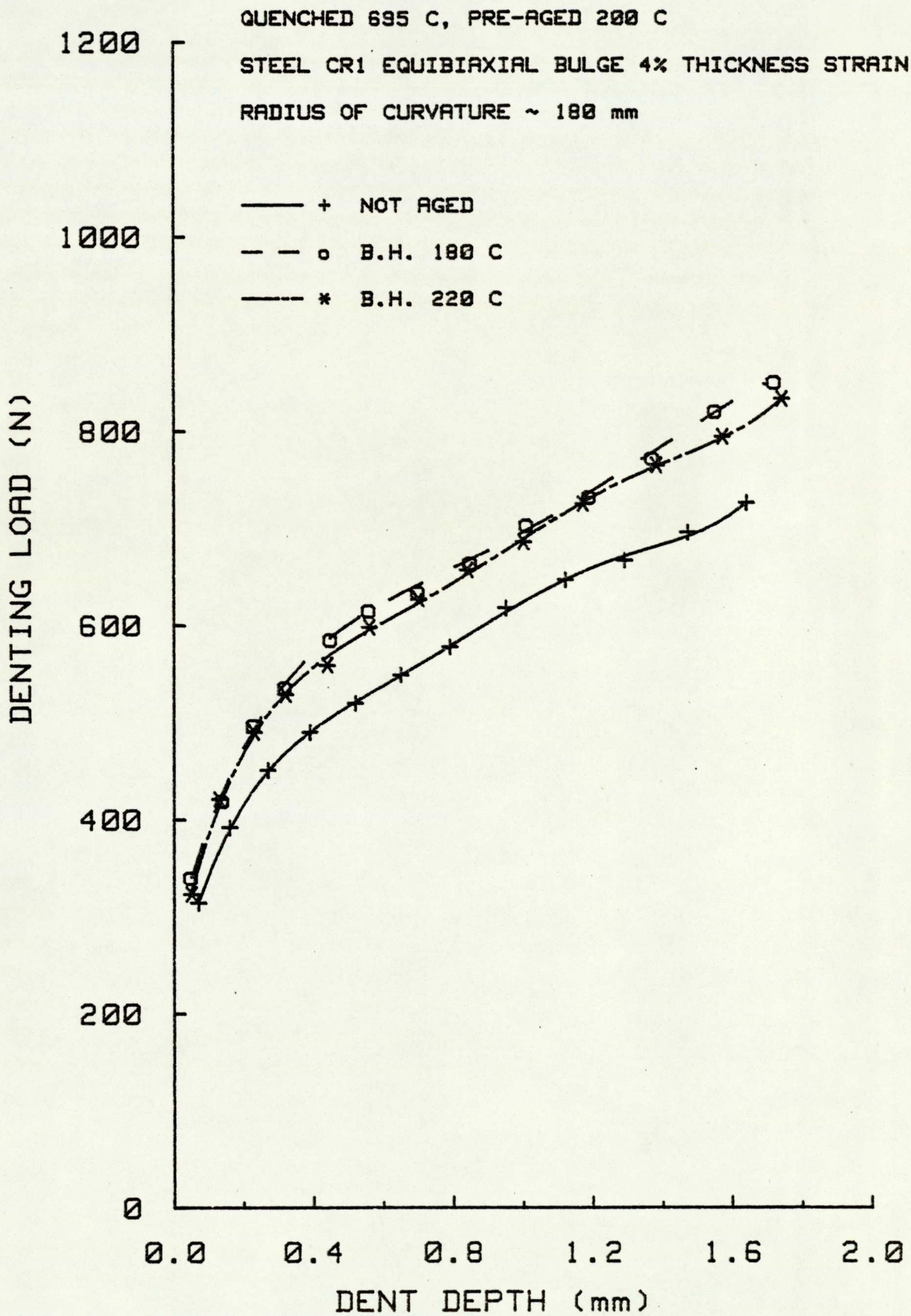


Fig 84. Effect of bake-hardening at 180°C and 220°C on dent-resistance of curved panels of steel CR1, quenched from 695°C and pre-aged at 200°C before bulging equibiaxially to 4% thickness strain and then bake-hardening.

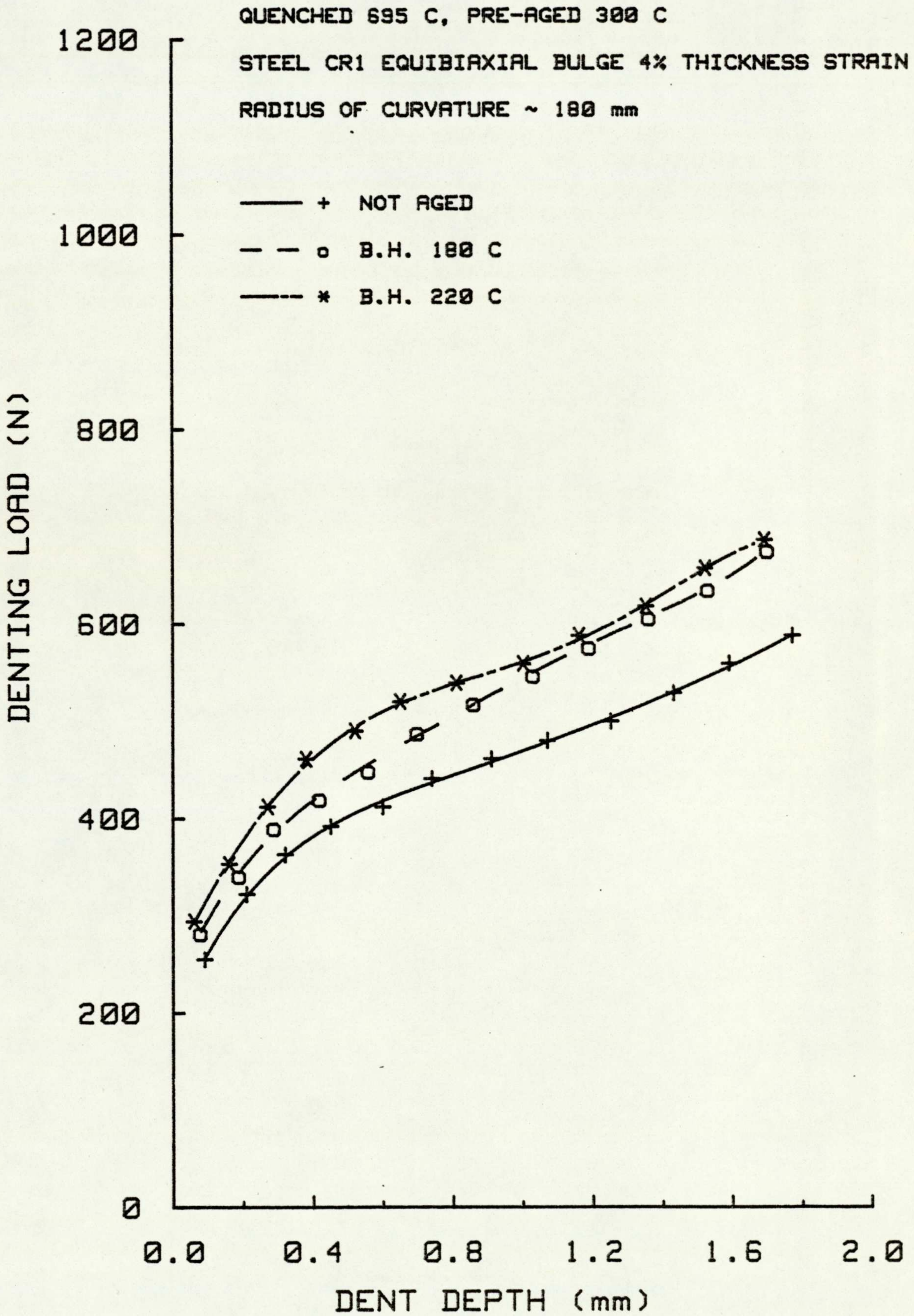


Fig 85. Effect of bake-hardening at 180°C and 220°C on dent-resistance of curved panels of steel CR1, quenched from 695°C and pre-aged at 300°C before bulging equibiaxially to 4% thickness strain and then bake-hardening.

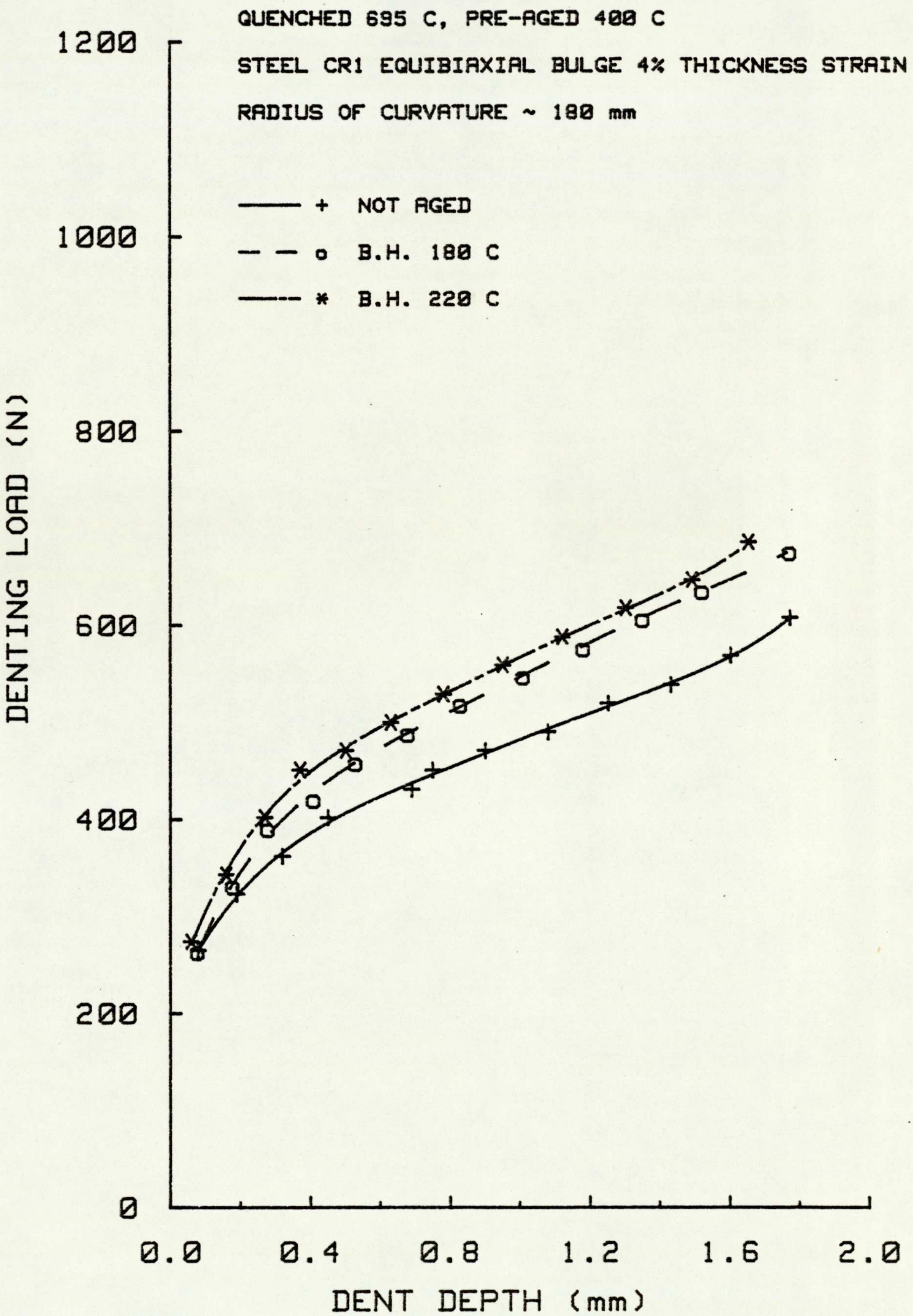


Fig 86. Effect of bake-hardening at 180°C and 220°C on dent resistance of curved panels of steel CR1, quenched from 695°C and pre-aged at 400°C before bulging equibiaxially to 4% thickness strain and then bake-hardening.

CHAPTER 5

DISCUSSION

The main objectives of the research were, first, to evaluate the directionality of the flow strength developed in lightly stretched sheets of two representative dual-phase steels and then to explore how the directional strength can be changed by strain-ageing in a paint-baking treatment applied after pressing.

In the first part of the work the evaluations were made using uniaxial tensile testpieces which were cut from sheets which had been prestrained in uniaxial stretching, plane strain stretching or equibiaxial stretching. Previous work has shown that maximum differences in uniaxial yield strength are developed between the directions in the sheet plane which are parallel and perpendicular to the direction of maximum extension in prestraining. The tests were therefore confined to these two directions and flow strengths were characterised in terms of the 0.5% proof stress and the UTS. These measurements define the resistance to deformation by static unidirectional stresses acting in the sheet plane. However, an important property of external body panels is resistance to denting caused by forces applied locally at a large angle to the sheet plane. Such denting involves complex stress systems and, although it is known to depend on sheet thickness, yield strength and panel curvature, the influence of directional strength in the sheet plane is not well understood. In the second part of the research, dent resistance was explored experimentally using the same materials, pre-treatments and ageing conditions as were used in the investigation with tensile tests. In this discussion the results of the tensile tests will be considered first.

The two dual-phase steels used in the present work were lean-alloy steels. Their carbon contents in weight % were 0.06 and 0.09 respectively. The dual-phase steel 60 having a composition similar to those of conventional batch annealed stabilised steels, while the dual-phase 80 had a higher manganese content of $\sim$ 0.5% weight % and higher phosphorous $\sim$ 0.06%. In the as-received conditions, the volume fractions of martensite were 0.11 for steel 60 and 0.14 for steel 80. For the purpose of comparison, parallel tests were made on a commercial stabilised steel of CR1 quality.

5.1 Investigation Using Tensile Tests

5.1.1 Tests Made in the 'As-Received' Condition

In these tests the results with uniaxial and plane strain stretching were very similar and they can be considered together. The 0.5% proof stress and UTS values of the aluminium stabilised steel showed little change after ageing at elevated temperatures of 90°, 135°, 180° and 220°C and they also show only small differences in the flow stress values measured at 0° and 90° to the direction of maximum extension in stretching. This is typical of a material which contains only a very small volume fraction of second phase particles and which develops only a small Bauschinger effect. The behaviour is different from that of the dual-phase steels which contain relatively large volume fractions of martensite.

