

SUPERPLASTIC DEFORMATION OF TITANIUM ALLOYS

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SUMMARY

The superplastic deformation of three alpha/beta titanium alloys; Ti-6Al-4V, IMI 550 (Ti-4Al-4Mo-2Sn-0.5Si) and Ti-8Al-1Mo-1V and a beta alloy; Ti-15V-3Cr-3Al-3Sn was investigated. The alpha/beta alloys exhibited high strain rate sensitivities and superplastic strain in these alloys tended to randomise the alpha and beta phase textures and caused grain growth. The beta alloy showed relatively low strain rate sensitivity and formed subgrains resulting in grain refinement.

Superplastic deformation reduced room temperature strength. In the alpha/beta alloys this was mainly the result of recrystallisation at the forming temperature rather than grain growth associated with the superplastic strain. The strength of the IMI 550 alloy after forming was raised by increasing the cooling rate from the forming temperature and by ageing after forming.

The anisotropy of superplastic deformation, measured in terms of the strain ratio R , was caused by microstructural directionality and was not related texture. The R values were also influenced by the test piece shape.

Uniaxial data were used to predict the optimum gas pressure cycle for the superplastic forming of a hemisphere from Ti-6Al-4V sheet. The calculated pressure cycle was found to be significantly different to that for an isotropic, non-strain hardening material.

The uniaxial data were also used in a computer model of necking in Ti-6Al-4V during superplastic deformation. The influence of strain hardening and strain rate hardening were considered and predictions of limiting strains for various initial neck sizes were made.

The work described in this thesis has shown in particular that:

- 1 The anisotropy of superplastic deformation in titanium alloys results from directionality in the microstructure, but that R value measurements can be influenced by test piece shape.
 - 2 The alpha/beta alloys investigated exhibit relatively high strain rate sensitivities and are more suitable for superplastic forming than the beta alloy Ti-15V-3Cr-3Al-3Sn.
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The term "superplastic" indicates the ability of a material to undergo large amounts of essentially neck free strain without failure. Superplasticity is generally restricted to temperatures greater than $0.4T_M$ (1), where T_M is the absolute melting temperature, and is promoted by a small grain size and slow grain coarsening at the deformation temperature. Superplasticity can also be induced in some materials by cycling through a phase transformation (1), although this "environmental superplasticity" is not the subject of this thesis.

The logarithmic stress/strain curve for the superplastic temperature usually has a sigmoidal shape and is conventionally divided into three regions. In regions I and III, the low and high strain rate regimes respectively, flow stress is relatively insensitive to strain rate. However, in region II, where superplasticity is encountered, the flow stress often increases rapidly with strain rate. Thus, any tendency of strain localisation during deformation is counteracted by a local increase in flow stress and necking is avoided. The potential superplasticity is usually measured by the strain rate sensitivity m where: $\sigma = K\dot{\epsilon}^m$. In this equation σ is the applied stress, $\dot{\epsilon}$ is the strain rate and K is a constant depending on testing conditions. This expression incorporates no strain hardening and an m value of 1.0 indicates viscous flow (1). In practice, m values for metals up to about 0.9 can be obtained and some strain hardening, due to grain growth, may occur.

The two phase Ti-6Al-4V alloy is by far the most widely used high strength titanium alloy and is highly superplastic, capable of undergoing tensile elongations of up to about 1000% (1,2). Consequently, the experimental data on superplasticity of titanium is most extensive for this alloy and it has been used for almost all of the superplastically formed titanium components so far produced.

The objectives of this thesis were to examine the characteristics of superplastic deformation of titanium alloys with emphasis on the influences of microstructure, texture and test piece shape on the uniformity of superplastic flow. In a review of the literature on superplastic deformation particular attention has been paid to those aspects most relevant to the experimental work described elsewhere in the thesis.

An inert atmosphere testing rig was designed and uniaxial superplastic testing was carried out on Ti-6Al-4V in sheet form and on sheet

test pieces machined from Ti-6Al-4V bar containing a strongly directional microstructure and a pronounced crystallographic texture. Three other sheet alloys, for which there was very little data on superplastic deformation were also investigated viz:

(1) IMI 550 (Ti-4Al-4Mo-2Sn-0.5Si). This is a high strength, two phase alloy with potential for lower forming temperatures and significantly higher post-formed room temperature strength than Ti-6Al-4V.

(2) Ti-8Al-1Mo-1V. This is a creep resistant, near alpha alloy with a higher modulus and lower density than most other titanium alloys.

(3) Ti-15V-3Cr-3Al-3Sn. This is a large grained, metastable beta alloy that is cold formable in the annealed condition and can be aged to very high strengths.

Biaxial sheet forming under gas pressure is often carried out assuming that the material is isotropic and does not strain harden. The superplastic data derived from the uniaxial tests on Ti-6Al-4V have been used to estimate the effects of anisotropy, strain hardening and strain rate sensitivity on the superplastic forming of a hemisphere and to predict the optimum pressure cycle. The uniaxial data has also been used to model the necking characteristics of Ti-6Al-4V during superplastic tensile deformation.

The experimental results are discussed in detail. The superplastic deformation in each of the alloys is compared and the alloys are assessed for superplastic formability. The causes of anisotropy of superplastic deformation are considered. Conclusions arising from the work are presented and several areas of possible future work are suggested.

2.1 Models of superplastic deformation

A variety of mechanisms have been proposed for superplastic deformation including slip, diffusion and dislocation creep and grain boundary sliding with various accommodation processes. There is evidence to support many of the models. However, no single theory comes close to explaining all of the experimental observations, which is not surprising in view of the wide range of single and multiphase materials exhibiting superplasticity (1-4).

Diffusion creep processes may involve lattice diffusion (Herring-Nabarro (5) creep, grain boundary diffusion (Coble (6) creep) or a combination of diffusion paths (7). Both Herring-Nabarro and Coble creep predict a strain rate sensitivity $m = 1$, where $m = d \ln \dot{\epsilon} / d \ln \dot{\epsilon}$, but a different grain size dependence of strain rate. Diffusion creep also predicts grain elongation (not usually observed during superplastic deformation), strain rates usually several orders of magnitude too low and the retention of texture after superplastic strain. However, there is some evidence that diffusion creep plays a role in the superplastic deformation of some materials (2) and Griffiths and Hammond (8) predicted the superplastic strain rates of two large grained beta titanium alloys and beta brass with reasonable accuracy using the Herring-Nabarro creep formula.

Models based on slip alone are not satisfactory (2), because they do not predict an equiaxed grain structure after superplastic strain and cannot explain the randomisation of texture or the low rate of work hardening. There is evidence (see section 2.3) that slip can play a role in the superplastic deformation of some alloys. However, in all these cases recovery, recrystallisation or grain boundary migration are assumed to take place. According to Naziri and Pearce (9,10) and Schmidt-Whitley (11,12) superplasticity occurs only when the observed flow stress would predict a dislocation cell size greater than the grain size. Thus, subgrains are not observed during superplastic deformation. Hayden et al (13,14) formulated a dislocation climb model based on experimental observations of superplasticity in Ni-Cr-Fe alloys in which a transition occurred from essentially dislocation free grains at low stress to cell formation and dislocation tangling at higher stresses.

Chaudhari (15) produced a model of superplastic flow in Zn-Al based on the motion of dislocations in internal stress fields and later (16) suggested a "dislocation cascade" mechanism involving diffusion creep at the heads of dislocation pile-ups. However, the extensive dislocation pile-ups predicted by both the models of Chaudhari (15,16) are not usually observed (1,2) although studies of the role of dislocations in superplastic deformation are likely to be hindered by dislocation recovery and disappearance during cooling from the forming temperatures.

The dislocation based models discussed generally seem (1) to be more applicable to the high stress regime ie region III than to region II where the highest strain rate sensitivities are exhibited.

It is grain boundary sliding that is usually assumed to play the dominant role in superplastic deformation. Observations of marker lines scratched onto the surface have revealed grain boundary offsets after superplastic strain in eutectoid Al-Zn (17-19), eutectic Pb-62%Sn (20), eutectic Mg-Al (21) and Pb-Tl (22). The grain rearrangement in the Pb-Sn eutectic has also been studied in the scanning electron microscope by straining in situ (23,24). It has been estimated (1,2,25-27) that grain boundary sliding contributes 60-80% of the total strain in region II and considerably less in regions I and III.

Grain boundary sliding cannot occur without some accommodation process to maintain coherency between grains at edges and triple points. The model of Ashby and Verrall (28) proposed diffusional accommodation for grain boundary sliding and introduced the "grain switching" concept, whereby a group of four grains rearrange themselves producing an axial strain of 0.55. The model predicts a strain rate of:

$$\dot{\epsilon} = \frac{100\Omega}{kTd^2} \left[\sigma - \frac{0.72\gamma}{d} \right] D_v \left[1 + \frac{3.3\delta D_B}{dD_v} \right]$$

where σ is the tensile stress, Ω the atomic volume, d the grain size, γ the grain boundary free energy, δ the width of the boundary diffusion path, D_v and D_B are the bulk and boundary diffusion coefficients, k is Boltzmann's constant and T the absolute temperature. The threshold stress $0.72\gamma/d$ arises from the high energy transition state in the grain switching process. The contribution of dislocation creep, which becomes significant at high stress levels, was considered to be additive. The Ashby/Verrall model has been applied with varying degrees of success to

a Zn-Al alloy (28,29), alpha/beta titanium (30) and a Zr based alloy (31), and the grain switching process has been observed in practice (32). However, the grain size dependence given by the model is not consistent with many of the experimental data (33-35), it predicts a stress exponent ($n = 1/m$) tending to unity and it is based on a two dimensional array of grains, which can deform without any increase in surface area of the specimen.

A three dimensional grain boundary sliding model with diffusional accommodation at grain corners has been proposed by Geckinli (36). The predicted strain rate is:

$$\dot{\epsilon} = \frac{1.25D_B \delta \Omega d^3}{K_V^2 kT} \left(\sigma - \frac{22.7\gamma}{d} \right)$$

where K_V is the volume of matter diffused for each grain for sliding to occur and the other variables are as defined above. Since $K_V \propto d^3$ this predicts a d^{-3} grain size dependence of strain rate, which agrees with results on Pb-Sn eutectic (23) and Sn-5%Bi (37). However, the stress dependence of strain rate $\dot{\epsilon} \propto (\sigma - \sigma_0)$, which gives $m = 1$ is not commonly observed.