With the dual-phase steel the value of the 0.5% proof stress measured at 90° to the direction of maximum extension in stretching was generally significantly lower than measured at 0° to the stretching direction. This is associated with the relatively large Bauschinger effects which are caused by the dispersion of hard martensite particles.

The 'as-received' dual-phase steels were aged, after rapid cooling from the annealing temperature, at the relatively low temperature of 260°C for $\sim$ 1 minute. The effect of this is to develop relatively high bake-hardenability in ageing at the paint-stoving temperature (usually close to 180°C) after pressing. There were useful increases in both the proof stress and UTS after ageing for 30 minutes above 135°C, but the effects of ageing on the proof stress were rather different from those in the UTS. The most important difference is that the increases in the proof stress values are much greater than those in the UTS after ageing at 180°C and 220°C.

A point of particular interest in the investigations is the extent to which the directionality in proof stress values, which is developed in stretching, can be reduced by ageing at paint stoving temperatures. In general, there were very little changes in directionality after ageing up to 135°C. There were significant reductions in the directionality after ageing at 180°C but ageing at 220°C was necessary in order to eliminate most of the directionality of the proof stress. These results demonstrate a limitation in the improvement of strength properties which can be achieved at the usual paint-stoving temperature of 170°-180°C.

Qualitatively, the results obtained in ageing after prestraining in rolling are rather similar to those obtained after stretching in uniaxial tension and plane strain, but the extent of the directionality in the 0.5% proof stress was much smaller in the rolled testpieces. This difference is believed to be associated with the more complex deformation applied in rolling which leads to a smaller Bauschinger effect than that developed in unidirectionally stretched sheets.

The results of ageing tests made on the three steels after stretching in balanced biaxial tension to a thickness strain of 4% showed no significant variation in the flow strength with direction in the sheet plane. However, in contrast to the behaviour of the CR1 steel, with the dual-phase steels the 0.5% proof stress values in the 'as stretched' condition were quite low relative to the level of the UTS values. Comparing these results with those obtained after uniaxial and plane strain stretching, the difference between the 0.5% proof stress and UTS values for the biaxially stretched sheets were closer to the corresponding difference between the proof stress and UTS values measured at 90° to the stretching direction. Ageing at temperatures above 135°C caused a considerable improvement in the proof stress values of the biaxially stretched sheets but, as in the case of the 90° proof stress values of unidirectionally stretched sheets, the increase in proof stress after ageing at 180°C was much less than that obtained by ageing at 220°C.

Considering all the results for the tests made on the dual-phase steel in the 'as-received' condition, it is concluded that, while paint stoving at 180°C can give useful increases in the strength values, the effect on the 0.5% proof stress measured at 90° to the stretching direction in unidirectionally stretched sheets is rather disappointing.

5.1.2 Tests Made on Sheets Preaged at Different Temperatures

The programme of tests on the re-heat treated sheets was designed to investigate the influence of variations in the ageing potential of the sheets. In this case, the dual-phase steels were quenched from the annealing temperature of 740°C and then, in most cases, they were aged for 3 minutes at 200°, 300°or 400°C before stretching. The CR1 steel was quenched from 690°C to maintain a single-phase structure but was also given similar ageing treatments before stretching in order to vary the strain-ageing potential of the steel. Preageing temperatures within the 200°C to 400°C temperature range had a large effect on the flow stress values of the steel.

Figs 87 to 95 summarise some of the effects of variations in both the pre-ageing temperature and the final bake-hardening temperature applied after 4% pre-strain in uniaxial tension on the 0° and 90° 0.5% proof stress values of the two dual-phase steels and CR1. In the as-quenched condition with dual-phase steels, bake hardening for 30 minutes at 180°C was sufficient to eliminate most of the directionality of the proof stress. Pre-ageing at 300°C or 400°C reduced the strain ageing potential severely, so that the improvement in the proof stress values after stretching and final ageing at 180°C gave relatively small increases in the proof stress values. As with the 'as-received' condition, much better 'bake-hardenability' was obtained with the sheets that had been pre-aged near 200°C.

5.2 The Underlying Causes of the Observed Changes in Bake-Hardening

As already mentioned, the much greater directionality of proof stress values obtained in the stretched dual-phase steels, as compared with sub-critically annealed CR1 steel, is believed to be a consequence of the directional internal stress system which is developed during plastic deformation of an alloy strengthened by non-deforming particles.

The tests made with the steels in a range of conditions which gave different ageing potentials show that the extent of the general rises in flow strength and the reduction in directionality of the 0.5% proof stress during the final ageing treatment depended on the ageing potential of the steels. It is concluded that these observed changes in strength were controlled mainly by the strength and completeness of dislocation locking during the final elevated temperature ageing treatment.

For the dual-phase steel sheets in the as-received condition or after preageing at 200°C before prestraining in uniaxial tension or plane strain, tensile tests made at 0° to the direction of maximum extension in the prestrain show a tendency towards return of the yield point after bake-hardening at 90°C and a more definite inflection in the load/elongation relationship after bake-hardening at 135°C. Bake-hardening at 180°C and 220°C gave well developed sharp yield points. For tests at 90° to the direction of maximum extension under the same conditions the load elongation curves were well-rounded after BH at 90° and 135°C. After BH at 180°C weak yield points were observed and after BH at 220°C the yield point was well developed. After stretching in equibiaxial tension sheets preaged at 200°C (3 minutes) or 260°C (1 minute) showed a return of the yield point in BH at 180°C and 220°C which was rather similar to that observed in the tests at 90°C with the unidirectionally prestrained sheets. With sheets of the dual-phase steels preaged at 300°C and 400°C the yield point returned much more gradually with increasing BH temperatures than in the case of sheets preaged at 200°C.

Overageing of the quenched steels at temperatures in the range 200°-400°C, even for times of 1 to 3 minutes, is expected to precipitate nearly all the carbon out of solid solution and most of the nitrogen will be precipitated as aluminium nitride. Thus, after overageing at 200°C or higher temperatures, strain ageing at temperatures close to room temperature, which depends on the concentration of dissolved interstitial solutes, is expected to be very weak. This expectation is confirmed by the results in ageing at 47°C after temper rolling. However, at the higher ageing temperatures applied in bake-hardening, fine carbide particles present within the microstructure have significant solubility and carbon can be transferred from the precipitates to sites close to the dislocation cores where binding of interstitial atoms is strong. The extent to which such dislocation locking, by transfer of carbon from precipitates, can develop depends on the solubility of the precipitates at the ageing temperature. As mentioned in the literature survey, overageing at relatively low temperature, such as 200°C, for very short ageing times, can give very small precipitates of an intermediate carbide (probably the ε-carbide) which have higher solubility than cementite particles. The presence of such intermediate carbides in samples overaged at 200°C for 3 minutes or 260°C for 1 minute, probably contributes to the distinctly higher ageing potential in these conditions, relative to those obtained by overageing at 300°C or 400°C, which gives cementite precipitates. However, very small cementite particles are expected to have significant solubility in the cold worked structure at temperatures in the 180°-220°C range. These effects of the size and composition of the carbide precipitates and the influence of temperature on their solubility are believed to underly the observed effects of preageing temperatures and the final bake-hardening temperature on the strength of the bake-hardening reaction.

However, it is not clear that dislocation locking is the only factor responsible for the changes observed at the higher bake-hardening temperatures, particularly those which accompany ageing at 220°C which cause a large reduction in the directionality of the proof stress. The 0.5% proof stress values are increased appreciably by increasing the bake-hardening temperature from 180°C to 220°C. The flow stress at high plastic strains, defined by the UTS, is not significantly increased by this rise in ageing temperature. This is possible evidence that some small degree of dislocation rearrangement or recovery occurs at the higher bake-hardening temperatures. If dislocation rearrangement does occur it would be expected to reduce the internal stresses which are responsible for the directionality of the proof stress in the dual-phase steels.

Results obtained with the CR1 steel are of interest in relation to the possible occurence of a recovery reaction. First it should be noted that in the 'as received' batch-annealed condition the results for the CR1 steel, which show negligible strain-ageing, give no indication of any softening of the cold worked structure in final ageing treatments up to 220°C. It follows that, if recovery does take place in the quenched and overaged conditions of the steels, it must be associated with the presence of the fine carbide particles and/or, in the dual-phase steels with tempering of the martensite. In considering possible changes of this kind it is helpful to examine results obtained with preageing at 100°C and 150°C. These results are clearly consistent with 'recovery' or tempering of the structure during ageing at 180°C or 220°C.

It is noticeable that the CR1 steel which was quenched from 695°C and aged at 150°C shows effects which are similar to those observed with the dual-phase steels. In the CR1, the changes must be associated with coarsening and resolution of the very fine carbide particles which are precipitated at the low preageing temperature. The fine carbide dispersion clearly contributes significantly to strengthening of the steel and changes affecting the carbide dispersion would be expected to influence the internal stresses present after cold working. Presumably, rather similar effects of changes in the fine carbide precipitates will occur in the dual-phase steels. This does not rule out additional affects associated with tempering of the martensite but the results do not provide independent evidence of this.

It should be noted that, for samples preaged at 200°C, the final ageing at 220°C will certainly cause changes in carbide distribution and further tempering of the martensite. The effect of final ageing at 180°C on tempering expected to be smaller in this condition but, due to the strain ageing reaction, it may still cause considerable changes in the carbide dispersion.

5.3 Dent Resistance

The most complete published work on the influence of strength, sheet thickness and panel curvature on static dent resistance is that of Miyahara and Yutori[73,74]. They showed that the load required to initiate a dent was given by the relationship:

$$Pd = k\ t^{m}\ \sigma y \qquad (1)$$

where k is a constant, t is sheet thickness and σy is the yield strength, Pd was taken = $P_{0.1}$, the load for 0.1mm dent depth. The range of m for curved panel was 2.2-2.5 and for flat panels 1.4-1.8. The strength of the panel had no systematic effect on the value of m. If, from the above earlier work, their average values of m, 2.4 for curved panels and 1.6 for the flat panels, are taken to apply to present results, then the effects of the change in thickness from 0.89mm for the dual-phase steels to 0.77mm for the CR1 steel would give a reduction of 21% with flat panels and 29% with the curved panels.

The curved panels used in this research were of a considerably smaller radius of curvature than those investigated by Yutori[74] and it is likely that the m coefficient was different from that which they found. The effects of sheet thickness were not investigated in this research but a comparison of results for the flat and curved panels allows an estimate to be made of the ratio of m values for the panel forms used.

For given values of k, t and σy, equation (1) predicts that the ratio of m values for flat and curved panels will be given by:

$$\frac{m\ (\text{curved})}{m\ (\text{flat})} = \frac{\log P_d\ (\text{curved})}{\log P_d\ (\text{flat})}$$

where P_d is the load required to form a small dent of standard depth. Values of m (curved)/m (flat) for uniaxial stretching and equibiaxial stretching were derived in this way and they are given in table (1). The derived ratios shown in table (1) show significant differences with the three steels. The indication is that mean m value ratios were significantly higher with the CR1 steel than with the dual-phase steels and that the values of dual-phase 60 were slightly higher than those for dual-phase 80. Considering the extremes and assuming that m for flat panels was about 1.6, the mean values of m for curved panels of CR1 would be about 2.8, that for the dual-phase 60 would be about 2.1 and for dual-phase 80 about 1.8.

Figs 96, 97 and 98 summarise the influences of variations in the preaged treatments within range 150°-400°C for 3 minute periods on the load which gave the 0.2mm and 1.0mm dents in as-stretched and bake-hardened conditions. All the preageing treatments gave substantial increases with the dual-phase steel and even with the CR1 steel.

For the high-bake hardenability condition given by preageing at 200°C the increases in dent resistance for the curved panels were substantially higher than those for the flat panels. More generally, for 0.2mm dents on sheets preaged in the 200°C-400°C range, the increments in dent resistance caused by bake hardening in the curved panels were greater than those with flat panels. It is possible that these differences in the behaviour of flat and curved panels is related to differences in the stress systems generated by denting. Yutori et al[74] have shown that with curved panels there is a larger compressive component in the stress system.

For sheets pre-aged at 150°C it is clear that tempering and/or recovery during bake-hardening has a large influence; for example, the increases in dent resistance caused by bake-hardening at 220°C are either less than or very little greater than those given by BH at 180°C. In view of the results obtained with pre-ageing at 150°C it seems possible that behaviour in conditions pre-aged at 200°C may also be influenced to some extent by tempering or recovery during the BH treatments; for example, with flat sheets of the 80 steel uniaxially prestrained the increment in dent resistance caused by increasing the BH temperatures from 180°C to 220°C was greater in the pre-aged at 300°C condition than in the pre-aged at 200°C. Curved panels of the 60 steel showed the unexpected result that, for the 0.2mm dent, BH at 180°C gave a somewhat better dent resistance than that after BH at 220°C.

The loads for 0.2mm dents in flat and curved panels in the as-stretched and BH conditions are summarised in tables 2, 3, 4 and 5. Individual results show a fairly wide scatter but, considering results of the two dual-phase steels, to a first approximation the absolute increases in denting load for a particular dent depth and BH treatment are of the same order of magnitude for dents with the four different conditions of prestraining, although on the whole the increments with prestraining in equibiaxial tension may tend to be slightly lower than those with prestraining in uniaxial stretching or plane strain. The increments for ageing at 220°C are significantly higher than those for ageing at 180°C, but while the mean ratio of the increments increase for BH at 220°C/increase for BH at 180°C, for the flat panels is about 1.2 , the corresponding ratio for the curved panels is about 1.6.

The absolute increments in denting load obtained with the CR1 steel tested under equivalent conditions were typically about 0.3 of the corresponding increments obtained with dual-phase steels. However, the dent resistance of the CR1 steel is very much lower than those of the dual-phase steels and, if proportionate, rather than absolute increases are considered, then values for the CR1 steel are generally rather similar to those for the dual-phase steels.

Comparing the results for increases in dent resistance caused by BH at 180°C and 220°C for 30 minutes with the increase in 0.5% proof stress values, given in tables 6 and 7, it is evident that the tensile test results do not give a quantitative prediction of the effects of BH on dent resistance. In all cases the proportionate increases in dent resistance were greater than would be predicted from the increase in the 0.5% proof stress values. The relationship between the 0.5% proof stress values and the dent resistance loads, for 0.2mm and 1.0mm deep dents, are summarised in Figs 99 to 104. The filled symbols refer to conditions bake-hardened at 180°C and 220°C which are not distinguished in this case. Open symbols refer to results in prestrained conditions which were not bake-hardened. Results for heat-treated conditions with the pre-ageing temperatures of 200°, 300°and 400°C are included with the as-received condition where they are available with dual-phase steel.

The rather wide scatter of the results shown in Figs 99 to 104 is in part due to the inclusion of 0.5% proof stress values measured at both 0° and 90° to the direction of maximum extension. However, relations with the proof stress values measured at 0° and 90°, are distinguished by plotting as small and large symbols respectively. It is clear from the plots that, in general, the P_d- proof stress relationships for bake-hardening conditions (filled symbols) tend to be higher than those for the stretched conditions but this is particularly true when results for proof stress values measured in the 0° direction are considered. The general level of results in this former condition is not only lower but also the slope of the P_d - proof relation tends to be less than corresponding to aged conditions. The difference is more clearly resolved in the case of the 1.0mm dents.

Differences in the effects of bake-hardening on the 0.5% proof stress values and the dent resistance are particularly evident in the case of the CR1 steel. Equation (1) indicates that the load required to make a small dent is expected to be directly proportional to the yield strength of the steel. The results show that when the proof stress or yield strength values are measured in the direction of maximum extension either in tensile tests or from specimens cut from actual pressings equation (1) will not necessarily give reliable predictions of the increases in dent resistance due to bake-hardening. This is important because yield stress values measured in these ways have been used to predict dent resistance before and after bake-hardening.[78]

Dent resistance depends on complex stress systems which involves bending stresses and compression as well as tensile stresses[74]. As results of tensile tests made on the dual-phase steel show the proof stress values in tests made at 90° to the direction of maximum extension can be much lower than those measured at 0° to this direction. However, these tests do not involve a complete reversal of the stress system as is involved in compressive straining after tensile straining.

In earlier work, Bauschinger tests made in tension-compression or reversed torsional straining, showed that the proof stress (0.1% offset for example) is extremely low after stress reversal in both dual-phase and sub-critically annealed low-carbon steels. In the case of the latter, measurements made in torsion tests, which allow reversal of the stress systems without cutting or machining, are available from the work of Ogram and Wilson[34]. They include results for a fine grained low-carbon steel having a relatively low ageing potential due to about 0.002 weight % dissolved nitrogen. After 5% prestrain and unloading the 0.1% proof stress in continued forward deformation was 145 MPa but in reversed straining it was only 100 MPa. However, ageing for 10 minutes at 89°C, was sufficient to increase the 0.1% proof stress in reversed straining to about 117 MPa. This was more than double the increase observed in continued forward straining. These results indicate that relatively light dislocation locking can cause rather large increases in the 0.1% proof stress measured in reversed straining. Changes affecting low proof stress values, such as the 0.1% proof stress, may be important in relationship to dent resistance.

As in existing tests, unfortunately it is very difficult to cut and machine testpieces from a large pressing without some ageing and, as a result, the measured 0.1%

proof stress values do not give a reliable indication of the 'as-stretched' condition. Probably the observed increase in dent resistance of CR1 steel after BH in the as-received condition is caused by a small amount of dislocation locking which might be expected to influence proof stress measured at small permanent sets, particularly when restraining involves a stress reversal. In the present work the measurements of the 0.5% proof stress do not detect a change of this kind but the 0.2% proof stress values shown in Fig 14 indicates a rise in elevated temperature ageing.

These comments on the difficulty of measuring reliable low proof stress values, such as the proof stress at 0.1% offset, apply to tests made on testpieces which have to be cut and machined from large prestrained sheets. The results in tables 6 and 7 and Figs 99 to 104 show that the increments in yield stress values made at 0° to the direction of maximum extension did not give a reliable indication of increases in dent resistance caused by BH. However, it is possible that, if reliable values of the 0.1% proof stress could have been measured in the as-stretched condition at 90° to the direction of maximum extension in prestraining, the increment in the 0.1% proof stress values caused by bake-hardening would have been considerably larger than the observed changes in the 0.5% proof stress values. If this were so, it seems that 0.1% proof stress made in the 90° direction might give a more reliable guide to the increments in dent resistance. In order to test this possibility it would be necessary to cut out tensile testpieces from the sheets without any significant ageing.

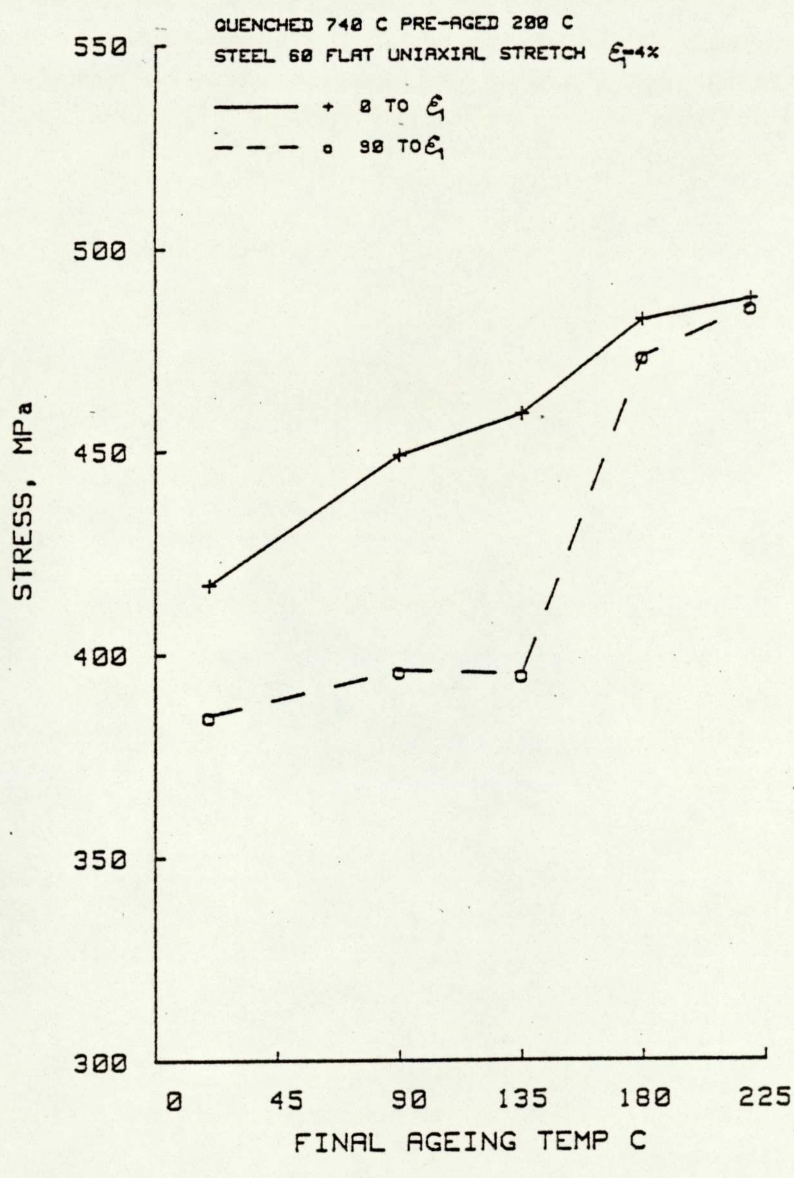


Fig 87.

Figs 87 to 89.

Effects of variations in preageing temperature (for 3 minutes) and bake-hardening temperature (for 30 minutes) on 0.5% proof stress measured at 0° and 90° to direction of extension in prestraining. For dual-phase steel 60 stretched 4% in uniaxial tension before bake-hardening. Fig 87 preaged at 200°C, Fig 88 preaged at 300°C, Fig 89 preaged at 400°C.

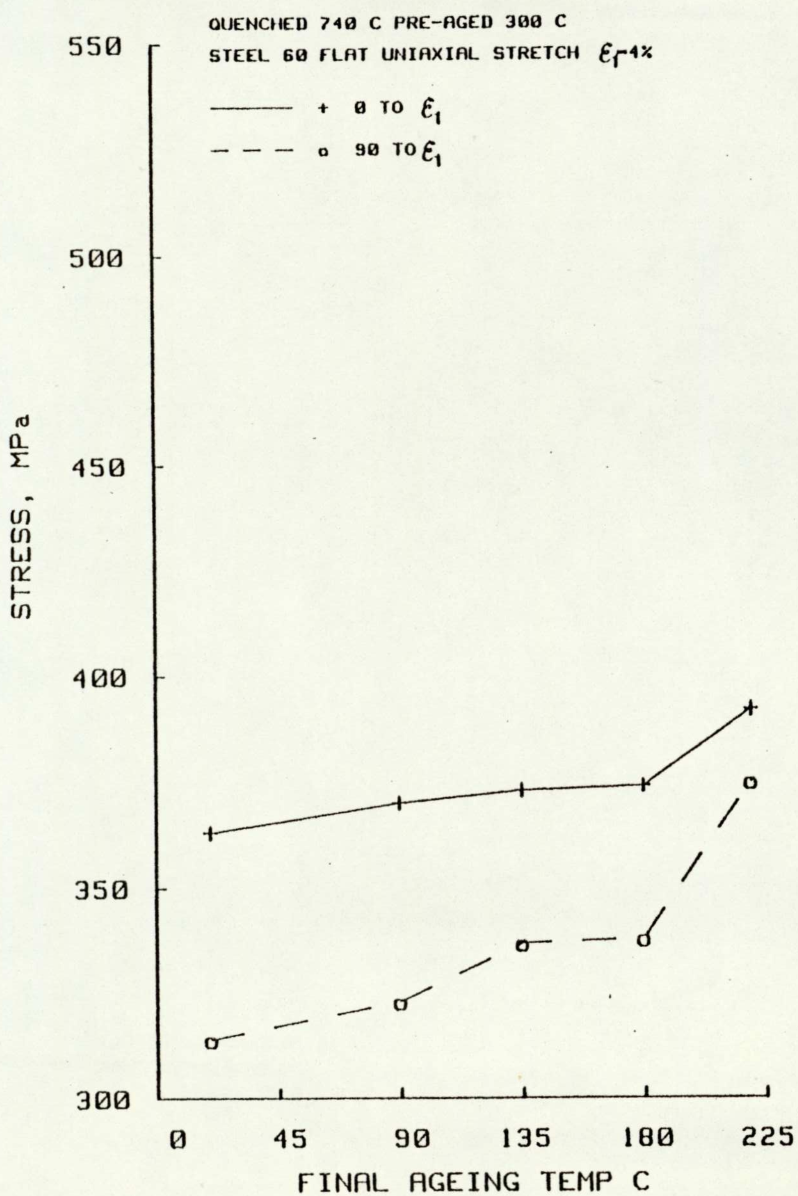

QUENCHED 740 C PRE-AGED 300 C
STEEL 60 FLAT UNIAXIAL STRETCH ε_f 4%
+ 0 TO ε_1
o 90 TO ε_1
STRESS, MPa
550
500
450
400
350
300
0
45
90
135
180
225
FINAL AGEING TEMP C

Fig 88.

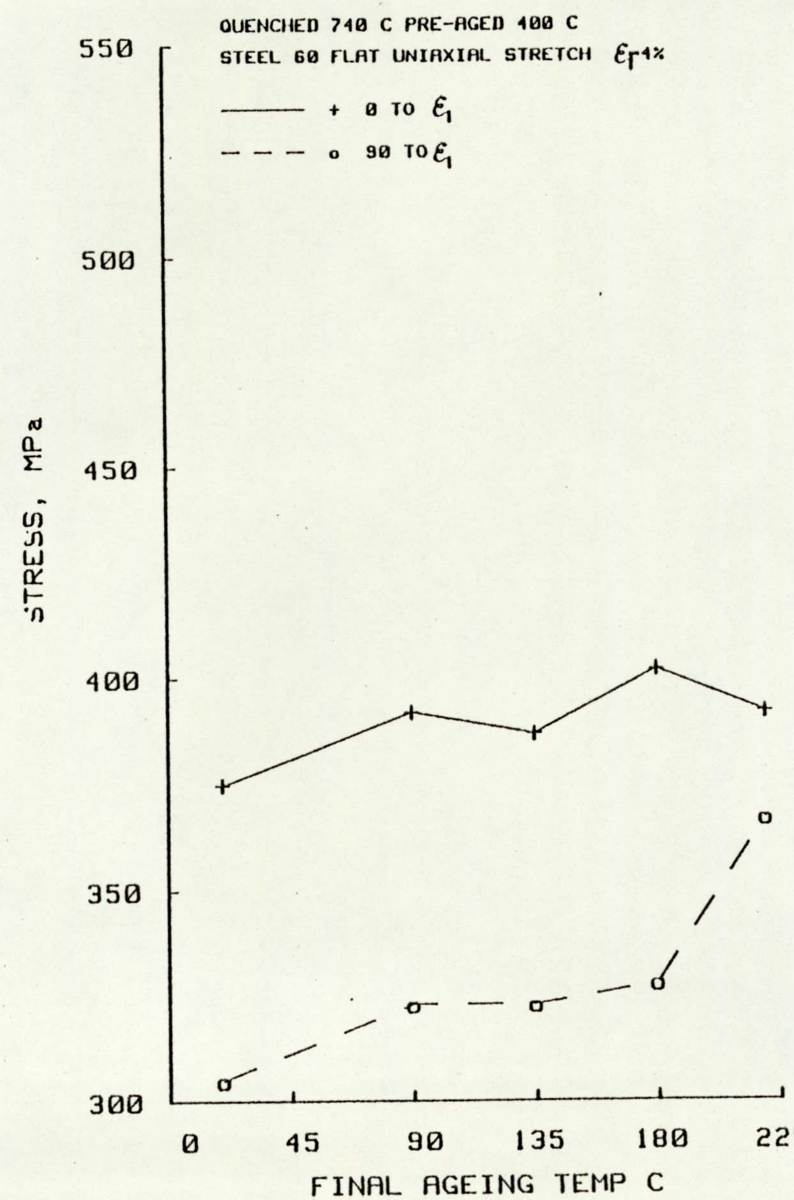

QUENCHED 740 C PRE-AGED 400 C
STEEL 60 FLAT UNIAXIAL STRETCH ε_f 4%
+ 0 TO ε_1
o 90 TO ε_1
STRESS, MPa
550
500
450
400
350
300
0
45
90
135
180
225
FINAL AGEING TEMP C

Fig 89.