Gifkins (38) suggested a "core and mantle" model, whereby each grain consists of a core region surrounded by a deformable mantle of variable width. At low stress (region I) diffusion creep takes place with diffusion confined to a relatively narrow mantle. In region II the mantle deforms by dislocation climb and glide and at higher stress extensive slip occurs in the core and subgrains may form. This model was subsequently extended (22) to three dimensions to allow grains from one layer to slide between two others in an adjacent layer. Good agreement with experiment was claimed (22).

Beere (39) proposed a deforming mantle model for cubic grains based on diffusion creep. It was shown (39) that if one set of interfaces has a greater sliding resistance than another then grain rotation is inevitable.

Ball and Hutchison (19) working on eutectoid Zn-Al suggested that groups of grains slide as units until they are blocked by a grain of unfavourable orientation. Dislocations generated in this grain pile up at the opposite grain boundary and eventually climb into the boundary itself.

This led to the rate equation:

$$\dot{\epsilon} = \frac{AD_B G}{kT} \left(\frac{\sigma}{G} \right)^2 \left(\frac{b}{d} \right)^2$$

where A is a constant.

According to Mukherjee (40) dislocations are generated by grains sliding individually rather than in groups and encountering ledges or other protrusions in adjacent grains. The rate of sliding is then controlled by the climb rate of dislocations into the boundary. This leads to a rate equation similar to that of Ball and Hutchison given above and gives a reasonable correlation with data on Zn-Al eutectoid (40).

Gittus (41) proposed a theory of superplastic flow involving grain switching specifically for two phase materials. According to the model sliding occurs by dislocations gliding in the interphase boundaries and piling up at triple edges before climbing away into disordered segments of the interphase boundary.

2.2 Superplastic deformation of two phase materials

Models of superplastic deformation in two phase alloys are likely to be complicated by quite different properties in the two phases and the presence of interphase as well as intergranular boundaries. Titanium alloys are likely to represent an extreme case in view of the relatively high lattice diffusion rates in the beta phase; $D_\beta/D_\alpha = 100-1000$ in the superplastic temperature range (42). Hamilton et al (30) produced, by hydrogenation, phase proportions from 40-100% beta in Ti-6Al-4V at 870°C and found that flow stress at 870°C increased with decreasing beta phase content, although this comparison was complicated by rapid grain growth in the near beta alloys at low strain rates, which tended to increase the flow stress. The highest flow stresses were recorded for an alpha alloy Ti-6Al. Similarly, Sastry et al (43) measured flow stresses at 900°C in an alpha titanium alloy (Ti-5Al-2.5Sn), a beta alloy (Ti-15V-3Cr-3Al-3Sn) and an alpha/beta alloy (Ti-6Al-4V). The alpha phase alloy was found to have the highest flow stress and the alpha/beta alloy the lowest, although the results were again influenced by excessive grain growth in the single phase alloys.

The superplastic properties of each phase can be studied separately using single phase alpha or beta alloys as indicated above. However, the problem remains of combining the results to predict the behaviour in two phase materials. Hamilton et al (30) stated that the Ashby/Verrall model (28) of the diffusion accommodated grain boundary sliding could be used to calculate the $\sigma/\dot{\epsilon}$ relationship for Ti-6Al-4V from the data for the single phase materials using a rule of mixtures approach and by assuming the same strain rate in each phase. According to this "isostrain rate" model the strain rate in the alloy is essentially determined by the deformation characteristics of the harder (alpha) phase. However, this is not consistent with other work on titanium alloys (44-46), alpha/beta brass (47,48) and the zirconium based "Zircaloy" (31), where it was suggested that the deformation is largely restricted to the beta phase, which tends to form a continuous matrix and behaves like a deforming mantle around the alpha grains. This type of behaviour is better described by the term "isostress", with the two phases deforming at different rates. Similarly, Springarn and Nix (49) proposed a model in which the faster diffusing beta phase collects at alpha/alpha boundaries and pinches off the alpha grains as observed by Naziri et al (50) in Zn-22%Al and by Hidalgo-Prada and Mukherjee (46) in Ti-6Al-4V-2Ni.

Boundary sliding measurements on a Pb-62%Sn alloy (20) and on Zn-22%Al (17) both revealed that the sliding rate for each type of interface depended on the value of δD_g , where δ is the boundary width and D_g is the coefficient of boundary diffusion. This was the basis for a deformation model (51) for anisotropic superplasticity in two phase alloys, which accounted for the break-up of banded microstructures and predicted a flow softening by the progressive and irreversible conversion of alpha/alpha interfaces into beta/beta interfaces.

Dunlop et al (52) observed that cavitation in an aluminium bronze (Cu-9.5Al-4Fe) occurred mainly at alpha/beta boundaries and suggested that sliding was concentrated at these boundaries as a result of the strong texture, which indicated a relatively low misorientation across alpha/alpha and beta/beta boundaries with a correspondingly high sliding resistance. Similarly, Chandra et al (53) found that the sliding rate in alpha/beta brass decreased in the order $\alpha/\beta > \alpha/\alpha > \beta/\beta$ and that cavitation occurred predominantly at alpha/beta interfaces. It was concluded that

cavitation was inhibited in the beta/beta boundaries by plastic flow in the relatively soft beta phase and Patterson and Ridley (54) showed that void content decreased with increasing beta phase proportion at 600°C in alpha/beta brasses with slightly differing compositions.

Attempts have been made (44-46,55,56) to improve the superplastic properties of Ti-6Al-4V by small additions of Ni, Co or Fe to increase the beta phase proportion and to increase effective diffusion rates in the beta phase. Both Ni and Co have tracer diffusivities about two orders of magnitude higher than that of titanium in beta titanium and it is reasoned (57) that, because gross solute segregation can be avoided by migration of the phase interfaces, the diffusivity can be controlled by the faster diffusing species even at concentrations of only a few percent, the effective diffusion rate in a two component system obeying: $D_{\text{eff}} = C_B D_A + C_A D_B$, where C_A and C_B are the atomic fractions of components A and B and D_A and D_B are the tracer diffusivities. The modified alloys were shown to have similar properties to the base alloy, but at temperatures typically 100°C lower. Some improvement in room temperature strength due to solid solution hardening was also recorded (55). It was shown (44) that the fast diffusing species ie Co or Ni tended to segregate to the boundaries perpendicular to the tensile axis during superplastic strain and that longitudinal boundaries were depleted. According to Ma et al (58) the reverse is true for the relatively slow diffusing element Mo in IMI 550.

2.3 Superplastic anisotropy

Anisotropic superplastic properties have been reported for a wide range of alloys (1,2,51,59-79) and have generally been ascribed either to the effects of pre-existing crystallographic texture, inferring slip controlled deformation, or to mechanical fibreing or grain elongation, in which case grain boundary sliding is considered to be dominant. There is substantial evidence for both, although in many cases either texture or microstructure are discussed in isolation and the results are not completely unambiguous.

In general (1,2,25,32), both microstructural directionality and texture intensity are reduced by superplastic strain, although when substantial amounts of slip occur the texture change may involve a systematic

rotation (69) or the stabilisation of certain texture components (76,80-82) and in two phase alloys the effect of superplastic strain on texture may be different in each phase (32,75,80-83). However, even when superplastic anisotropy appears to be texture related, it is not suggested that slip is the only deformation mode in operation. Superplastic deformation does not result in grain elongation normally associated with slip at relatively low temperatures and high strain rates. To take account of this, additional processes such as recrystallisation, recovery or grain boundary migration are assumed to take place. It is usually accepted (75,76,78,80-82) that texture randomisation is associated with grain boundary sliding, whereas the stabilisation or even intensification of some texture components may be due to slip (76,81,82,84-86) or anisotropic diffusion (80).

Packer et al (66) working on eutectic Zn-Al noticed that round tensile specimens developed elliptical cross-sections during superplastic strain. This was accompanied by both a reduction in texture intensity and the microstructure becoming more equiaxed. However, the elliptical specimens were re-machined to a round cross-section and were still found to behave anisotropically after the microstructural directionality had been removed. It was concluded that the strain anisotropy was the result of texture in the zinc rich phase and that the deformation was occurring largely by slip, although concurrent grain boundary migration or recrystallisation was invoked to account for the fact that grain elongation did not occur. Similarly Johnson et al (67) produced spheroidised grains in both eutectic and eutectoid Zn-Al by hot rolling, which also resulted in a strong texture. This material behaved anisotropically. However, in material quenched to produce an equiaxed grain structure, but with a random texture, the superplastic deformation was isotropic. Nuttall also reported (72) isotropic superplastic properties in quenched Zn-Al eutectoid alloy.

The anisotropy in Zn-0.4%Al sheet was recorded (68,69) in terms of the plastic strain ratio R , defined as the ratio of width to thickness strain. The R values were found to increase towards 1.0 with superplastic strain and strain rate sensitivity became more isotropic (71). The deformation at superplastic strain rates was apparently slip controlled with simultaneous recovery occurring, although the microstructure was not studied. Heubner et al (70), working on eutectoid Zn-Al, also noticed R

values increasing with superplastic strain and related this to decreasing texture intensity. Kaibyshev et al (87) reported texture intensities decreasing with superplastic strain in a range of Zn-Al alloys and also noticed a transition from single to multiple slip in the Zn rich phase with increasing strain rate in region II. Investigations of superplastic deformation in Al-Li alloys (61,79), Supral 220 (61) and the aluminium alloy 7475 (61) also revealed anisotropy decreasing with strain.

According to Kaibyshev et al (83) the presence of a strong rather than a random texture in Zn-22% Al with equiaxed grains resulted in a decrease in superplastic flow stress, an increase in elongation and in the strain rate corresponding to maximum m , although no anisotropy was observed in the flow stress at the superplastic temperature. It was suggested that the strong texture increased the rate of grain boundary sliding and facilitated the climb or glide of dislocations into the boundaries. Thus there are inconsistencies between this work and other results on Al alloys, which generally indicate some degree of texture controlled anisotropy.

Superplastic anisotropy has also been investigated in Pb-Sn eutectoid alloy (73-75). The evidence consistently suggests that, for this material, microstructural banding or grain elongation are more important than texture effects. In the material used by Melton et al (73) the initial texture was weak and was practically unchanged by superplastic strain. Break-up of the banded microstructure with strain was said to explain the increasingly isotropic behaviour and compression testing confirmed that the flow stress was highest parallel to the rolling direction. These results are consistent with those of Kashyap and Murty (74) who investigated Pb-Sn both with grains elongated in one direction by extrusion and swaging and grains elongated in two directions by rolling. Texture was not discussed in this work. However, Cutler and Edington (75) found that textures in both the Pb and Sn rich phases were slowly randomised by superplastic strain, consistent with grain boundary sliding rather than slip. Similar observations of texture randomisation were made on aluminium bronze by Dunlop et al (78).