Figs 90 to 93.

Effects of variations in preageing temperature (for 3 minutes) and bake-hardening temperature (for 30 minutes) on 0.5% proof stress measured at 0^{o} and 90^{o} to direction of extension in prestraining. For dual-phase steel 80 stretched 4% in uniaxial tension before bake-hardening.

Fig 90 as quenched from $740^{o}C$.

Fig 91 preaged at $200^{o}C$.

Fig 92 preaged at $300^{o}C$.

Fig 93 preaged at $400^{o}C$.

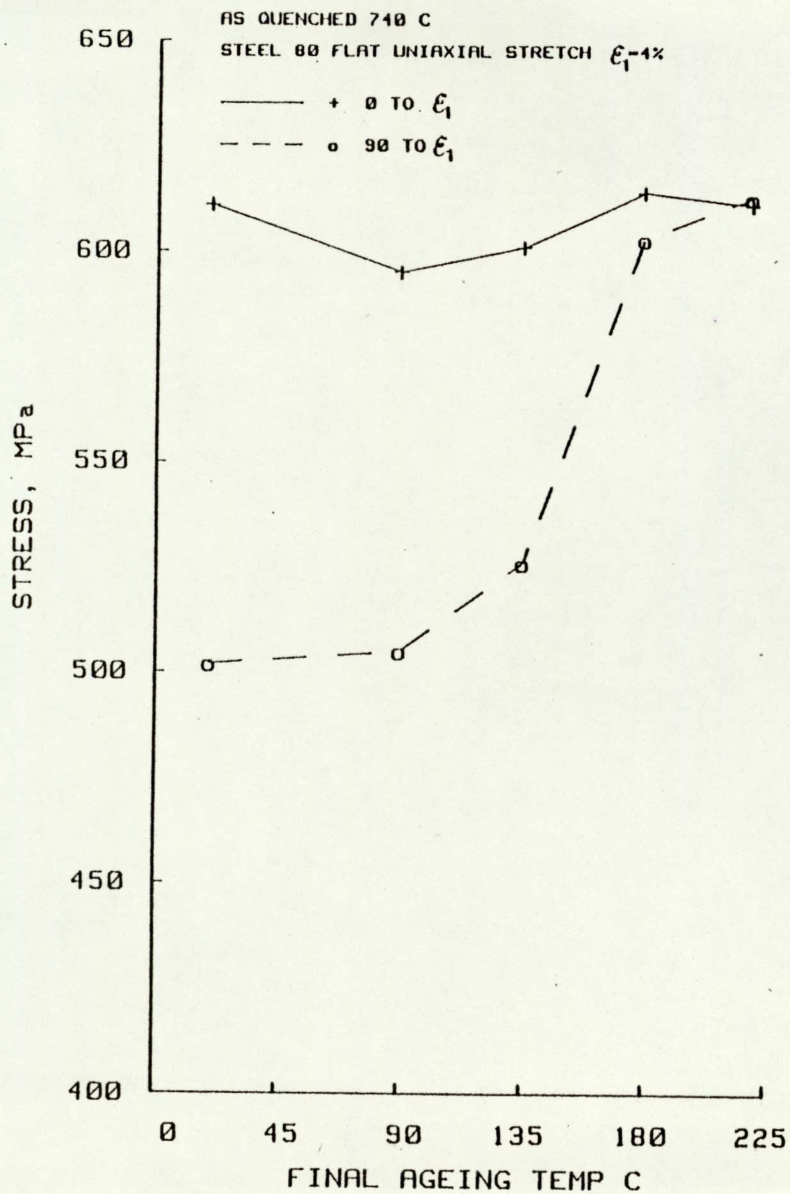

AS QUENCHED 740 C
STEEL 80 FLAT UNIAXIAL STRETCH ε_1-4%
+ 0 TO ε_1
o 90 TO ε_1
STRESS, MPa
650
600
550
500
450
400
0
45
90
135
180
225
FINAL AGEING TEMP C

Fig 90

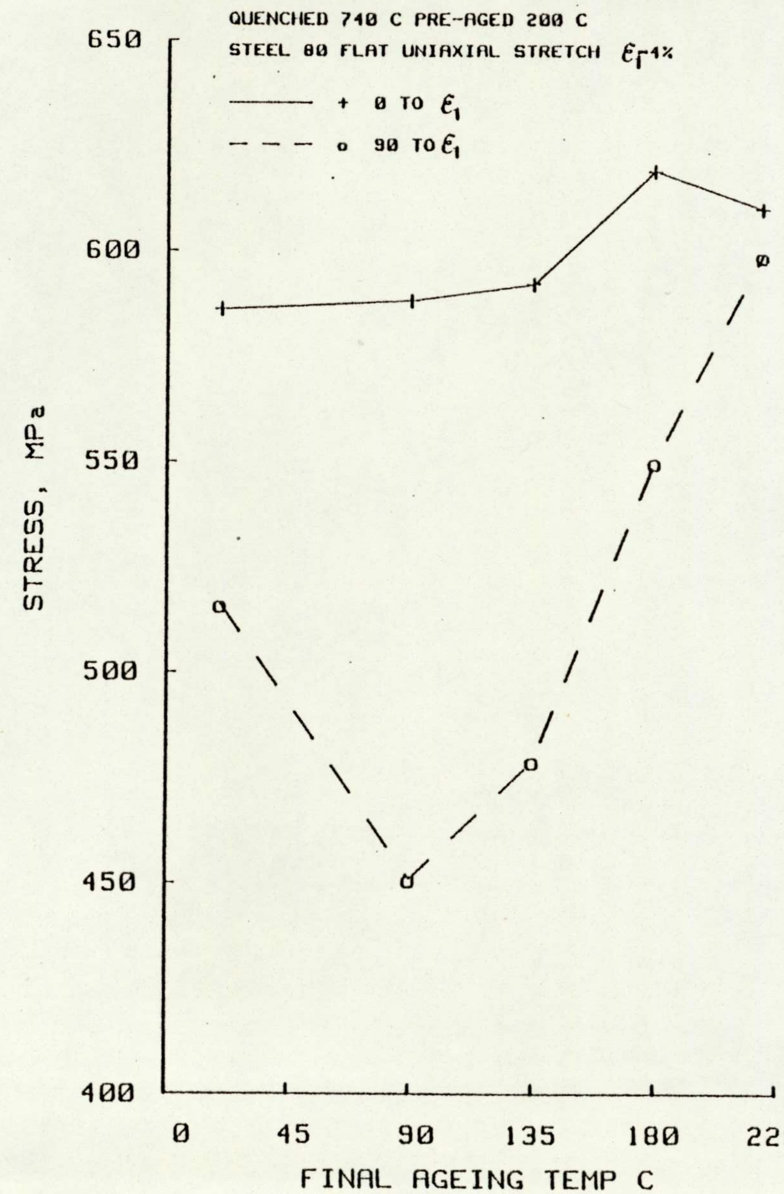

QUENCHED 740 C PRE-AGED 200 C
STEEL 80 FLAT UNIAXIAL STRETCH ε_1-4%
+ 0 TO ε_1
o 90 TO ε_1
STRESS, MPa
650
600
550
500
450
400
0
45
90
135
180
225
FINAL AGEING TEMP C

Fig 91

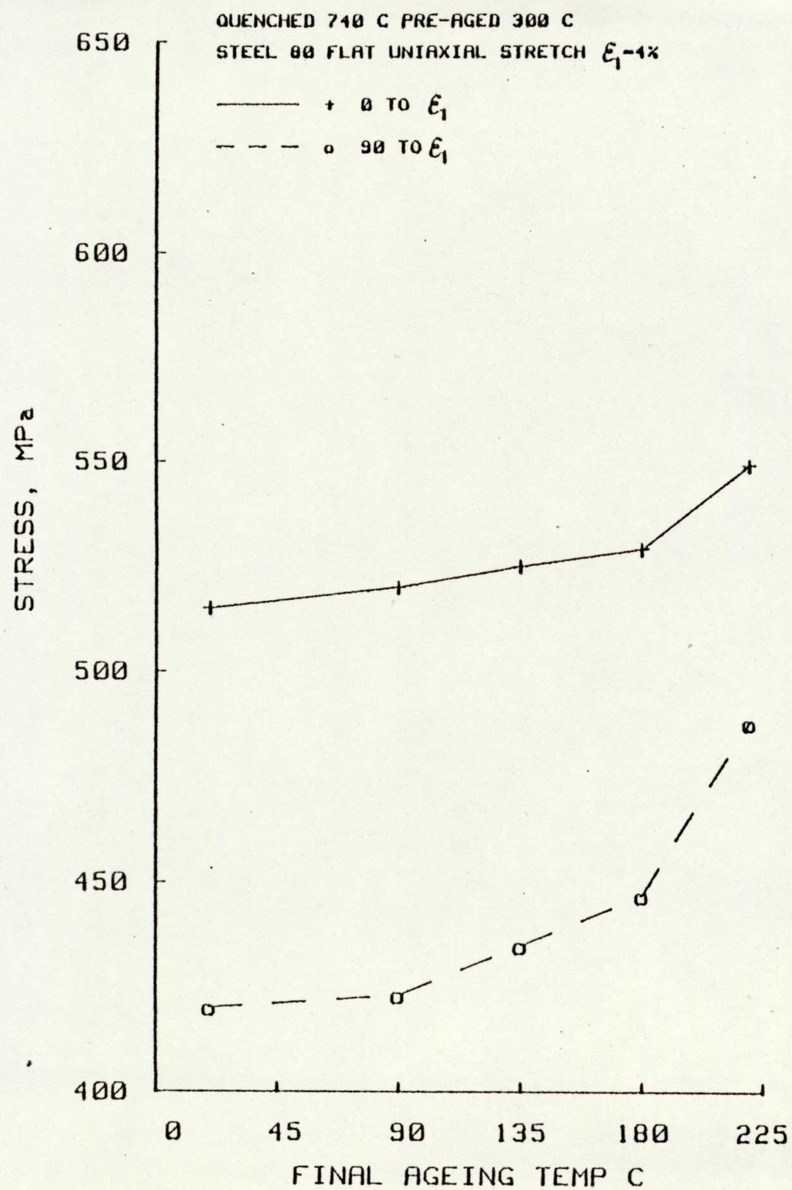

QUENCHED 740 C PRE-AGED 300 C
STEEL 80 FLAT UNIAXIAL STRETCH ε_1-4%
+ 0 TO ε_1
o 90 TO ε_1
STRESS, MPa
650
600
550
500
450
400
0
45
90
135
180
225
FINAL AGEING TEMP C

Fig 92

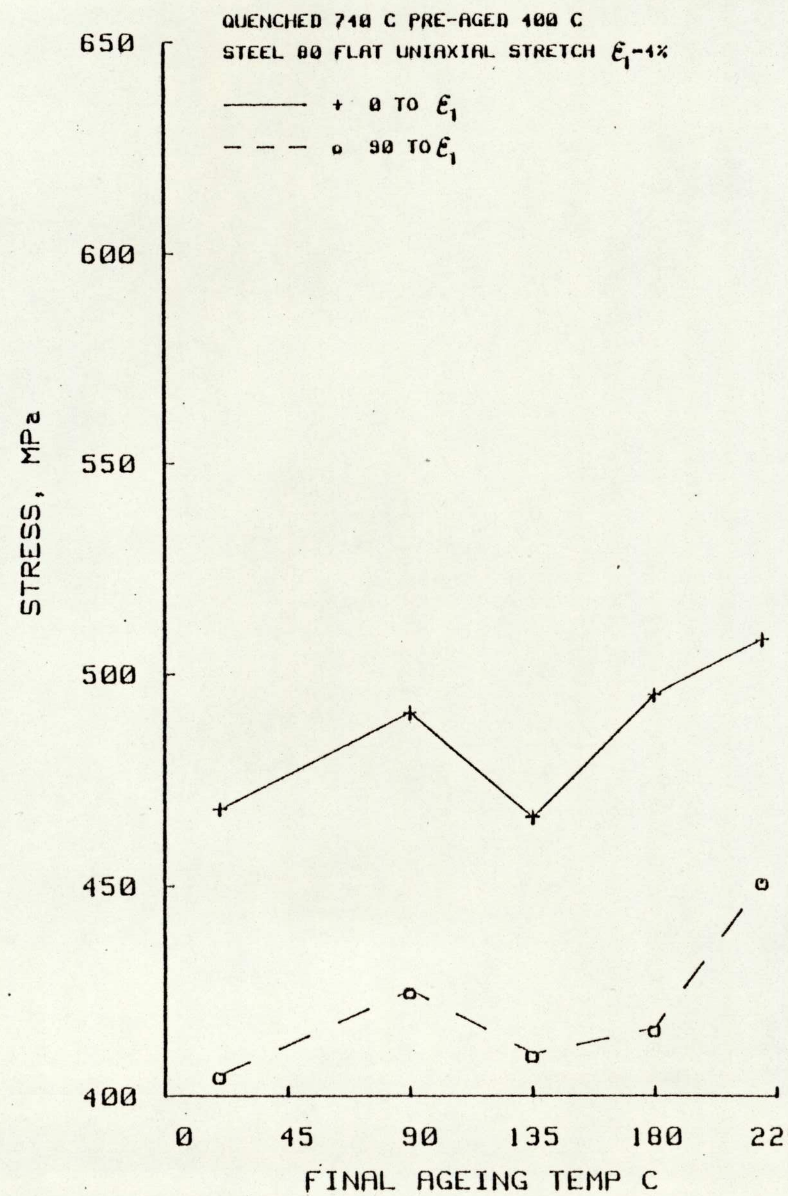

QUENCHED 740 C PRE-AGED 400 C
STEEL 80 FLAT UNIAXIAL STRETCH ε_1-4%
+ 0 TO ε_1
o 90 TO ε_1
STRESS, MPa
650
600
550
500
450
400
0
45
90
135
180
225
FINAL AGEING TEMP C

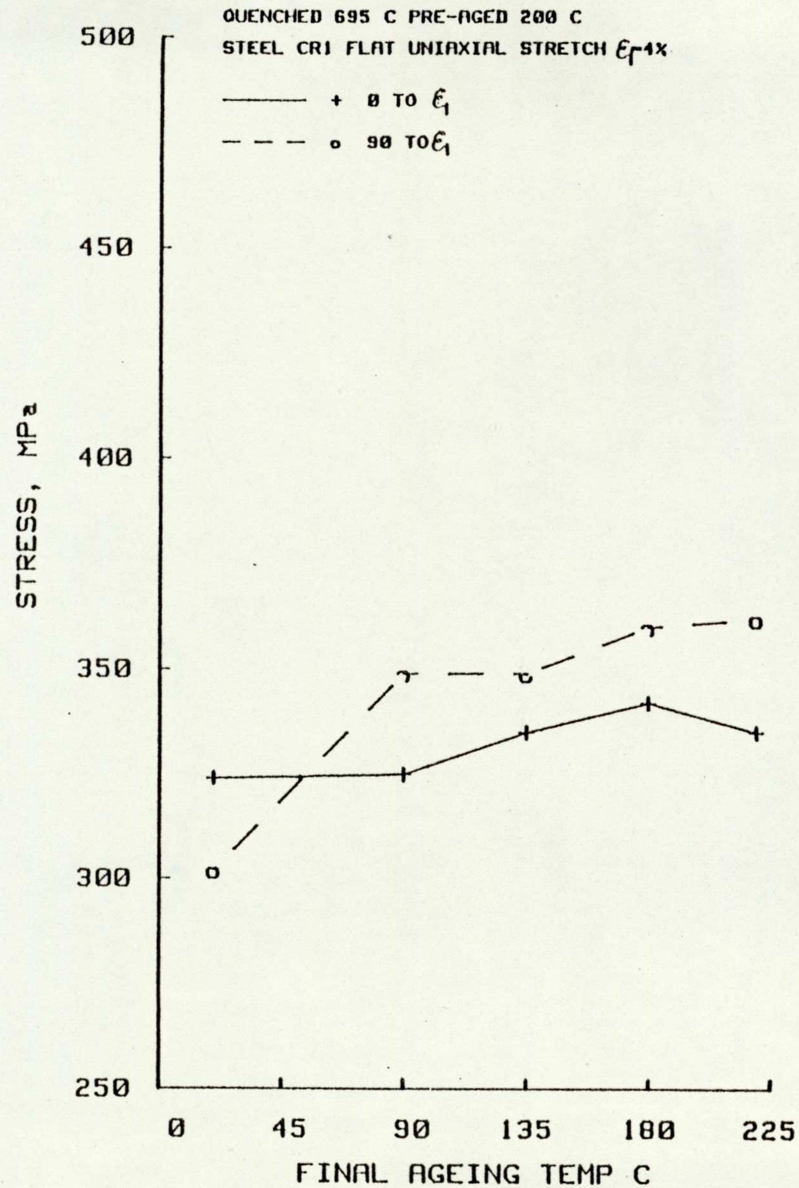
Fig 93

Figs 94 to 95.

Effects of variations in preageing temperature (for 3 minutes) and bake-hardening temperature (for 30 minutes) on 0.5% proof stress measured at 0° and 90° to direction of extension in prestraining. For steel CR1 stretched 4% in uniaxial tension before bake-hardening.

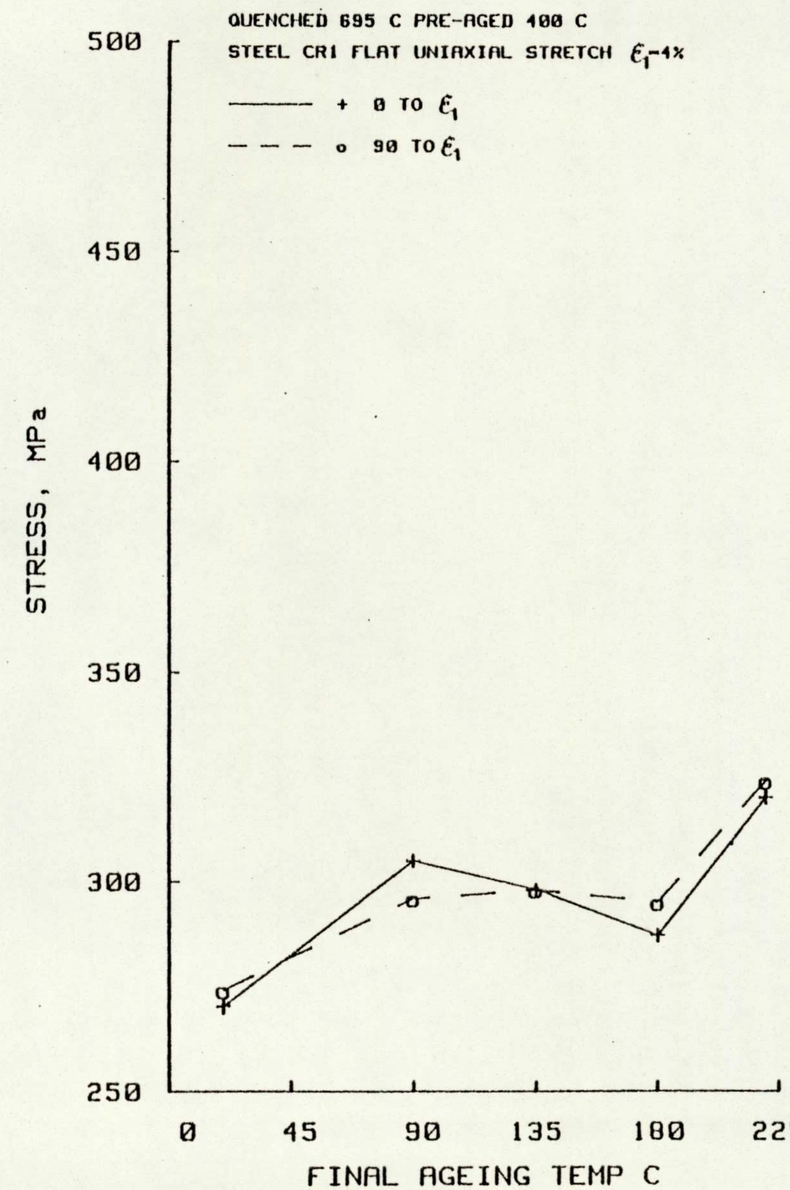
Fig 95 preaged at $200^{\circ}C$.

Fig 96 preaged at $400^{\circ}C$.



QUENCHED 695 C PRE-AGED 200 C
STEEL CR1 FLAT UNIAXIAL STRETCH ε_1-4%
+ 0 TO ε_1
o 90 TO ε_1
STRESS, MPa
500
450
400
350
300
250
0
45
90
135
180
225
FINAL AGEING TEMP C

Fig 94



QUENCHED 695 C PRE-AGED 400 C
STEEL CR1 FLAT UNIAXIAL STRETCH ε_1-4%
+ 0 TO ε_1
o 90 TO ε_1
STRESS, MPa
500
450
400
350
300
250
0
45
90
135
180
225
FINAL AGEING TEMP C

Fig 95

Table 1

Ratios of m values for curved and flat panels estimated from loads giving 0.2mm dents.

Ratios of Log P for flat panels/Log P for curved panels, where dent depth is 0.2mm and the radius of curvature is ~ 180mm. An effective prestrain of 4% was applied in each case.

Mode of prestrain and BH treatment	Dual-phase 60	Dual-phase 80	CR1
Uniaxial tension			
As-stretched	1.6	1.08	1.77
BH at 180°C	1.24	1.04	1.31
BH at 220°C	1.18	1.07	1.58
Mean values	1.34	1.06	1.55
Balanced biaxial tension			
As-stretched	1.29	1.23	1.69
BH at 180°C	1.24	1.10	1.52
BH at 220°C	1.20	1.14	1.70
Mean values	1.24	1.16	1.64

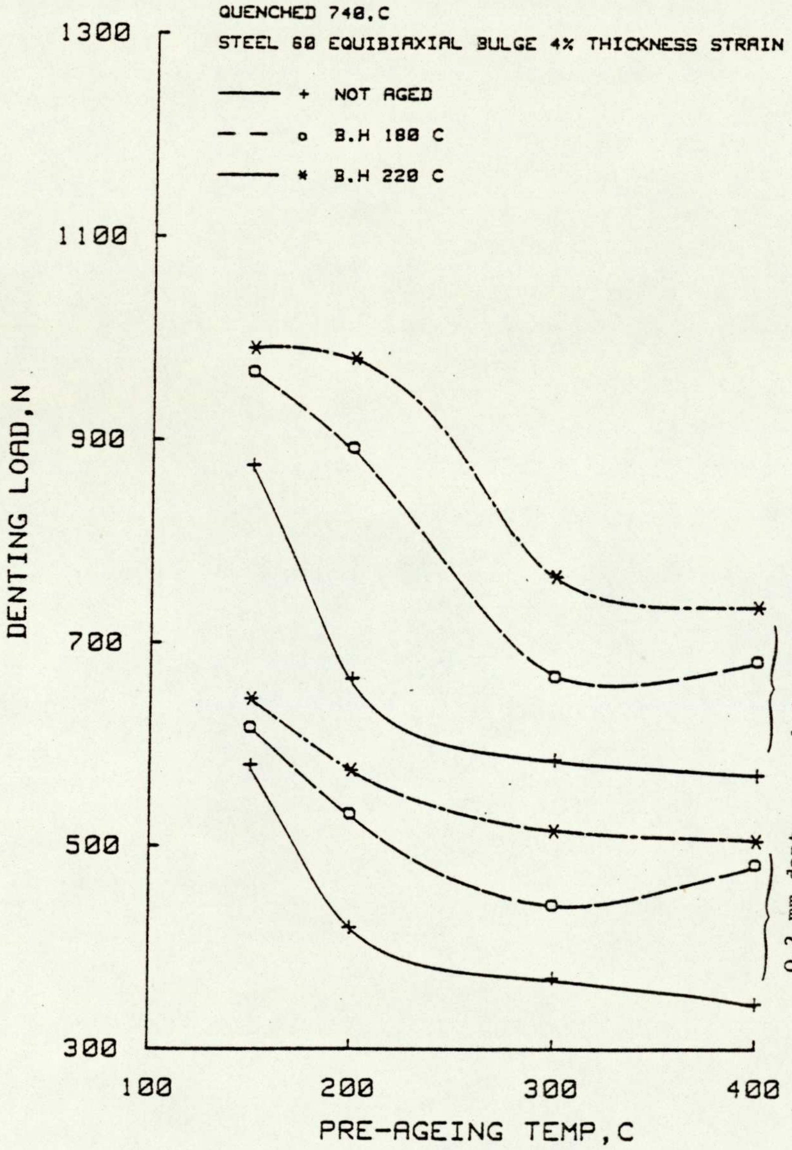


Fig 96. Effect of pre-ageing temperature on loads giving 0.2mm and 1.0mm deep dents in curved panels of dual-phase steel 60 equibiaxially stretched to 4% thickness strain with ~ 180mm radius of curvature. Tested as-stretched and after bake-hardening at 180°C and 220°C for 30 minutes.

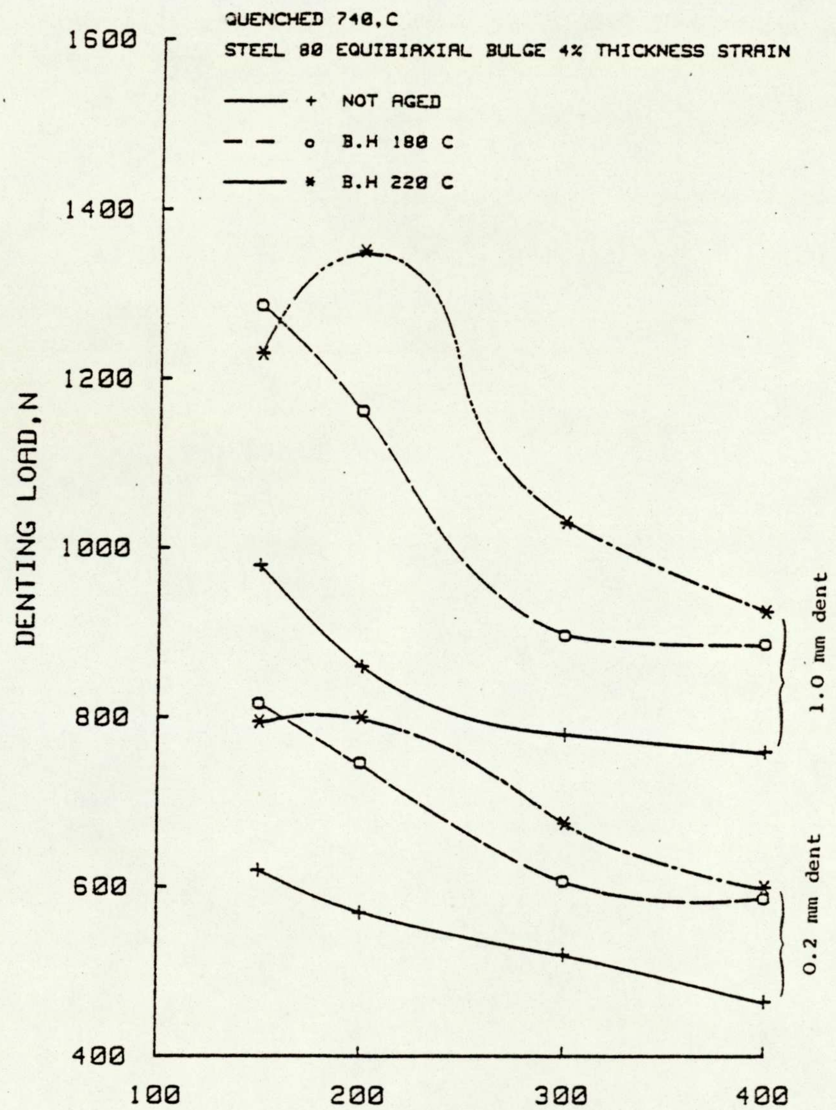


Fig 97. Effect of pre-ageing temperature on loads giving 0.2mm and 1.0mm deep dents in curved panels of dual-phase steel 80 equibiaxially stretched to 4% thickness strain with ~180mm radius of curvature. Tested as-stretched and after bake-hardening at 180°C and 220°C for 30 minutes.