Bricknell and Edington used detailed texture measurements in the form of crystallite orientation distribution functions (77) to predict (76) the effect of superplastic strain on R value of Al-6Cu-0.3Zr based on

single and multiple $\{111\}$ $\langle 110 \rangle$ slip. The predictions were in good agreement with experimental data for high strain rates, but at lower rates the predictions were increasingly inaccurate suggesting a gradual transition to grain boundary sliding. The increase of R value with strain was ascribed to the development of a $\langle 111 \rangle$ fibre texture at high strain rate and to the break-up of microstructural banding at lower rates.

The most thorough analysis made so far of superplastic anisotropy in titanium alloys is by McDarmaid et al (51,59-62,64,65) working on strongly banded Ti-6Al-4V with a pronounced texture. A gradual randomisation of alpha phase texture was found with anisotropy apparently arising from contiguous alpha grains aligned in the rolling direction. However, at relatively high strain rates or low temperatures a transition to slip controlled deformation occurred (61,64,65). At a temperature close to the beta transus when the alpha phase content was small the deformation was nearly isotropic and was not influenced by strain rate. Paton and Hamilton investigated (88) Ti-6Al-4V sheet containing similar microstructural banding and found anisotropic flow stress and strain rate sensitivity. Russian data (63) on the VT6 alloy Ti-6.5Al-5.1V are apparently contradictory in that superplastic anisotropy was reported in material with nominally equiaxed grains. This was said to result from the strong texture present. However, the sense of the anisotropy and its reversal at lower temperatures were the same as reported for the banded Ti-6Al-4V (60-62, 64,65) suggesting that microstructure rather than texture may have been responsible.

2.4 Superplastic forming of sheet

Despite the wide range of materials in which superplasticity has been investigated (1-4), the commercial exploitation of the phenomenon has concentrated on the sheet forming of aluminium (89-91) and titanium (89-95) alloys with microduplex stainless steel (96) and eutectoid Zn-22%Al (91, 92, 97) occasionally used.

Alloys based on Al-Ca and Al-Cu-Zr (Supral) have been specially developed for superplastic forming and a wide range of Supral components from ejector seat head boxes to machine covers, cladding for internal and external walls and roofs and even car body parts have been produced. For higher strength applications Si, Mg and Ge additions have been made to the Supral material and thermomechanical grain refining treatments have been developed (98-101) for the high strength 7000 series alloys, which improve the superplastic properties. All the aluminium alloys suffer

from cavitation during superplastic strain (79,102,103) which degrades the service properties, although forming with a superimposed hydrostatic pressure can alleviate the problem (104).

The titanium alloy Ti-6Al-4V is highly superplastic and unlike aluminium alloys does not cavitate significantly during superplastic forming. It also readily diffusion bonds at temperatures in the superplastic range so that forming and bonding can often be carried out in the same operation and complicated, internally stiffened components can be produced from a lay-up of several sheets. Superplastic forming or the combined superplastic forming/diffusion bonding (SPF/DB) fabrication processes have already been successfully applied to a number of demonstration and production aerospace components with cost savings up to 50% and weight savings up to 30% (95) compared with conventionally produced titanium parts. These savings are achieved because fewer fasteners are required, much machining is eliminated and more structurally efficient designs are possible. Titanium SPF/DB components can also economically replace aluminium or steel parts in some cases.

For superplastic forming of the aluminium Supral alloys aluminium tools can be used, whereas steel tools are required for the 7000 series alloys, which ^{can} have higher superplastic flow stresses. Titanium alloy forming requires much more expensive cast or machined steel or nickel based dies capable of withstanding temperatures up to about 950°C.

A variety of superplastic forming techniques for aluminium alloys have been investigated (89,91) in order to minimise the thickness variation of the forming and to reduce material wastage. Simple gas pressure blowing into a female mould results in a decrease in thickness from the edge of the forming to the corners, which contact the die wall last. Drape forming over a male die sitting in a cavity improves the thickness distribution, but leads to some material wastage. Reverse billowing is another technique used to reduce thickness variation. In this process the sheet is initially free-blown away from the mould by gas pressure to a height greater than the mould depth. The pressure is then reversed and the sheet blown into the mould so that the peak of the dome is the first part of the forming to touch the mould surface. For relatively deep formings it is usual to use male tools moving into the sheet. In plug assisted forming a prestretch is carried out using a moving tool and then pressure is used to complete the forming into a female mould.

Titanium sheet formings are almost always carried out using argon gas pressure and a female mould. When diffusion bonding is also required then a lay-up or pack of several sheets is assembled beforehand with a stop-off compound, usually yttria, applied where bonding is not required. The sheets may also be profiled by chemical milling before forming or be stamped out, depending on the reinforcement or the geometry of the internal stiffening required. Diffusion bonding may be by platen pressure before or during the forming operation or by gas pressure. Joining by TIG welding before inflation of the pack is also occasionally carried out.

Resistance to neck growth during plastic deformation at room temperature is largely determined by the rate of work hardening with the strain rate sensitivity m becoming important in the post uniform stage (105). The normal anisotropy, measured by the R value can also significantly affect sheet formability. Sheet drawability (106), for example, is enhanced by high R values as the potential failure site at the bottom of the cup wall, where plane strain deformation occurs, is effectively strengthened, whereas the flange area (pure shear), near the cup rim, is effectively weakened. The influence of R on limiting strain (the strain at which localised necking begins) also depends on the stress state. In uniaxial tension increasing the R value delays the development of the plane strain deformation that is required for localised necking (105) and thus increases the limiting strain. This is true for all cases where the minor strain ϵ_2 is negative (107,108). However, under plane strain conditions (105,109) the limiting strain is independent of R and for positive minor strains ($\epsilon_2 > 0$) the limiting strain decreases with R value. Strongly basal textured titanium alloys can have R values up to 12 (105,107).

The deformation limits for cold forming are determined by minimum bend radii and by strain and strain ratio, which are considered together in the forming limit diagram. Failure by localised necking or fracture usually occurs if the limits are exceeded. Each of the processes of drawing, bending, or stretching over a punch have different constraints and friction considerations associated with them and different mechanical properties may be critical in each case.

Superplastic forming differs from cold sheet forming in a number of ways. The m value is usually assumed to be the most important factor

in promoting large elongations and the material may be processed beforehand to give a fine, stable grain size for this reason. The R value has the same influence on superplastic formings as on cold sheet formings. However, superplastic deformation is not usually influenced by texture and the normal anisotropy under superplastic conditions is unlikely to be as pronounced as that observed during room temperature deformation.

Superplastic sheet forming is normally carried out by gas pressure blowing such that the stress state is biaxial tension. This mostly precludes localised necking (106) and even where localised necking could occur such as in corners or in long trough-shaped formings, the material is usually sufficiently strain rate sensitive to avoid localised necking and diffuse necking only will occur. Thus, failure of non-cavitating material during superplastic forming is rarely a problem.

3 EXPERIMENTAL PROCEDURE

3.1 Material

The Ti-6Al-4V sheet was supplied by Reactive Metals Inc. in four thicknesses: 0.9, 1.8, 2.0 (batch A) and 3.3mm. A further two batches (B and C) of 2.0mm Ti-6Al-4V sheet and 2.0mm IMI 550 sheet (Ti-4Al-4Mo-2Sn-0.5Si) were obtained from IMI Titanium. All of the Ti-6Al-4V and IMI 550 sheet had been cross rolled and annealed at 700°C for 2 hr. The Ti-6Al-4V bar 160 x 55mm was also supplied by IMI and was annealed for 4hr at 700°C before machining. The 2.0mm Ti-8Al-1Mo-1V (duplex annealed: 790°C 8hr, furnace cooled, 790°C 1/4 hr, air cooled) and the 2.0mm Ti-15V-3Cr-3Al-3Sn (annealed 790°C, 2hr) were supplied by Timet.

3.2 Superplastic deformation

Sheet test pieces were machined with the final rolling direction parallel to the tension axis (L) and perpendicular to the tension axis (T). The test piece heads were reinforced by plates of 2mm thick Ti-6Al-4V sheet spot welded onto each side in order to minimise head distortion during superplastic testing. A diagram of the test piece is given in Fig 1. For the determination of superplastic properties of sheet material (section 3.1) a gauge length (l_0) of 25mm and a gauge width (w_0) of 16mm was used, except for the Ti-8Al-1Mo-1V tests ($w_0 = 12\text{mm}$). For the series of tests investigating the effect of test piece shape on R value L orientation test pieces of 3.3mm thick Ti-6Al-4V sheet were used either (a) with a gauge length of 10mm and gauge widths of 4, 8, 16 or 19.5mm or (b) with a gauge width of 16mm and gauge lengths of 2, 5, 10 or 25mm as shown in Fig 1.

Sheet test pieces were machined from the 55mm thick Ti-6Al-4V bar in three orientations; TL, ST and LT where the first letter indicates the tensile axis of the test piece and the second letter refers to the orthogonal direction in the plane of the test piece and L, T and S are the longitudinal, transverse and short transverse directions of the original bar. A cutting diagram is shown in Fig 2. The bar thickness was insufficient for the extraction of the ST orientation test pieces. Therefore, 4mm thick blanks for the gauge length were machined from the bar in the ST orientation and extended by electron beam welding pieces of 4mm thick Ti-6Al-4V sheet to each end.

Final machining of the test pieces to 2mm thickness was then carried out with a gauge length of 25mm and a gauge width of 12mm.

Uniaxial superplastic straining was carried out on a screw driven tensile testing machine with the test piece enclosed in a quartz tube sealed at each end by water cooled brass caps. A gas tight seal was maintained by heat resistant O-ring seals around the pull rods and an inert atmosphere was provided by a slow flow of argon, which was passed over titanium swarf at 900°C in a separate furnace in order to getter the oxygen before entering the testing chamber. The pull rods, test piece grips and loading pins were made from nimonic 75 and were assembled before each test with graphite flakes on the threads to prevent sticking. A vertical three zone furnace provided a hot zone to $\pm 2^\circ\text{C}$ over a length of 100mm. A series of circular stainless steel heat shields push-fitted onto the pull rods inside the quartz tube both above and below the test piece helped to maintain the hot zone and the furnace was lagged top and bottom to reduce heat loss. Temperature was monitored by three thermocouples which passed through the top cap and upper heat shields and were wired onto the test piece. Cross-head displacement was measured by an integral digital extensometer and load, extension and temperature were recorded on a moving chart. The testing apparatus is shown in Fig 3.