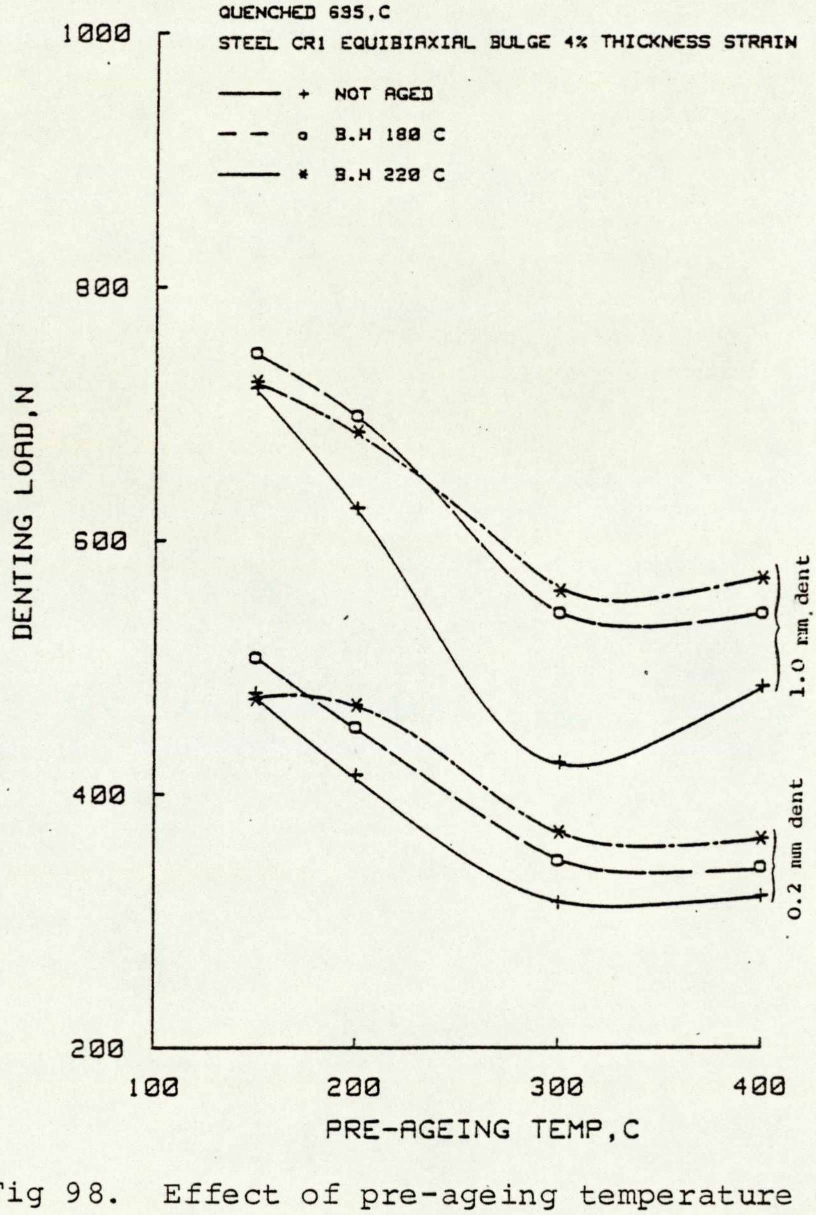


Fig 98. Effect of pre-ageing temperature on loads giving 0.2mm and 1.0mm deep dents in curved panels of steel CR1 quenched from 695°C, equibiaxially stretched to 4% thickness strain with ~180mm radius of curvature. Tested as-stretched and after bake-hardening at 180°C and 220°C for 30 minutes.

Table 2 Effect of bake-hardening on loads giving 0.2mm and 1.0mm deep dents for flat sheets prestrained in uniaxial tension to give $\mathcal{E}_1 = 0.04$.

Material and BH treatment	$P_{0.2}$ Load at 0.2mm dent N	Increase in $P_{0.2}$ N	% Increase at $P_{0.2}$ o/o	$P_{1.0}$ Load at 1.0mm dent N	Increase in $P_{1.00}$ N	% Increase at $P_{1.0}$
Dual-phase 60						
As-stretched	260			810		
BH at 180°C	420	160	61.5	1225	415	51.2
BH at 220°C	465	205	79	1350	540	66.7
Dual-phase 80						
As-stretched	420			1175		
BH at 180°C	595	175	41.7	1680	505	43
BH at 220°C	670	250	59.5	1825	650	55
Steel CR1						
As-stretched	175			505		
BH at 180°C	237	62		640	135	26.7
BH at 220°C	202	27		615	110	21.8

Table 3. Effect of bake-hardening on loads giving 0.2mm and 1.0mm deep dents for flat sheets prestrained with plane strain to give $\varepsilon_1 = 0.04$.

Material and BH treatment	$P_{0.2}$ Load at 0.2mm dent N	Increase in $P_{0.2}$ N	% Increase at $P_{0.2}$	$P_{1.0}$ Load at 1.0mm dent N	Increase in $P_{1.00}$ N	% Increase at $P_{1.0}$
Dual-phase 60						
As-stretched	278			738		
BH at 180°C	413	135	48.6	1088	350	47.4
BH at 220°C	475	197	70.8	1268	530	71.8
Dual-phase 80						
As-stretched	392			1047		
BH at 180°C	565	173	44	1420	373	35.6
BH at 220°C	590	198	50.5	1500	453	43.3
CR1 steel						
As-stretched	125			460		
BH at 180°C	200	75	60	600	140	30.4
BH at 220°C	175	50	40	535	75	16.3

Table 4 Effect of bake-hardening on loads giving 0.2mm and 1.0mm deep dents for sheets stretched in equibiaxial tension to a thickness strain of 4%.

Material and BH treatment	$P_{0.2}$ Load at 0.2mm dent N	Increase in $P_{0.2}$ N	% Increase at $P_{0.2}$	$P_{1.0}$ Load at 1.0mm dent N	Increase in $P_{1.00}$ N	% Increase at $P_{1.0}$
Dual-phase 60						
As-stretched	325			1090		
BH at 180°C	420	95	29	1315	225	20.6
BH at 220°C	450	125	38.5	1380	290	26.7
Dual-phase 80						
As-stretched	425			1390		
BH at 180°C	595	170	40	1790	400	28.8
BH at 220°C	645	220	51.7	1870	480	34.5
CR1 steel						
As-stretched	130			505		
BH at 180°C	170	40	30.7	615	110	21.8
BH at 220°C	200	70	35	680	175	34.8

Table 5 Effect of bake-hardening on loads giving 0.2mm and 1.0mm deep dents for curved panels stretched in equibiaxial tension to a thickness strain of 4%.

Material and BH treatment	$P_{0.2}$ Load at 0.2mm dent N	Increase in $P_{0.2}$ N	% Increase at $P_{0.2}$	$P_{1.0}$ Load at 1.0mm dent N	Increase in $P_{1.00}$ N	% Increase at $P_{1.0}$
Dual-phase 60						
As-stretched	460			720		
BH at 180°C	595	135	29.3	910	190	26.4
BH at 220°C	610	150	32.6	1025	305	42.4
Dual-phase 80						
As-stretched	595			912		
BH at 180°C	715	120	20	1115	203	22.2
BH at 220°C	825	230	38.6	1285	373	40.9
CR1 steel						
As-stretched	270			425		
BH at 180°C	310	40	14.8	465	40	9.4
BH at 220°C	325	55	20.4	485	65	14

Table 6 Effect of bake-hardening on 0.5% proof stress values for flat sheets prestrained in uniaxial tension to give $\varepsilon_1 = 0.04$

Material and BH treatment	$\sigma_{0.5}$ (MPa)		Increase in $\sigma_{0.5}$ MPa		Percentage increase in $\sigma_{0.5}$	
	0°	90°	0°	90°	0°	90°
Dual-phase 60						
As-stretched	422	387				
BH at $180^{\circ}C$	492	422	70	35	16.6	9
BH at $220^{\circ}C$	527	505	105	118	24.9	30
Dual-phase 80						
As-stretched	534	480				
BH at $180^{\circ}C$	605	553	71	73	13.3	15
BH at $220^{\circ}C$	645	635	111	155	20.8	32.3
Steel CR1						
As-stretched	270	263				
BH at $180^{\circ}C$	267	277	3	14	1.1	1.5
BH at $220^{\circ}C$	267	276	3	13	1.1	5

Table 7

Effect of bake-hardening on 0.5% proof stress values for sheets stretched in equibiaxial tension to a thickness strain of 4%.

Material and BH treatment	$\sigma_{0.5}$ (MPa)	Increase in $\sigma_{0.5}$ MPa	Percentage increase in $\sigma_{0.5}$
Dual-phase 60			
As stretched	444		
BH at 180°C	472	32	7.2
BH at 220°C	519	75	16.9
Dual-phase 80			
As stretched	549		
BH at 180°C	588	39	7.1
BH at 220°C	633	84	15.3
Steel CR1			
As stretched	318		
BH at 180°C	322	4	1.2
BH at 220°C	328	10	3.1

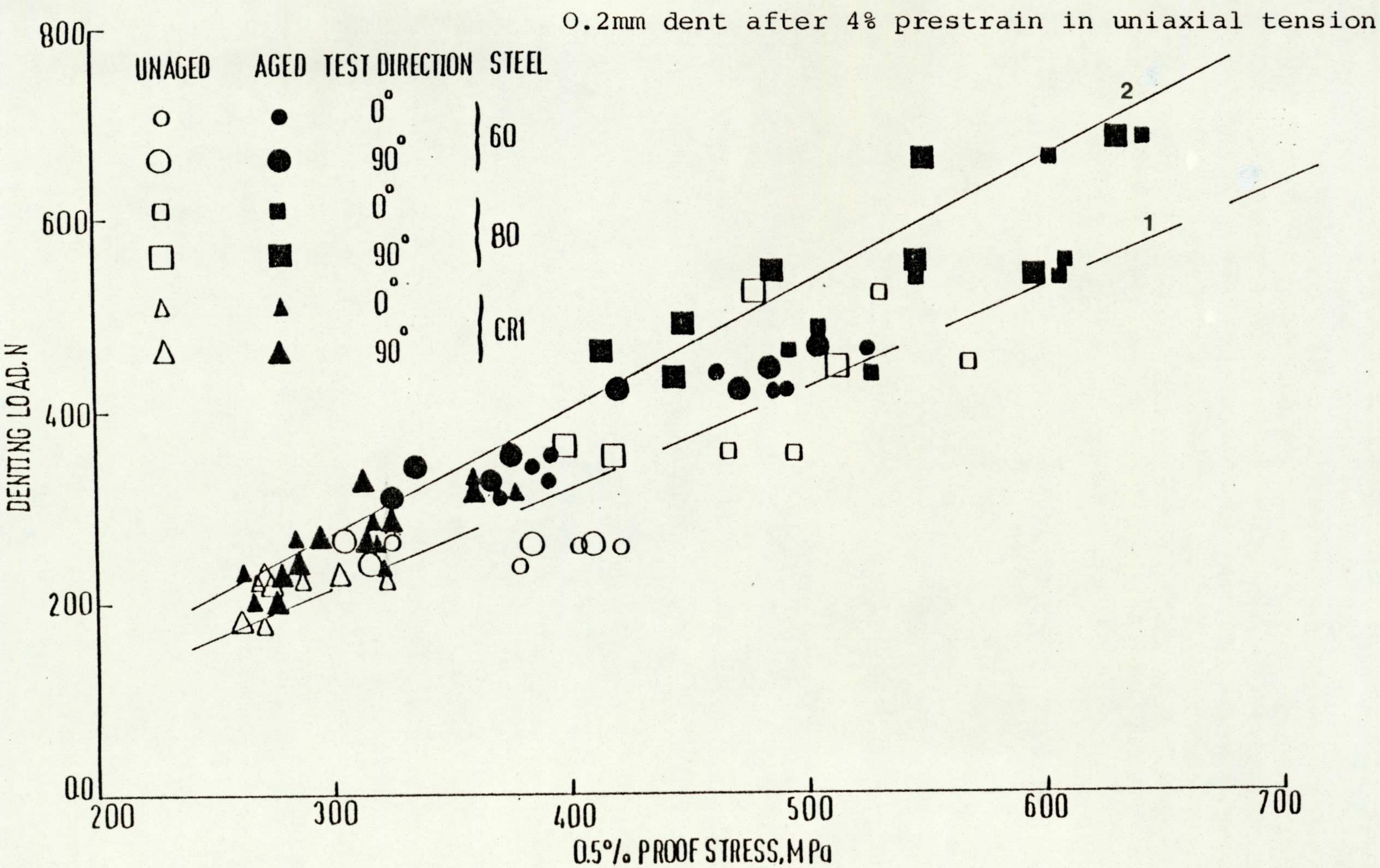


Fig 99 Comparison of loads for 0.2mm dents made in unaged and aged conditions after 4% strain in uniaxial tension. Line 1 is an estimate of the mean slope for tests in the unaged condition for the 0.5% proof stress measured at 0° to ε_1. Line 2 is a similar estimate of the mean slope of the relationship for conditions aged at $180^{\circ}C$ and $220^{\circ}C$ and with the 0.5% proof stress measured in both the 0° and 90° direction.

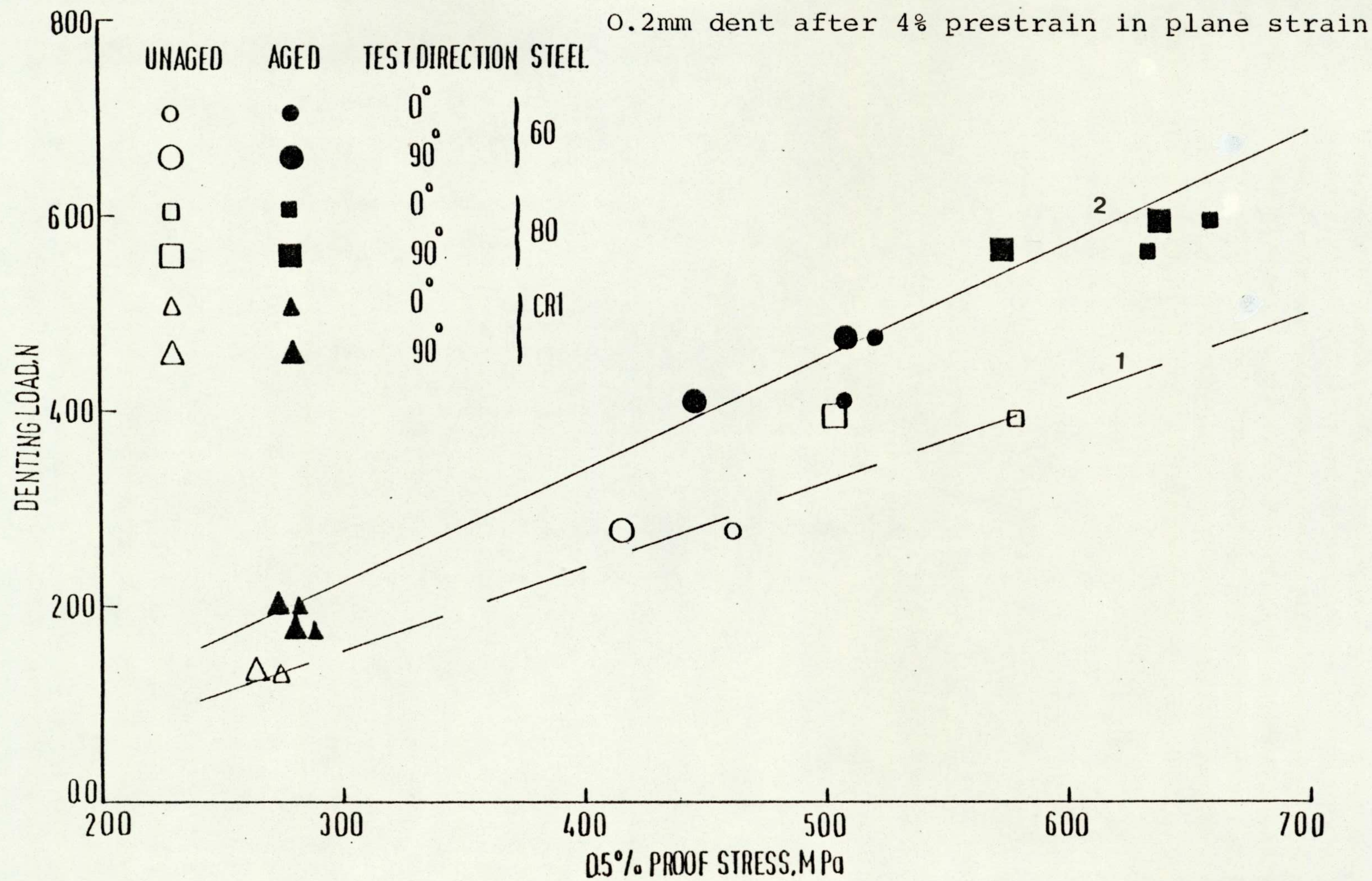


Fig 100 Comparison of loads for 0.2mm dents made in unaged and aged conditions after 4% strain in uniaxial tension. Line 1 is an estimate of the mean slope for tests in the unaged condition for the 0.5% proof stress measured at 0^{o} to $\mathcal{E}_1$. Line 2 is a similar estimate of the mean slope of the relationship for conditions aged at $180^{o}C$ and $220^{o}C$ and with the 0.5% proof stress measured in both the 0^{o} and 90^{o} direction.

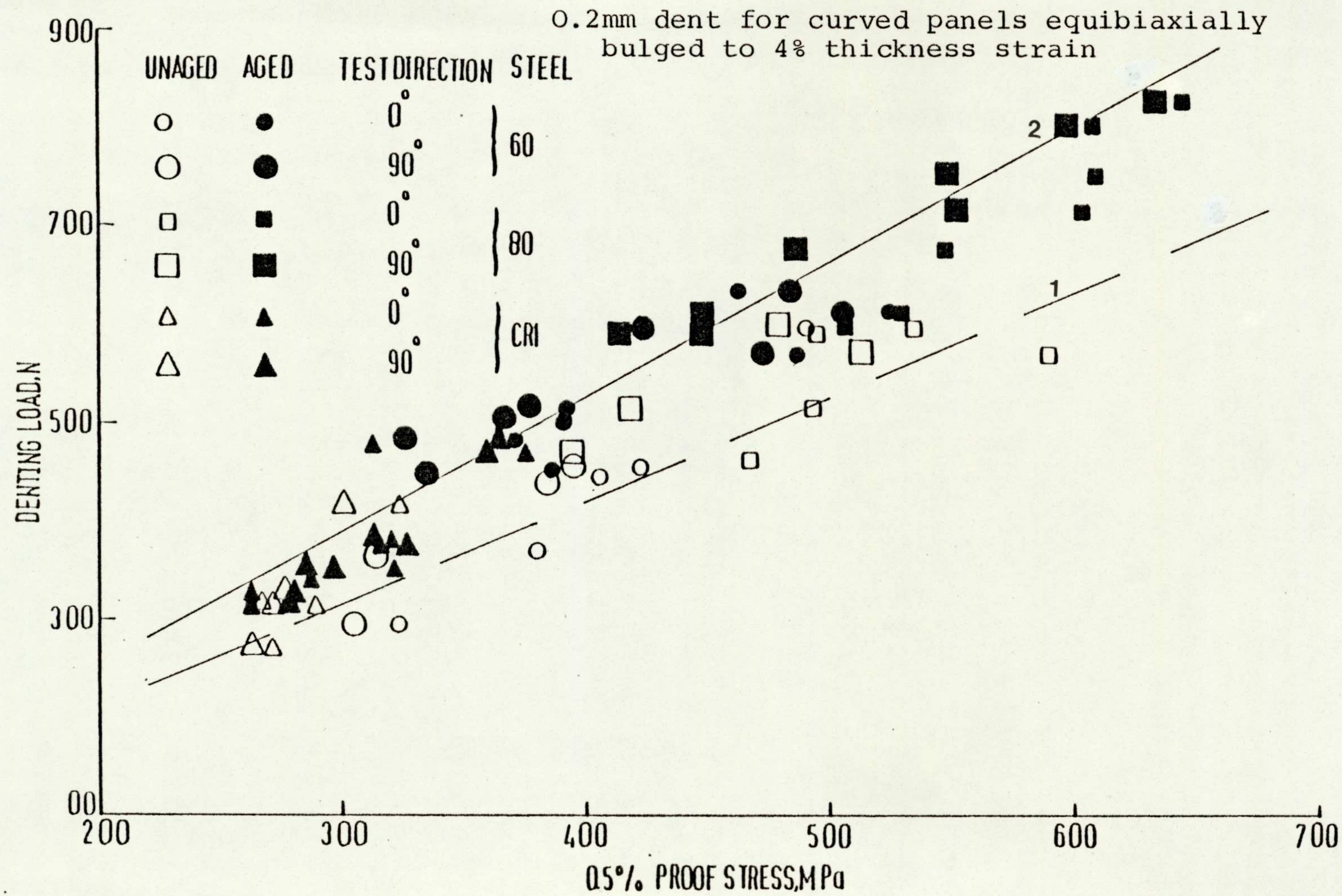


Fig 101 Comparison of loads for 0.2mm dents made in unaged and aged conditions after 4% strain in uniaxial tension. Line 1 is an estimate of the mean slope for tests in the unaged condition for the 0.5% proof stress measured at 0° to ε_1. Line 2 is a similar estimate of the mean slope of the relationship for conditions aged at 180°C and 220°C and with the 0.5% proof stress measured in both the 0° and 90° direction.

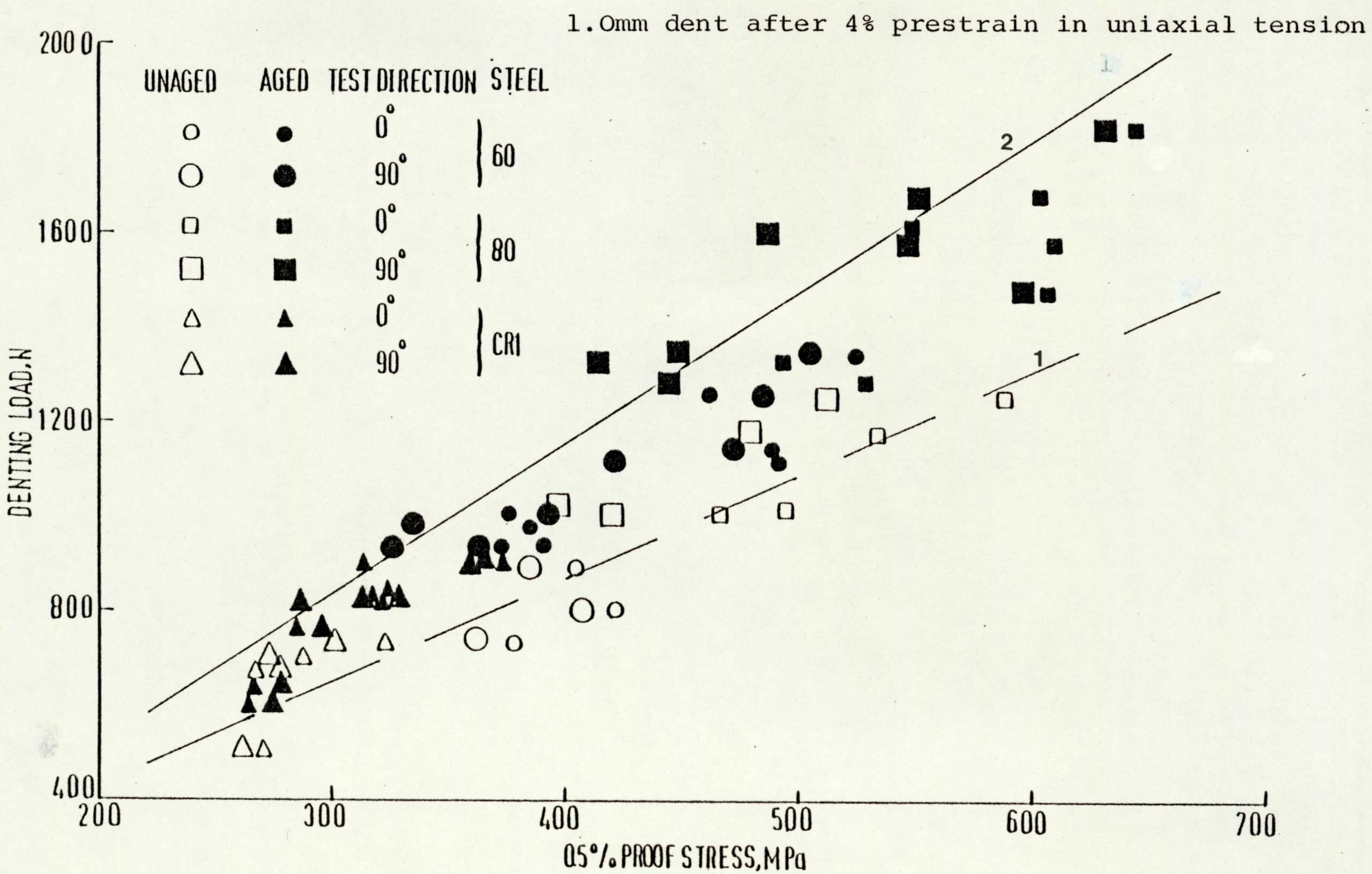


Fig 102 Comparison of loads for 1.0mm dents made in unaged and aged conditions after 4% strain in uniaxial tension. Line 1 is an estimate of the mean slope for tests in the unaged condition for the 0.5% proof stress measured at 0° to ε_1. Line 2 is a similar estimate of the mean slope of the relationship for conditions aged at 180°C and 220°C and with the 0.5% proof stress measured in both the 0° and 90° direction.

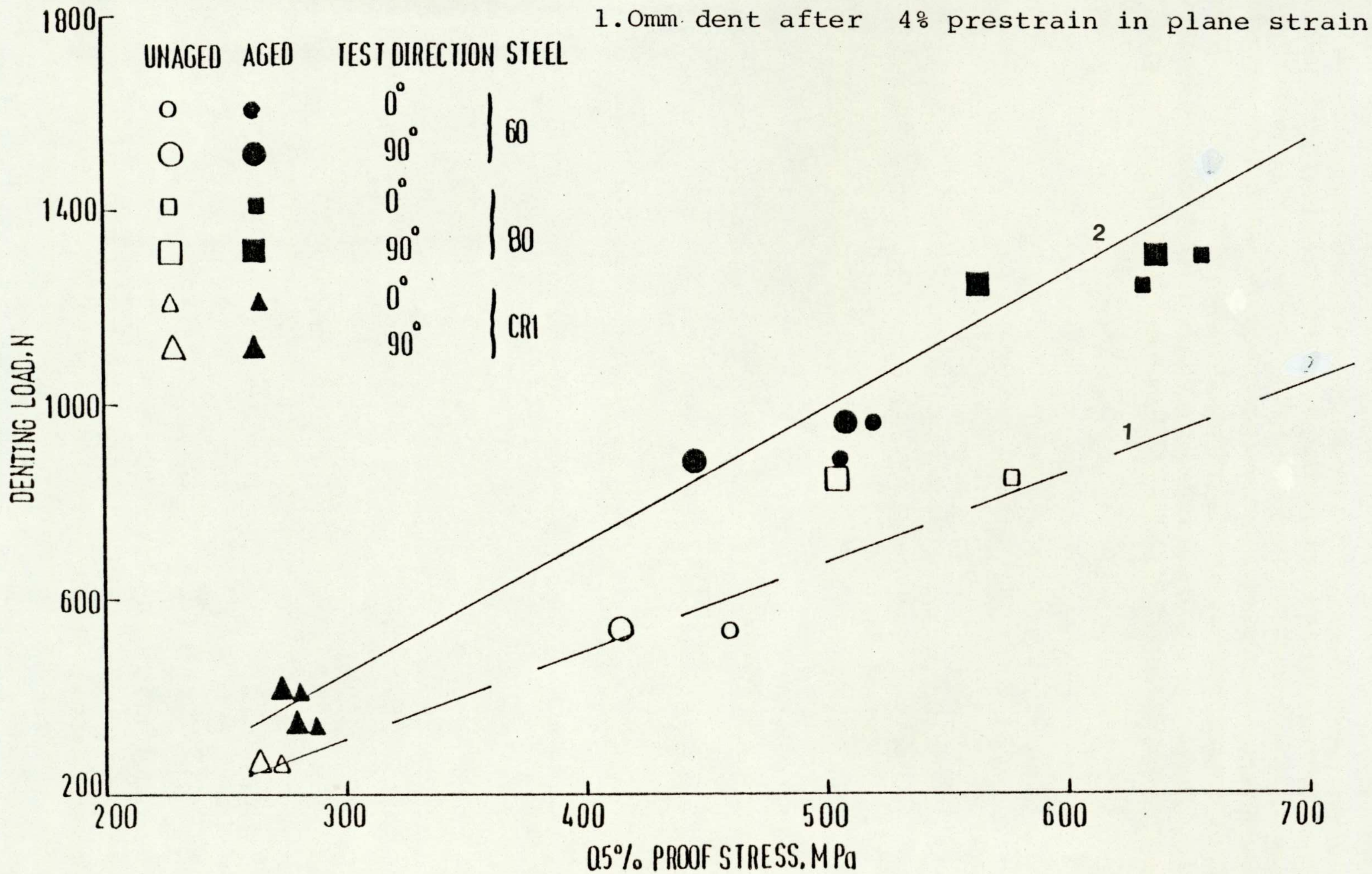


Fig 103 Comparison of loads for 1.0mm dents made in unaged and aged conditions after 4% strain in uniaxial tension. Line 1 is an estimate of the mean slope for tests in the unaged condition for the 0.5% proof stress measured at 0^o to ε_1. Line 2 is a similar estimate of the mean slope of the relationship for conditions aged at 180^oC and 220^oC and with the 0.5% proof stress measured in both the 0^o and 90^o direction.