The test temperatures used for the rolled sheet test pieces were: Ti-6Al-4V: 925°C, IMI 550: 900°C, Ti-8Al-1Mo-1V: 910, 940, 970 and 1010°C, Ti-15V-3Cr-3Al-3Sn: 810, 860 and 910°C. A strain rate of $3 \times 10^{-4}\text{s}^{-1}$ was used for the majority of the tests with some tests on the 3.3mm Ti-6Al-4V sheet carried out at 10^{-3}s^{-1} and $2.5 \times 10^{-3}\text{s}^{-1}$. For the Ti-6Al-4V test pieces machined from bar four test temperatures of 800, 875, 925 and 975°C and two strain rates of $3 \times 10^{-4}\text{s}^{-1}$ and $1.5 \times 10^{-3}\text{s}^{-1}$ were employed.

The strain rate was maintained nominally constant by increasing the cross-head speed as appropriate after each 50% strain increment. Flow stress and strain rate sensitivity were determined for the 3.3mm Ti-6Al-4V sheet for up to 400% superplastic strain by interrupting the test after 200% strain and remachining the gauge length parallel. Flow stress and m value data for the other Ti-6Al-4V sheet thicknesses and for the other alloys were determined for superplastic strains nominally up to 200% using single, rather than interrupted tests.

The strain rate sensitivity index m was measured over the strain rate range $2 \times 10^{-5} \text{s}^{-1}$ - $4 \times 10^{-3} \text{s}^{-1}$ for each alloy by repeatedly doubling the cross-head speed and recording the maximum load after each speed increment. The m value was calculated from the approximation (1):

$$m = \frac{d \ln \sigma}{d \ln \dot{\epsilon}} \approx \frac{\ln(P_2/P_1)}{\ln(V_2/V_1)}$$

where P_1 and P_2 , and V_1 and V_2 are the load and cross-head speed before and after the strain rate increase. This "step-strain rate" procedure was carried out after a 4% prestrain at $2 \times 10^{-4} \text{s}^{-1}$. In order to minimise the effects of strain it was desirable to cover the entire strain rate range in the smallest strain increment possible, whilst allowing steady state flow to be achieved after each increase in cross-head speed. It was found that, for the lowest stresses, a constant or maximum load was not reached and, in these cases, the strain rate steps were carried out after 1% strain increments. In nearly every case the entire strain rate range was covered within a total strain increment of 25%.

Strain rate sensitivity was also determined as a function of strain by temporarily increasing the cross-head speed by 25% after each 50% strain increment.

Calculations of flow stress were made from estimates of cross-sectional area at various strains during the test based on micrometer measurements of width and thickness strain made on the gauge length centre after the test. These measurements were also used to calculate the R values: $R = \epsilon_w / \epsilon_t = \ln(w/w_0) / \ln(t/t_0)$

The cooling rate from the test temperature was $25^\circ \text{C min}^{-1}$, except for some of the IMI 550 tests, for which faster cooling rate of $150^\circ \text{ min}^{-1}$ was used. The appropriate cooling rate was maintained down to 700°C and air was not admitted to the test chamber until the temperature of the test piece was 300°C or lower.

3.3 Room temperature tensile testing

The test pieces used for the determination of room temperature tensile properties of the Ti-6Al-4V sheet were standard "D-size" test pieces (gauge length 15.9mm) machined from the (larger) gauge lengths

of the superplastically deformed specimens. It was subsequently found to be more convenient simply to remachine the edges of the superplastic tensile test pieces and retest at room temperature and this approach was used for the other alloys. The tensile test pieces used for the as-received conditions and for material annealed at the superplastic temperatures were standard "B-size" (gauge length 32mm). The annealing was carried out in a vacuum furnace at a pressure of $< 10^{-4}$ torr with the appropriate cooling rate achieved by slowly sliding the furnace off the vacuum tube. The room temperature testing was carried out at a strain rate of $5 \times 10^{-5} \text{s}^{-1}$.

3.4 Microstructural examination

The etchants used for the microstructural examinations were the stain etch (40% methanol, 40% glycerol, 15% bezalkonium chloride, 5% hydrofluoric acid) for the quenched microstructures and Kroll's reagent (1% hydrofluoric acid, 12% nitric acid, 87% water) in all other cases. Phase proportions in the quenched Ti-6Al-4V sheets were determined by point counting using a 10 x 8 grid over 10 different areas ie 800 points for each microstructure. The mean linear intercept (mli) of the alpha phase was measured ie alpha/alpha grain boundaries were ignored and then the contiguous alpha grain size g was calculated in the two principal directions on the LS and TS sections from:

$$g = 2 \times P \times \text{mli}$$

where P is the alpha phase proportion. The alpha phase aspect ratios were then determined from the grain size measurements. Alpha phase aspect ratios were also determined for the 3.3mm Ti-6Al-4V sheet after superplastic strain and after reheating to the forming temperature and quenching into water.

3.5 Analysis of necking Ti-6Al-4V during superplastic deformation

An attempt has been made to analyse neck development during superplastic deformation using uniaxial flow stress data measured on L orientation Ti-6Al-4V sheet test pieces of initial thickness 3.3mm at a testing temperature of 925°C. A computer programme was written to predict the effect of a small strain increment in the uniform region of an imaginary test piece on the strain, strain rate, strain rate sensitivity and strain hardening in a pre-existing neck or inhomogeneity.

The initial area of the inhomogeneity was varied from 50% to 99% of that of the uniform region. For the purposes of the calculation the neck and uniform regions were treated as two separate test pieces pulled in series such that the strain rate in the uniform region was constant at $3 \times 10^{-4} \text{s}^{-1}$ (approximately that of maximum m at low strain). The neck strain rate was assumed to be constant during each strain increment, but both the uniform and the necked regions were allowed to strain harden during each increment according to the true stress-true strain curves for the strain rate in each region. After each increment a new value of neck strain rate was calculated.

Interpolations of the experimental data were made by a curve fitting subroutine and the values of strain rate sensitivity $d \ln \sigma / d \ln \dot{\epsilon}$ and strain hardening exponent ($n = d \ln \sigma / d \ln \epsilon$) obtained by differentiation.

The experimental data used is shown in Fig 4. Step-strain rate tests were used to determine the $\ln \sigma / \ln \dot{\epsilon}$ relationships at true strains of approximately 0.25 and 0.8, true stress-true strain curves were determined for three different strain rates of $3 \times 10^{-4} \text{s}^{-1}$, 10^{-3}s^{-1} and $2.5 \times 10^{-3} \text{s}^{-1}$ and the gradient ($d \ln \sigma / d \ln \dot{\epsilon}$) was determined as a function of strain at a strain rate of $3 \times 10^{-4} \text{s}^{-1}$. Flow stresses at zero strain were estimated by extrapolation of the true stress-true strain curves back to zero strain. The experimental data, shown by circles in Fig 4, were extrapolated up to a strain rate of $5 \times 10^{-2} \text{s}^{-1}$ and down to $4.5 \times 10^{-5} \text{s}^{-1}$. However, data for strain rates $< 3 \times 10^{-4} \text{s}^{-1}$ were not required in the calculation.

4 RESULTS

4.1 Superplastic deformation and subsequent tensile properties

4.1.1 Ti-6Al-4V

Plots of $\log \sigma$ against $\log \dot{\epsilon}$ are given in Fig 5 for L and T orientations for sheets of initial thickness 0.9–3.3mm. The $\ln \sigma / \ln \dot{\epsilon}$ relationships for the three batches of 2mm sheet are shown in Fig 6. For each of the sheets there was apparently little effect of test piece orientation on flow stress. The flow stresses were similar for each of the sheets tested except that the 2mm thick batch A material showed slightly lower flow stress at low strain rates, although this is exaggerated by the logarithmic axes. The effect of strain rate on strain rate sensitivity m for each of the Ti-6Al-4V sheets is shown in Figs 7 and 8. In each case a peak in m value occurred at a strain rate of about $2 \times 10^{-4} \text{ s}^{-1}$. At a given strain rate the variation in m value between the sheets was typically ± 0.05 and repeat tests on several of the sheets resulted in variation in m values of about ± 0.02 at a given strain rate.

The effect of superplastic strain on flow stress at 925°C and $3 \times 10^{-4} \text{ s}^{-1}$ is recorded in Fig 9. A solid line has been drawn through the data points for the sheet of initial thickness 3.3mm. The flow stress increased with strain up to a true strain of about 1.4 (300%). There was little effect of strain on flow stress for higher strains.

Strain rate sensitivity m decreased with strain at 925°C as shown in Fig 10, which gives data for the 3.3mm Ti-6Al-4V sheet.

The effects of superplastic strain on room temperature 0.2% proof stress (0.2PS) and tensile strength (TS) are shown in Figs 11 and 12. The full tensile results for the 3.3mm sheet are given in Table 1 and the data for the 2.0mm batch A material can be found in ref 110. No room temperature tensile tests were performed on the 2mm batch C material. The effect of annealing at the superplastic temperature, corresponding to zero strain in Figs 11 and 12, was to reduce the 0.2PS by 8–12% and the TS by 1–10% compared with the as-received (mill annealed) material. Superplastic true strain of 0.9 reduced the 0.2PS and TS typically by a further 2%, although for the T orientation there was apparently no effect of superplastic strain on strength for several of the sheets when compared with the heat cycled condition. Uniform and total elongations and moduli were generally unaffected by superplastic strain.

The starting textures ranged from a basal "edge" texture (basal poles parallel to the T direction) in the thickest sheet (Fig 13) to a "sheet" texture (basal poles at about 20° to the sheet normal inclined towards the final rolling direction) in the thinnest sheet. The 2.0mm thick sheets showed intermediate textures. Despite the cross rolling process only the 2.0mm batch B material showed four-fold symmetry in the pole figure. The starting textures for each of the sheets, except for the 1.8mm and 2.0mm batch C sheets, are given elsewhere (111). Only the texture of the 3.3mm sheet is considered further here.

The basal pole figure for the sheet surface (Fig 13a) showed concentrations of poles parallel to the transverse direction and normal to the sheet plane. The intensities of peaks close to the edge of the pole figure cannot be accurately determined or the peaks may remain undetected. Hence, the intensities of poles close to the transverse direction are likely to be higher than indicated. The prism planes at the sheet surface (Fig 13b) lay in the plane of the sheet or inclined towards the L direction. The $(110)_\beta$ texture pattern (Fig 13c) was similar to that of the $(0002)_\alpha$. The pole figures for the sheet centre were generally more distinct and with more intense peaks than those of the sheet surface. The $(0002)_\alpha$ pole figure (Fig 13d) showed an edge texture component and an annular concentration around the sheet normal. The $(10\bar{1}0)$ pole figure for the sheet centre (Fig 13e) is different from that of the sheet surface (Fig 13c) and indicates a rotation of 30° about the c axis such that for the sheet centre $\langle 11\bar{2}0 \rangle$ was parallel to the sheet normal, whereas for the sheet surface $\langle 11\bar{2}0 \rangle$ was parallel to the L direction. The $\{100\} \langle 110 \rangle_\beta$ texture (Fig 13f) is a typical beta rolling texture (112-114) with $\{110\}$ poles parallel to the T direction and the L direction (not detected) and at 45° to the sheet normal.

Annealing at 925° (Fig 14) caused a sharpening of the texture (112,115) with significantly higher peak intensities in the sheet centre. The $(0002)_\alpha$ and $(110)_\beta$ pole figures for the sheet surface showed the development of four new intensity peaks (Fig 14a and c).

Superplastic strain caused a substantial reduction in texture intensities (Fig 15). Neither the alpha nor the beta phase textures were completely removed, although for the sheet surface only the $(0002)_\alpha$ and $(110)_\beta$ peaks close to the sheet normal were detected (Figs 15a and c).

The microstructure of the LS section of the 2mm batch A material, shown in Fig 16a consisted of alpha grains elongated in the final rolling direction with particles of beta phase between the alpha grains. The microstructures of the other Ti-6Al-4V sheets were similar (111). Heating to the forming temperature (925°C 1hr , furnace cooled at 25°C min⁻¹) resulted in more equiaxed and larger grains (Fig 16b). Superplastic strain to 150% followed by cooling at 25°C min⁻¹ (Fig 16c) caused a further increase in grain size compared with the heat cycled material and largely removed directionality in the microstructure.

The microstructures in Fig 16 are not representative of the structure at the forming temperature because of the transformation which occurred during cooling. Fig 17, which shows the microstructure of the mill annealed 3.3mm sheet after quenching into water from 925°C provides a better indication of the high temperature structure. The LS section microstructure (Fig 17a) was much more strongly directional than that of the TS section (Fig 17b). However, this difference was much less marked in the other batches of Ti-6Al-4V sheet (111). Superplastic strain in the L direction (Fig 18) increased the grain size and reduced the directionality in the microstructure. Each of the microstructures in Fig 18 have been reheated to the forming temperature and water quenched.

The alpha aspect ratio ie ignoring alpha/alpha boundaries was measured for the quenched LS microstructure and is shown as a function of superplastic strain in Fig 19. The alpha phase appeared to be completely equiaxed after a strain of about 1.5 in the L direction.

4.1.2 IMI 550 (Ti-4Al-4Mo-2Sn-0.5Si)

The flow stress of IMI 550 at 900°C is given as a function of strain rate in Fig 20. There was almost no effect of test piece orientation on flow stress and the m values for the L and T orientations were also very similar (Fig 21) with the peak value corresponding to a strain rate of about 10^{-4} s^{-1} . The flow stress at 900°C increased gradually with superplastic strain (Fig 22) and the strain rate sensitivity apparently showed a slight increase up to a strain of about 0.8 (Fig 10).

The effect of superplastic strain at 900°C at a strain rate of $3 \times 10^{-4} \text{ s}^{-1}$ on the 0.2PS and TS of IMI 550 is shown in Fig 23 and the full tensile results are given in Tables 2-4. Annealing at the forming

temperature caused a slight reduction in 0.2PS and TS, compared with the as-received material. Superplastic strain up to 190% apparently had little further effect on room temperature strength, although there was some scatter in the L orientation results (Fig 23). Total elongations were significantly lower after superplastic strain than in the as-received condition.

The effect of post-forming heat treatment on the room temperature tensile properties of the L orientation was investigated. Ageing at 500°C for 24 hrs after forming increased the 0.2PS and TS by about 14%. The standard post-forming cooling rate was 25°C min⁻¹. However, a faster cooling rate of 150°C min⁻¹ produced, in the as-formed condition, 0.2PS and TS increases of about 11% compared with the slow cooled samples. Ageing of the fast cooled material further increased the 0.2PS by 10% and the TS by 8%.

Tensile tests were also carried out on material annealed at the forming temperature for 1/2 hr to simulate the heat cycle associated with superplastic strain. Cooling rates of 25 and 150°C min⁻¹ were used and both aged and unaged specimens were tested. The proof, tensile strengths and total elongations for these conditions (Table 2) were found to be slightly higher than the corresponding values for the superplastically formed and heat treated conditions (Table 4).

Table 2 also includes tensile properties for material annealed at 900°C for 3½ hrs and cooled at either 25 or 150°C min⁻¹. This relatively long anneal reduced proof and tensile strengths, although the material remained sensitive to cooling rate.

The as-received (annealed) LS section microstructure of IMI 550 (Fig 24a) contained highly deformed alpha grains elongated in the longitudinal direction. The TS section was similar with alpha grains elongated in the transverse direction. The alpha phase was much more equiaxed after heat cycling (900°C 1hr, cooled at 25°C min⁻¹, Fig 24b) and superplastic strain up to 150% further reduced the directionality in the microstructure and slightly increased the grain size (Fig 24c).

4.1.3 Ti-8Al-1Mo-1V

The flow stress of Ti-8Al-1Mo-1V increased with strain rate and decreasing temperature as shown in Fig 25. The strain rate corresponding to peak $\dot{\epsilon}_m$ value increased with temperature (Fig 26) and a maximum $\dot{\epsilon}_m$

value of 0.82 was found for both orientations for a test temperature of 1010°C for both orientations. The m value was highly strain rate sensitive at this temperature (Fig 26). Superplastic strain caused some increase in flow stress at the higher test temperatures (Fig 27), but at 910°C strain softening apparently occurred. Strain rate sensitivity at 1010°C decreased rapidly with strain (Fig 10).

The room temperature transverse tensile properties of Ti-8Al-1Mo-1V before and after superplastic strain are given in Table 5. The effect of annealing at each of the forming temperatures of 940, 970 and 1010°C was to reduce the 0.2PS and TS by about 5% compared with the as-received condition. Superplastic strain of 200% further reduced the proof and tensile strengths by another 5%. There was apparently no effect of forming temperature in the range 940-1010°C. The room temperature properties following forming at 910°C could not be evaluated because each of the test pieces pulled at this temperature failed during superplastic testing after about 200% strain.

The alpha and beta phase textures (not included) of the Ti-8Al-1Mo-1V sheet after 120% strain at 1010°C were similar to those of the as-received sheet. The pole figures showed $(0002)_{\alpha}$ and $(110)_{\beta}$ peaks close to the sheet normal tilted slightly towards the rolling direction with indications of another set of peaks close to the transverse direction. Pole figures for material annealed at the forming temperature were not determined. However, the texture results on this alloy were consistent with those on Ti-6Al-4V (Figs 13-15), which indicated a sharpening of texture associated with recrystallisation during annealing and then a gradual reduction of texture intensity with superplastic strain.

The microstructures of the LS sections of the Ti-8Al-1Mo-1V in the as-received (duplex annealed) condition, after annealing 1010°C for 1hr and cooling at $25^{\circ}\text{C min}^{-1}$ and after superplastic strain of 150% in the L direction at 1010°C are shown in Fig 28. Annealing at the forming temperature caused a marked increase in grain size and superplastic strain produced a further slight increase. The superplastically formed Ti-8Al-1Mo-1V consisted of about 50% transformed beta in the form of Widmanstatten alpha plates (Fig 28c), whereas the heat cycled material contained no such secondary alpha plates.

4.1.4 Ti-15V-3Cr-3Al-3Sn

Flow stress and m values for Ti-15V-3Cr-3Al-3Sn are plotted against strain rate in Figs 29 and 30. The m values decreased slowly with strain rate and there was little effect of temperature in the range 810-910°C. The flow stress remained roughly constant up to superplastic true strains of 1.0, but some strain softening occurred thereafter (Fig 31). Strain rate sensitivity at 910°C increased from 0.35 to about 0.4 after a strain of 1.4 (Fig 10).

The effect of superplastic strain on room temperature tensile properties of Ti-15V-3Cr-3Al-3Sn are shown in Table 6. Both strength and ductility were reduced by 300% superplastic strain, although the post-forming properties appeared to be independent of forming temperature in the range 810-910°C.

The $(110)_\beta$ pole figures for the Ti-15V-3Cr-3Al-3Sn before and after superplastic strain are given in Fig 32. The X-ray analysis suggested that the presence of a small amount of alpha phase may have slightly increased the apparent texture intensities in Fig 32a. However, it is clear that the texture in this alloy is not randomised by superplastic strain, unlike that of the Ti-6Al-4V (Figs 13-15).

Superplastic strain of Ti-15V-3Cr-3Al-3Sn at each of the temperatures investigated produced surface rumpling as shown in Fig 33. This extended into the test piece heads in the regions close to the ends of the gauge length.

The microstructures of the material in the as-received (annealed) condition and after various amounts of superplastic strain at 810°C are shown in Fig 34. Annealing at the forming temperature caused a large increase in grain size (Fig 34a), but superplastic strain resulted in grain refinement (Figs 34b-g). The effect of superplastic strain at 810 and 860°C on grain length in the L direction is shown in Fig 35. Both the heat cycled and the superplastically formed material contained alpha precipitates in two forms: relatively large intragranular precipitates and much finer particles decorating the grain and subgrain boundaries (Fig 34h). This precipitation probably occurred during cooling from the forming temperature.

4.2 Superplastic deformation of sheet test pieces machined from Ti-6Al-4V bar

4.2.1 Test piece shape after superplastic strain

The appearance of the test pieces after superplastic deformation under each of the conditions of temperature and strain rate is shown in Fig 36.

4.2.1.1 TL orientation

The TL orientation test pieces after testing were characterised by a very irregular strain distribution along the gauge length with bands of relatively undeformed material running across the test piece. This behaviour was most marked for the 875°C and 925°C test temperatures and the banding is shown at A in Fig 36. The banding was more apparent in the low strain rate test pieces than in the corresponding high strain rate test pieces for test temperatures 800-925°C. The test pieces pulled at 975° failed at a relatively low strain with no evidence of banding across the gauge width.

4.2.1.2 ST orientation

For the ST orientation the banding direction was through the thickness of the test pieces and small blocks of relatively undeformed material were left protruding out of the surface eg Fig 37a. This effect was more clearly visible near the ends of the test piece gauge lengths ie in the low strain regions than in the centre of the gauge lengths or in the necked regions, where the test piece surfaces were relatively smooth. The test pieces pulled at 975°C exhibited relatively low total elongations and, for both strain rates, showed shallow trenches in the test piece surface running roughly perpendicular to the tensile axis (Fig 38). Each of these features was an incipient neck as shown schematically in Fig 38b. This behaviour was peculiar to the ST test piece orientation. Final necking at 975°C occurred along a line at an angle to the tensile axis. The 800°C ST test pieces also showed this "slant" type local necking (Fig 36).