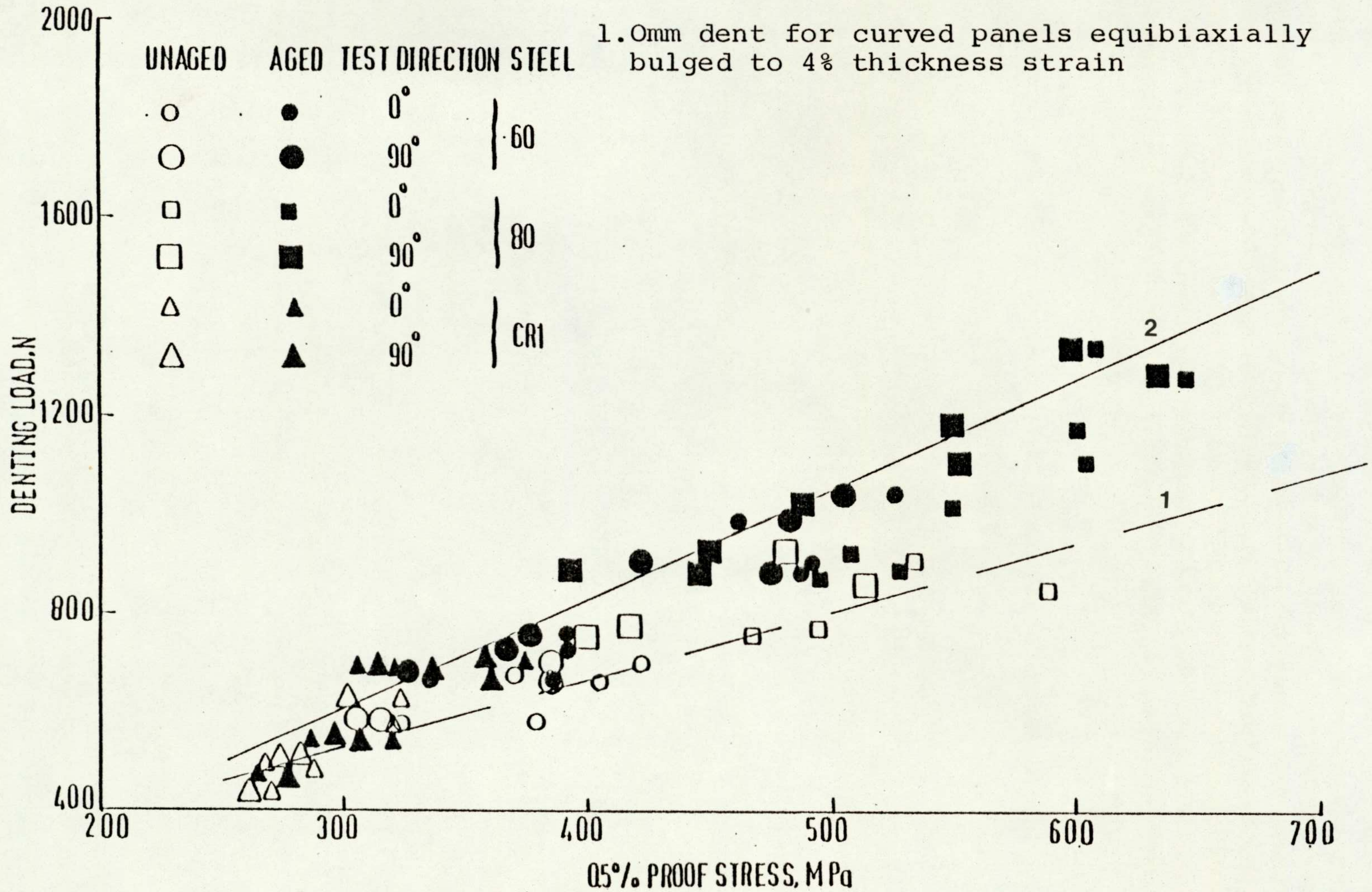


Fig 104 Comparison of loads for 1.Omm dents made in unaged and aged conditions after 4% strain in uniaxial tension. Line 1 is an estimate of the mean slope for tests in the unaged condition for the 0.5% proof stress measured at 0^o to ε_1. Line 2 is a similar estimate of the mean slope of the relationship for conditions aged at 180^oC and 220^oC and with the 0.5% proof stress measured in both the 0^o and 90^o direction.

CHAPTER 6

CONCLUSIONS

The following conclusions are based on tensile tests and static dent resistance tests made on the two lean alloy dual-phase steels designated 60 and 80 and ~~on~~ comparative On tests made on a batch annealed CR1 steel. The steels were pre-strained in uniaxial tension, plane strain and equibiaxial straining before bake-hardening (BH) for 30 minutes at various temperatures in the range $90^{\circ}C$-$220^{\circ}C$. Usually the effective strain applied in pre-straining was 4% but in some cases higher pre-strains were used and tests were also made after rolling reductions of 1% and about 4%.

Conclusions 1 to 3 relate to tests made in the 'as-received' commercially processed condition. In the case of dual-phase steels this involved water quenching from $740^{\circ}C$ followed by overageing for 1 minute at $260^{\circ}C$ and temper-rolling $\sim$ 1%.

1. In the as-stretched condition (before bake-hardening) the dual-phase steels which had been stretched in uniaxial tension or plane strain showed large differences (40 MPa) in the 0.5% proof stress values and smaller difference in UTS values measured at 0° and 90° to the direction of maximum extension. The batch annealed CR1 steel showed only relatively small differences in these flow stress values

(12 MPa) measured at 0° and 90° direction.

After 4% pre-strain in rolling the directionality of flow stress observed with the dual-phase steels was qualitatively similar to those obtained with pre-straining in uniaxial tension and plane strain but the directionality in the 0.5% proof stress was significantly smaller in the rolled pieces.

Results after pre-straining in balanced biaxial stretching were not directional but the differences between 0.5% proof stress and UTS values were closer to those seen in the 90° direction obtained in uniaxial tension or plane strain.

2. The differences in the directionality of observed with dual-phase steels and the CR1 steel after stretching in uniaxial tension or plane strain is believed to be related to internal stresses arising from the presence of hard second phase particles in dual-phase steels. Such directionality in the sheet plane was less in the rolled sheets than in unidirectionally stretched sheets because of the more complex deformation in rolling and it was absent after equibiaxial stretching.

3. Bake-hardening above about $135^{\circ}C$ caused increases in UTS and 0.5% proof stress values. After BH at $180^{\circ}C$ or $220^{\circ}C$ the increases were greater in the proof stress values, particularly those measured at 90° to the direction of maximum extension. Thus

directionality in the proof stress values in sheets stretched uniaxially and in plane strain was reduced by BH at 180°C and it was almost eliminated by BH at 220°C.

After equibiaxial stretching the increases in the 0.5% proof stress values after BH at 180° and 220°C were rather similar to those observed at 90° to the direction of maximum extension in sheets pre-strained in uniaxial stretching or plane strain.

4. In sheets re-heat treated, water quenched and pre-aged for 3 minutes at different temperatures in the range of 20°C-400°C, it was observed that pre-ageing at 300°C or 400°C caused a marked reduction in the improvement in 0.5% proof stress values given by BH at 180°C and 220°C. Pre-ageing at 200°C for 3 minutes gave bake-hardenabilities which were similar to those observed with the as-received sheets. With dual-phase sheets quenched from 740°C and not pre-aged before stretching the rises in the strength and reduction in directionality occurred at lower BH temperatures.

5. The results of tests on sheets with different strain ageing potentials outlined in conclusion 4 indicate that dislocation locking by interstitial atoms, mainly carbon, was the dominant factor promoting the increase in flow strength developed by BH treatments. Earlier work[34] suggests that during strain-

ageing at high BH temperatures, some of the carbon involved in the dislocation locking may be derived from resolution of fine precipitates.

6. Tensile tests made on sheets which had been pre-aged at various temperatures before temper-rolling about 1% and then strain-ageing at 47°C for 6 weeks showed negligible increases in the 0.5% proof stress and UTS values where the pre-ageing temperature was 200°C or higher. It was concluded that room temperature strain-ageing will be negligible provided the pre-ageing temperature is above 200°C.

7. Dislocation locking may not be the only factor responsible for the changes in directionality of flow strength developed in the dual-phase steels by the higher BH temperatures. Tempering of the micro-structure with some relief of the internal stresses may also be involved, particularly in the case of sheets which are pre-aged for a short time at relatively low temperatures. With sheets which are stretched in this condition, BH for 30 minutes at temperatures of 180°C or 220°C probably causes some coarsening of the fine carbide precipitates and further tempering of the martensite.

8. In all conditions examined, bake-hardening at 180°C or 220°C gave substantial increases in static dent resistance. With the dual-phase steels, the absolute increases in denting load for a given depth

of dent and bake-hardening treatment were, to a first approximation, rather similar for dents made with the four different conditions of pre-straining (flat uniaxial stretching, flat plane strain, flat equi-biaxial stretching and equibiaxial bulging).

9. The loads required to form small dents in flat panels using a tool of 50mm diameter were quite small even with the dual-phase steels, for example, in steel 60 prestrained 4% with a 0.5% proof stress value of about 460 MPa. The load giving a 0.2mm deep dent, $P_{0.2}$, was 260 N. Bake-hardening for 30 minutes at 220°C gave about 80% increase in this denting load.

10. With the bulged panels of $\sim$ 180mm radius of curvature, $P_{0.2}$ were substantially higher than in the case of flat panels. Thus the proportionate increases in denting load developed by a given bake-hardening treatment was less in the curved panels than in the flat panels. For example, with the dual-phase steel 60 bulged to give a thickness strain of 4%, the percentage increase in $P_{0.2}$ caused by bake-hardening was $\sim$ 30%.

11. The increments in dent resistance caused by bake-hardening at 220°C were significantly higher than those caused by bake-hardening at 180°C. However, while the mean ratio of the increments with these two temperatures, increment with 220°C/ increment with 180°C, was only about 1.2 with flat panels, it was

about 1.6 with the curved panels. The difference may arise from differences in the stress systems developed during denting of the two forms of panels.

12. In all cases the proportionate increases in dent resistance were greater than would be predicted from observed increases in the 0.5% proof stress values in tensile tests. In terms of the relationship proposed by Miyahara[73] which assumes that dent resistance with a given thickness of sheet is directly proportional to its yield stress, the dent resistance of the dual-phase steels observed at a given value of the 0.5% proof stress tended to be higher in the bake-hardened condition than in the as-stretched conditions. This was particularly true with the higher strength steel 80.

13. Tests made with re-heat treated sheets, pre-aged at different temperatures to give different strain ageing potentials, showed that the expected trend of higher increments in dent resistance with higher ageing potentials for a given bake-hardening treatment. However, even with relatively high pre-ageing treatments (e.g. 3 minutes at 300^{o}C or 400^{o}C) the increments in dent resistance obtained in bake-hardening at 180^{o}C or 220^{o}C were substantial. Also the batch annealed CR1 steel showed significant increases in dent resistance after these bake-hardening treatments.

These results would not be predicted from measurements of changes in the 0.5% proof stress in tensile tests.

14. The observation that bake-hardening gave significant improvements in dent resistance with the conditions which are expected to be of low ageing potential probably indicates that even light dislocation locking can give useful increases in dent resistance. The influence of such light locking on the effective elastic limit developed in denting, which involves compressive components of stress, is not necessarily shown in the 0.5% proof stress values measured in tensile tests. It is possible that proof stress values measured at a very low offset (e.g. 0.1%) at 90° to the direction of maximum extension in prestraining would give better agreement but there are considerable difficulties in obtaining reliable 0.1% proof stress values for the unaged condition in testpieces which have to be cut from large panels.

15. The results summarised in conclusions 12 to 14 indicate that measurements of the yield strength made in tensile tests will not in general provide a reliable quantitative prediction of increments in dent resistance generated by bake-hardening and, whenever practicable, actual dent resistance tests should be made.

Abbreviations Used:

AIME	American Institute of Mining, Metallurgical and Petroleum Engineers
Conf	Conference
Eng	Engineering
Inds	Industries
Inst	Institute
ISI	Iron and Steel Institute
J	Journal of
Mat	Material
Met	Metal, Metallurgy
NPL	National Physical Laboratory
Phs	Physics
Proc	Proceedings of
Ref	Reference
Rep	Report
Roy	Royal
SAE	Society of Automotive Engineers
Sci	Science
Soc	Society
Spe	Special
TMS/AIME	The Metallurgical Society of AIME
Trans	Transaction of
Trans ISI J	Transaction of Iron and Steel Institute of Japan

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