4.2.1.3 LT orientation

By contrast, the LT orientation test pieces pulled at temperatures up to 925°C showed more uniform deformation with relatively smooth surfaces and parallel gauge lengths, except that shallow troughs and ridges, running parallel to the test piece axes were visible (Fig 37b). The height and depth of these features was much less than that of the surface irregularities occurring in the test pieces of the other

orientations. These surface grooves were also observed on round test pieces pulled in the L direction.

4.2.2 Microstructure

The microstructure of the as-received material contained bands of heavily deformed material with strongly directional microstructure eg at A in Fig 39 and more homogeneous, non-directional regions such as at B in Fig 39. The microstructures of material quenched into water from each of the forming temperatures are shown in Fig 40. The alpha phase volume fractions, estimated from the data in ref 64, were, at 800, 875, 925 and 975°C respectively, 81%, 68%, 50% and 10%. As discussed in refs 59, 64, 65 and 111, the banded regions, such as at A in Fig 39, contain equiaxed, contiguous alpha grains aligned in the rolling direction. These aligned alpha grains can be seen on the LS and LT sections in Fig 40c. There is also evidence from Figs 39 and 40 of alpha phase alignment, or possibly grain elongation in the T direction, although to a lesser extent than in the L direction. The distribution of aligned microstructure is shown schematically in Fig 41, reproduced from ref 51. The nature of the surface protrusions in Fig 37a lend some support to the shape suggested in Fig 41 for the bands of aligned alpha grain. In Fig 40d the etchant has attacked the martensite matrix and martensite plates are visible between the alpha grains. Each of the microstructures in Fig 40 shows the alpha phase alignment, although this is less marked for the higher forming temperatures as a result of the lower alpha phase content. The material quenched from 975°C (Fig 40d) showed some variation in primary alpha phase proportion associated with the regions of aligned and non-aligned alpha grains, although it was not clearly determined that the alpha rich regions corresponded either to the aligned or to the non-aligned regions. The TS section of Fig 40d shows an alpha rich region at A and an alpha lean region at B.

A section of 925°C TL test piece, through a region such as at A in Fig 36b, is shown in Fig 42. Two distinct areas of different microstructure can be seen; aligned alpha grains such as at A and equiaxed microstructure at B. Other banded regions on the TL test pieces were all found to contain the aligned microstructure. Similarly, the protrusions on the surfaces of the ST orientation test pieces (Fig 37a) were found to correspond to areas of aligned microstructure, such as at A in Fig 43, and depressions in the surface were associated with more equiaxed microstructure (at B in Fig 43).

The effect of superplastic strain was to increase the average grain size, particularly during the first 100% strain increment, and to break up the banded microstructure. The LS section is shown in Fig 44 after various amounts of strain in the S direction at 875°C and at a strain rate of $3 \times 10^{-4} \text{ s}^{-1}$. After testing, the material was reheated to 875°C and quenched into water. The microsections for Figs 44a-d were taken from parts of the test piece where the irregular surface suggested the microstructure was banded. The micrographs in Figs 44e and f corresponded to regions near the centre of the gauge length where the test piece surface was relatively smooth. The values for axial strain given in Fig 44 were calculated from measurements of local width and thickness strain and by assuming constant volume.

A section through an ST test piece pulled at 975°C revealed that the microstructure in the necked regions contained practically no primary alpha phase and showed relatively large alpha plate colonies indicative of a large prior beta grain size (Fig 45). The neck which propagated to failure contained only a few isolated primary alpha grains indicating that the composition at the forming temperature was almost completely beta (Fig 46a). By contrast, the microstructure of a 975°C TL test piece showed a uniform distribution of primary alpha phase eg Fig 46b which shows the microstructure close to the point of failure (compare with Fig 46a).

The ST test pieces B2 and B3 and the TL test piece A4, which had almost necked to failure (Figs 36c,d,f), showed cavitation in the necked region eg Fig 47. The cavitation seemed to originate either in the beta phase or at the alpha/beta phase boundaries (Fig 47) and the cavities appeared to elongate or coalesce parallel, rather than perpendicular to the tensile axis.

A scanning electron micrograph of the edge of a test piece after superplastic strain is shown in Fig 48. The surfaces, which had a machined finish prior to testing, have been roughened by grain boundary sliding and the emergence of new surface grains.

4.2.3 Flow stress

The superplastic flow stresses for each orientation and test temperature are given in Fig 49. These values were calculated from the maximum load, which occurred in each test at 15% strain or less, and the initial cross-section area. At such low strains the test piece gauge

lengths would still be approximately of uniform cross-section and parallel sided. Flow stresses were higher for the high strain rate and decreased with increasing temperature. For each strain rate the flow stresses were highest for the LT orientation and lowest for the ST orientation.

4.2.4 Strain rate sensitivity

The m values, measured at the point of maximum load by temporarily increasing the crosshead speed by 25%, are plotted against temperature in Fig 50. For each temperature the m values were higher for the low strain rate tests and for each orientation and strain rate reached a maximum at 925°C. With the exceptions of the high strain rate tests at 800°C and the low strain rate tests at 975°C, the strain rate sensitivity values were in the order $m_{ST} > m_{TL} > m_{LT}$. This was the reverse of the order of flow stresses (Fig 49).

4.2.5 R values

R values were measured at various points along the gauge length of each of the test pieces and have been plotted against the local axial strain ϵ_1 in Fig 51. The two strain rates given in Fig 51 were only nominal values as the irregular strain distributions in the TL and ST test pieces implied a wide variation in local strain rate.

The most significant feature of the R values for the ST test pieces was that for test temperatures of 800-925°C all the R values measured were greater than 1.0, although a large amount of scatter was recorded. For the 975°C test pieces R values ranged from 0.77 inside the necked regions to about 1.0 in the adjacent unnecked area (see Fig 38).

For the LT orientation there was much less variability in R than for the other orientations. All the R values measured for this orientation were less than 0.9.

The R values recorded for the TL orientation were measured both on the banded regions of the test pieces, such as at A in Fig 36, and in the regions between the bands and have been plotted separately as closed and open symbols respectively. No trend of R value with strain was apparent, although for the low strain rate tests (Fig 51a) the R values corresponding to the banded regions tended to be higher than those measured in the non-banded areas, particularly for the 875°C test temperature. This is shown in Fig 52 where the true width and thickness

strains ϵ_w and ϵ_t have been plotted against ϵ_l for the 875°C low strain rate test and the measurement locations on the test pieces have been identified.

4.2.6 Texture

The pole figures for the material before and after superplastic strain are given elsewhere (116). They revealed that, unlike the case of the rolled sheet (section 4.1.1), annealing at the forming temperature did not appreciably sharpen the texture and superplastic strain at 925°C caused a gradual reduction in texture intensity. However, even after 374% strain neither the alpha nor the beta phase texture intensities were completely randomised.

4.3 R values of Ti-6Al-4V sheet after superplastic strain

4.3.1 Effect of sheet thickness

The effect of uniaxial superplastic strain on R for the L and T orientations of each of the sheets is shown in Fig 53. All the R values determined were less than 1.0 ie strain through the thickness exceeded strain across the width. The R value increased with superplastic strain for both orientations of each sheet, with the R values generally higher for the L orientation. It also appeared that the R value decreased with initial sheet thickness.

The microstructure of the 3.3mm Ti-6Al-4V as-received sheet (Fig 17 and ref 111) showed some elongation and alignment of the alpha grains along two orthogonal directions in the plane of the sheet (the rolling directions) though this was much less severe than that observed in highly textured Ti-6Al-4V bar as discussed in section 5.2 where the anisotropy of superplastic flow was clearly related to the effects of microstructure. In Fig 54 alpha phase aspect ratios of the material quenched from the forming temperature are plotted against the R value corresponding to a true axial strain of 0.4. For each point the aspect ratio plotted was measured on the cross-section of the corresponding specimen ie for the L orientation specimens the alpha phase aspect ratio on the TS section is recorded. This figure shows that the exceptionally low R value measured for the T orientation of the 3.3mm sheet was accompanied by a relatively high alpha phase aspect ratio and when the R value was high (0.9mm thick sheet, L orientation) a correspondingly low alpha phase aspect ratio was observed. The other orientations of the 0.9 and 3.3mm sheet and the other sheet thicknesses showed intermediate values of R and aspect ratio.

4.3.2 Effect of test piece geometry

The effect of test piece geometry was investigated using test pieces of 3.3mm Ti-6Al-4V sheet. The true width and thickness strains, ϵ_w , ϵ_t , are plotted against axial strain ϵ_1 in Fig 55a for test pieces with the same initial gauge length, but different gauge widths. The axial strain at the centre of each specimen was calculated from the width and thickness measurements assuming constant volume : $\epsilon_1 = - (\epsilon_w + \epsilon_t)$. The strain was more anisotropic for larger gauge widths, but points for the two narrowest specimens (4mm and 8mm gauge widths) lay on the same curve. This indicated that there was no effect of gauge width below 8mm. Similarly, varying the gauge length (Fig 55b) resulted in more anisotropic behaviour at shorter gauge lengths.

Test pieces of various initial widths are shown after straining in Fig 56. In each case constraint by the test piece head near the ends of the gauge length has reduced the width strain there, resulting in tapering along the gauge length. However, the taper was less severe for the narrower specimens and for the smallest gauge width (4mm) the gauge length was approximately parallel and free from the influence of the head over most of its length (Fig 56d).

Plotted in Fig 57 are width and thickness strain measurements taken from different points along the gauge length of a specimen pulled to failure (test piece 9 in Fig 56). These points are compared with values for the 4mm gauge width specimen (taken from Fig 55) extrapolated to $\epsilon_1 = 4.0$ which approximately represent the width and thickness strain that would be measured with no constraint from the test piece head. The two sets of curves are coincident for a value of axial strain of about 2.0, but diverge at lower and higher strains indicating increasingly anisotropic behaviour towards the failure site and towards the test piece head in the test piece pulled to failure.

4.3.3 R values of other alloys

The R values of the 3.3mm Ti-6Al-4V sheet after superplastic strain are compared with those of the other alloys investigated (all of initial thickness 2mm) in Fig 58.

The R values for the alpha/beta alloys in this comparison were measured after superplastic strain at the temperatures corresponding to equal phase proportions in each alloy.

The data for the Ti-6Al-4V are interpolated data calculated from the width and thickness strain curves in Fig 55a for the 4mm gauge width test piece, whereas the other points have been calculated directly from width and thickness measurements. The IMI 550 behaved in a very similar way to the Ti-6Al-4V, although with apparently little effect of test piece orientation. Less data were available for the other two alloys. However, the R values seemed to be relatively high and similar for the L and T orientations in each case.

4.4 Analysis of necking of Ti-6Al-4V during superplastic deformation

The computer programme predicted the development of a neck after each of a series of small strain increments in the uniform region of the gauge length. The analysis assumed that the neck strain rate was constant during each increment (see section 3), whereas, in practice, the neck strain rate would change continuously. Thus, the accuracy of the prediction of neck development depends on the size of the strain increment chosen and is reduced by increasing the strain increment size. The effect of different uniform strain increment size on predicted neck strain is shown in Fig 59 for an initial neck area of 0.95 (normalised with respect to the starting area of the uniform section of the gauge length). The true strain increment size was varied from 0.025 to 0.1. The difference only became significant at high total strains and for all subsequent analyses a strain increment of 0.05 was used.

Neck strain has been plotted against uniform strain in Fig 60 for initial (normalised) neck cross-section areas ranging from 0.5 to 0.99. In each case the uniform strain at the point of failure is indicated by the point at which the gradient becomes infinite. For neck areas of 0.5, 0.8, 0.9, 0.95 and 0.99 this occurred at uniform strains of about 60, 200, 300, 450 and 850% respectively. The solid line in Fig 60 indicates equal strain in the "neck" and uniform regions ie the inhomogeneity size is zero. The effect of strain on strain rate in the neck for an inhomogeneity of normalised area 0.8 is shown in Fig 61. The neck strain rate initially decreased slightly and reached a minimum at a strain of about 10%. The (constant) strain rate in the uniform region is shown by a solid line in Fig 61. The neck strain rate at zero strain is indicated by the point on the vertical axis. The load, normalised with respect to the initial load, is also shown in Fig 61.

The flow stresses in the uniform and necked regions are shown as a function of strain in Fig 62. Strain hardening and strain rate hardening caused the neck flow stress to rise more rapidly than the flow stress in the uniform region where the strain rate was constant.

Strain rate sensitivity m and strain hardening exponent n are given in Fig 63. The m values decreased with increasing strain, with the m value in the neck decreasing sharply as the failure point was approached. The n values in both the uniform and necked regions initially increased and passed through maxima. A small degree of strain softening at high strain is indicated in Fig 63 by negative n values. This was due to a slight misfit of the polynomial to the true stress-true strain data points. There is no experimental evidence of strain softening and the dashed line in Fig 63 indicates the more likely effect of strain on n .

The total strain hardening rate can be written as the sum of strain hardening and strain rate hardening terms:

$$\frac{d\sigma}{d\epsilon} = \left(\frac{\partial \sigma}{\partial \epsilon} \right)_{\dot{\epsilon}} + \left(\frac{\partial \sigma}{\partial \dot{\epsilon}} \right)_{\epsilon} \frac{d\dot{\epsilon}}{d\epsilon} = \frac{\partial \sigma}{\partial \epsilon} + \frac{\sigma m}{\dot{\epsilon}} \left(\frac{d\dot{\epsilon}}{d\epsilon} \right) \quad (1)$$

where σ is the flow stress and ϵ is the strain. This ignores any effects of increasing surface area and applies to isothermal conditions only. The values of each of the two terms on the right hand side of equation (1) were determined as a function of strain in the neck and plotted in Fig 64 along with the total strain hardening rate $d\sigma/d\epsilon$. This figure also includes $(\partial \sigma / \partial \epsilon)_{\dot{\epsilon}}$ for the uniform deformation. The term incorporating $d\dot{\epsilon}/d\epsilon$ is zero for this region. The value of $\partial \sigma / \partial \epsilon$ for the neck and uniform regions decreased with strain and would tend towards zero in the absence of strain softening. The value of $\frac{\sigma m}{\dot{\epsilon}} \left(\frac{d\dot{\epsilon}}{d\epsilon} \right)$ for the neck increased from close to zero and became dominant above about 85% uniform strain. Consequently the total strain hardening rate for a neck of initial (normalised) area decreased with strain, reached a minimum at about 70% strain and then increased rapidly.

5 DISCUSSION

5.1 Titanium alloy sheet

5.1.1 Superplastic properties and microstructures

All of the Ti-6Al-4V sheets tested were nominally of superplastic quality. However, Figs 7 and 8 show that for a given strain rate the m values varied significantly (by up to about 0.18). It is well known (117-119) that high strain rate sensitivity is favoured by fine grain size, but m values are also sensitive to grain size distribution, grain aspect ratio and small variations in composition (88,118,120,121). It is likely that all of these factors contributed to the variability of m value and flow stress between the different Ti-6Al-4V sheets examined here. The data in Figs 5-8 indicate the range of m values and flow stresses likely to be found in commercial superplastic quality Ti-6Al-4V sheet and suggest that a very detailed microstructural analysis would be required to predict the relative superplastic forming potential of various batches.

The superplastic properties of each of the alpha/beta alloys were qualitatively similar with peaks in m value in the strain rate range 10^{-4} s^{-1} - $3 \times 10^{-4} \text{ s}^{-1}$ (Figs 7,8,21,26) and an increase in grain size with superplastic strain (Figs 16,18,24,28). By contrast, the Ti-15V-3Cr-3Al-3Sn showed relatively low m values decreasing with increasing strain rate (Fig 10) and grain size decreasing with strain (Figs 34 and 35).

It is interesting to note that at a strain rate of $1.5 \times 10^{-4} \text{ s}^{-1}$ corresponding roughly to that of maximum m value the flow stress of each of the alpha/beta alloys at the appropriate temperature for equal phase proportions was practically identical, whereas the flow stress of the Ti-15V-3Cr-3Al-2Sn at this strain rate was much higher (Fig 65). However, it is unlikely that the similarity of flow stress in the alpha/beta alloys is of particular significance in view of the difference in grain size between the alloys (Figs 16,24,28). At the strain rate corresponding to maximum m the deformation is likely to result largely from grain boundary sliding (1,2,18,21,32,37,122), but the average grain boundary shearing rate will be lower for the finer grained IMI 550 than for the other two alpha/beta alloys. This would tend to reduce the flow stress. However, the IMI 550 contains a relatively high proportion of molybdenum, which compared to other alloying elements, has a low

diffusivity in beta titanium reducing the effective lattice and grain boundary diffusion rates (44,45,55,56). Paton and Hall (56) reported that Mo additions to Ti-6Al-4V increased the superplastic flow stress. The flow stress is also likely to depend on shear modulus (123), which in turn varies with the composition. Thus, the flow stress at any particular strain rate or phase proportion is determined by a combination of factors and it should not be inferred from Fig 65 that there is a unique value of flow stress for alpha/beta titanium alloys at the 50/50 phase temperature and at the strain rate corresponding to maximum m .

It has been established (88,124,125) that the increase in flow stress of Ti-6Al-4V during superplastic strain (Fig 9) is almost entirely due to grain coarsening. It can be seen from Fig 9 that the flow stress did not increase above about 1.7 true strain, consistent with the stable grain size at this strain (Fig 18). Grain growth during superplastic strain also occurred in the other alpha/beta alloys (Figs 24 and 28) and the strain hardening in these cases can presumably be explained in the same way. The small amount of strain softening that occurred in the Ti-8Al-1Mo-1V was possibly associated with the break-up of the directional microstructure, as reported for extruded 60/40 brass (126) and Cu-P alloys (127), or may have been due to inaccurate estimation of the instantaneous cross-section area towards the end of the superplastic test (the Ti-8Al-1Mo-1V test pieces pulled at 910°C either failed or necked substantially).

The strain hardening rate (Fig 22) and, therefore the rate of grain growth, was lower for the IMI 550 at the 50/50 phase proportion temperature (900°C) than for the Ti-6Al-4V or the Ti-8Al-1Mo-1V at 925°C and 1010°C respectively. This was partly due to the relatively low temperature and partly as a result of the low diffusivity of the Mo in the IMI 550. The grain size stability of the IMI 550 was also reflected in the increase in m value with superplastic strain up to 130% (Fig 10). This increase in m was probably related to the microstructure becoming more equiaxed, which appeared to override the tendency of the simultaneous grain coarsening to reduce m . Higher superplastic strains (128) would presumably have decreased the m value as the effect of grain coarsening became dominant.

Superplastic strain produced subgrains and refined the grain size of Ti-15V-3Cr-3Al-3Sn, whereas static annealing, which occurred in the test piece head, caused rapid grain growth. It has been reported (129,130)

that sliding can occur in subgrain boundaries in aluminium, with the subgrain sliding preceded by a gradual increase in misorientation between adjacent subgrains, leading to the formation of high angle boundaries. This is analogous to the method proposed by Hayden et al (122) for the break-up of fibrous microstructures during superplastic strain, whereby high aspect ratio grains are divided along their lengths by subgrain structures. Alternatively, Griffiths and Hammond (8), working on beta titanium alloys and beta brass, suggested that subgrain walls form continuously by dislocation climb with equiaxed grains and grain refinement resulting from the migration of subgrain boundaries and the pinching-off of grains. Therefore, it seems likely that the reduction in grain size in Ti-15V-3Cr-3Al-3Sn occurred by the development of a subgrain dislocation structure with sliding eventually taking place on the subgrain boundaries. The surface rumpling (Fig 33) indicated that grain boundary sliding was occurring. The decreasing grain size resulted in an increase in strain rate sensitivity at 910°C (Fig 10) up to 300% strain and a small degree of strain softening.

After superplastic deformation and furnace cooling the Ti-8Al-1Mo-1V contained approximately 50% transformed beta (Fig 28c), whereas the annealed material taken from the test piece head contained only intergranular beta and no secondary alpha. There was also evidence of this behaviour in the Ti-6Al-4V (Figs 16b and c). The likely explanation for this is simply that the cooling rate in the test piece heads, which were covered by the grips, would have been lower than that in the gauge lengths. Consequently, the beta-alpha phase transformation in the test piece heads occurred with migration of the alpha/beta boundaries and enlargement of the existing alpha grains. In the gauge lengths the greater undercooling and slightly larger grain size favoured the nucleation and growth of Widmanstatten alpha plates (131) inside the beta grains. This explanation is consistent with the grain size decreasing in the order Ti-8Al-1Mo-1V > Ti-6Al-4V > IMI 550 after superplastic strain. Higher superplastic strains and, hence larger grain sizes, can result in the formation of Widmanstatten alpha plates in IMI 550 also (132).

The highest m values measured were for the Ti-8Al-1Mo-1V at 1010°C (Fig 26). However, for commercial forming operations other factors must be taken into consideration when selecting an alloy. A low forming temperature is desirable so that cheap die materials can be used,

die life prolonged and contamination minimised. An m value which is roughly constant over a wide range of strain rates is also an asset as some variation in strain rate is to be expected in a complicated forming. The variation in m value for the IMI 550 (Fig 21) over the strain rate range was relatively small, although the peak values were slightly lower than those of the Ti-6Al-4V or Ti-8Al-1Mo-1V. By comparison, the m value of the Ti-8Al-1Mo-1V was highly strain rate sensitive at 1010°C (Fig 26). A further disadvantage of the Ti-8Al-1Mo-1V is that for this alloy the phase proportions are relatively sensitive to temperature (133). Thus, close temperature and strain rate control would be required in order to maintain optimum superplastic properties in this material.

The advantages of the high lattice diffusivity of the beta alloy Ti-15V-3Cr-3Al-3Sn in promoting high strain rate sensitivity were outweighed by the large initial grain size and the rapid grain coarsening that occurred during annealing before superplastic deformation. Consequently, the Ti-15V-3Cr-3Al-3Sn appeared to be the least attractive of the alloys for superplastic forming showing relatively low strain rate sensitivity and an irregular surface after forming. Low total elongations are reported (8,134) for beta titanium alloys during superplastic deformation.

5.1.2 Room temperature tensile properties after superplastic deformation

The most marked effect of superplastic strain on the room temperature tensile properties of Ti-6Al-4V was the reduction of 0.2PS and TS by about 10% associated purely with the thermal cycle. This was caused by recrystallisation (134,135) that removed the worked structure resulting in larger, more equiaxed grains (136). By comparison the effect of superplastic strain on room temperature strength, which was the result of grain coarsening (137), was small. These observations on the effects of annealing at the forming temperature and superplastic strain agree with other uniaxial data (62,134,137,138) on Ti-6Al-4V. The results for biaxially formed material (139-141) show similar trends, but more scatter, probably reflecting slightly different forming and testing techniques. The British standard (142) for Ti-6Al-4V sheet specifies minimum 0.2PS and TS levels of 900 and 960 MPa respectively. The strength loss inherent in superplastic forming often means that formed components do not reach these minimum values.

The anisotropy in room temperature properties was due to crystallographic texture. The anisotropy was particularly pronounced for the edge textured 3.3 mm Ti-6Al-4V sheet (Figs 11 and 12 and Table 1) for which the modulus and strength were greater in the transverse direction than in the longitudinal direction.

The IMI 550 alloy also showed some loss in room temperature strength after annealing at the forming temperature (Table 2). The additional effect of superplastic strain on strength was slight (Fig 23). The standard heat treatment for IMI 550 involves an air cool from the solution temperature (900°C) and ageing at 500°C to produce a fine alpha precipitation. This ageing response is enhanced by increasing the cooling rate as more metastable beta phase is retained and this phase is richer in alpha stabilising elements. The post-forming cooling rate of 25°C min⁻¹ was sufficient to give a 0.2PS and TS increase of about 10% on ageing. Resolution treatment at 900°C, air cooling and ageing can increase room temperature strength at least to that of the as-received (mill annealed) material and usually beyond (110,128,132). Such a heat treatment applied to a formed sheet component may, in practice, cause contamination and distortion problems. However, since the optimum forming temperature is the same as, or close to, the solution temperature the same result can be achieved simply by increasing the cooling rate after the forming operation and subsequently ageing. It is likely that cooling rates higher than the 150°C min⁻¹ used in this work could easily be achieved in practice.

The as-formed 0.2PS and TS advantage of the IMI 550 over the Ti-6Al-4V was about 5% for a cooling rate of 25°C min⁻¹. Ageing of the IMI 550 at 500°C after forming further increased the 0.2PS and TS by 5%. The highest strengths achieved in the IMI 550, by cooling at 150°C min⁻¹ and ageing, gave 0.2PS and TS levels typically 18% greater than those of the as-formed Ti-6Al-4V.

Tables 2-4 indicate that superplastic strain reduced total elongations of IMI 550 at room temperature. This contradicts other results (110) on IMI 550 sheet, although Duffy (128) also reported some loss of ductility. It is possible that slight surface contamination by oxygen occurred during superplastic testing accounting for the reduction in elongation.

The as-formed strength of the Ti-8Al-1Mo-1V was low by comparison with the Ti-6Al-4V and the IMI 550, although modulus and elongation were both higher.

The strength and ductility of the Ti-15V-3Cr-3Al-3Sn alloy were both reduced by superplastic deformation (Table 6). Ageing to precipitate alpha would presumably strengthen the alloy considerably and increase the modulus after superplastic strain. There appears to be little advantage in superplastically forming this alloy for commercial applications in view of its cold formability (143) and competition from the highly superplastic alpha/beta alloys.

5.1.3 Texture

The texture variation through the thickness of the 3.3 mm Ti-6Al-4V sheet (Fig 13) was indicative of different deformation conditions at the sheet centre and the sheet surface during rolling, although this was not reflected in the microstructure. The $(0002)_\alpha$ and $(110)_\beta$ pole figures were similar suggesting that the Burgers orientation relationship $(0002)_\alpha // \{110\}_\beta$, $\langle 11\bar{2}0 \rangle_\alpha // \langle 111 \rangle_\beta$ was generally obeyed. However, examination of the $(10\bar{1}0)_\alpha$ pole figures 13b,e and 14b,e shows that the Burgers variant with $(0002)_\alpha$ parallel to the T direction was preferred. This occurred because the transverse basal texture component develops during sheet rolling (112), whereas the remaining variants can only form by transformation during cooling.

Superplastic strain of Ti-6Al-4V caused a reduction in texture intensity at both the sheet surface and the sheet centre (Fig 15) consistent with grain boundary sliding throughout the material. The peak intensities in the alpha and beta pole figures were similar both before and after superplastic strain suggesting that the grain rotation was not restricted to one phase. These observations of the effect of strain on the alpha texture are in agreement with other work (59,138) on uniaxial superplastic deformation of Ti-6Al-4V and similar results were obtained for IMI 550 (132). Biaxial forming of Ti-6Al-4V (139-141) appears to reduce the alpha phase texture intensity in a similar way. Much less attention has previously been paid to the effect of superplastic strain on beta textures. McDarmaid et al (60) reported that in highly textured Ti-6Al-4V bar beta texture was almost completely removed by 300% superplastic strain, whereas the alpha texture was retained, although with a reduced intensity. It was suggested (61) that this was caused by a

preference for sliding in the beta/beta boundaries (51). Ma and Hammond also found evidence (44) by high temperature electron microscopy that superplastic deformation in Ti-6Al-4V is associated with much larger distortions of the beta phase than of the alpha phase. Therefore, these results (44,60) differ from the work reported here and the precise role of the two phases remain unclear.

The Ti-6Al-4V sheets investigated had a wide range of initial textures, but there was no evidence of any influence of texture on superplastic properties. This agreed with other work on Ti-6Al-4V (59,88) and was consistent with the assumption that the superplastic strain was largely accommodated by grain boundary sliding.

The texture of the Ti-15V-3Cr-3Al-3Sn sheet Fig 32 was either $\{110\}\langle 110 \rangle$ ($\{110\}$ planes parallel to the sheet plane and $\langle 110 \rangle$ parallel to the rolling direction) or $\{110\}\langle 100 \rangle$. Neither of these textures is a common bcc rolling texture (113,114). Superplastic strain apparently did not reduce the texture intensity. This was consistent with the occurrence of slip deformation, which led to the formation of subgrains.

5.1.4 Activation energy

Estimates of the activation energy Q for superplastic flow in Ti-6Al-4V at 850-925°C were made by Arieli et al (124) using the Dorn creep equation (123):

$$\dot{\epsilon} = \frac{AGb}{kT} \left(\frac{b}{d}\right)^p \left(\frac{\sigma}{G}\right)^n D_0 \exp - \frac{Q}{RT}$$

where A and p are constants, G is the shear modulus, b is the Burgers vector, k is Boltzmann's constant, T is the absolute temperature, R is the gas constant, d is the grain size, D_0 is a frequency factor and n is the stress exponent ($n = 1/m$). The estimated values of Q (124) ranged from 224-299 kJmol⁻¹ and increased with superplastic strain.

An attempt was made to calculate the activation energy for superplastic deformation for the Ti-8Al-1Mo-1V alloy based on the data in Fig 25 and using a similar method to that of Arieli et al (124). The flow stress was determined for each test temperature for a strain rate of 10⁻⁴ s⁻¹, which was assumed to lie within region II in each case. A plot of $R \ln \sigma^{-n} G^{n-1}$ against $1/T$ was made using the appropriate values of n and G for each temperature. However, the points did not fall on, or even near a straight line and a meaningful estimation of Q could not be

made. Instead, it is useful to consider the sources of error in the determination of Q .

Very little information is available for the shear modulus of titanium alloys at high temperature. The estimations of Q for the Ti-8Al-1Mo-1V alloy and that for Ti-6Al-4V (124) used extrapolated shear modulus data (144) for alpha phase titanium, which is likely to be a source of significant error.

The calculated value of Q is particularly sensitive to the strain rate sensitivity and its temperature dependence. There are several ways (1,145) in which the m value can be estimated from load displacement curves and these can yield significantly different results (145). Arieli and Rosen (145) claim that the m value for Ti-6Al-4V at 900°C is really constant at 0.5 throughout region II. However, the overwhelming weight of evidence (1) indicates that this is not the case and that m phases through a maximum in region II. The m values given in this thesis were calculated using one technique only, as described in section 3.2, in order that the results should be comparable. In the determination of Q for Ti-6Al-4V Arieli et al (124) assumed that the m value did not vary (for a particular strain rate) in the temperature range 850-925°C. Clearly, a similar assumption is not justified for the Ti-8Al-1Mo-1V alloy (Fig 26). Other workers (8,128,146) have chosen simpler rate equations to describe superplastic flow and have ignored either the influence of shear modulus or the temperature variation in m when calculating the activation energy.

Further sources of inaccuracy are the variability in test piece machining and microstructure. The small variations in m , recorded during repeat tests on material of the same batch could not be entirely accounted for by the random inaccuracies in load measurement.

A number of different models have been formulated for superplastic flow (reviewed in section 2.1 and in refs 1,35,40,147) with activation energies corresponding to grain boundary diffusion, lattice diffusion or phase boundary diffusion. Measured values of Q (reviewed in ref 1) are generally of the order of that for grain boundary diffusion in some systems and of the order the activation energy for lattice diffusion in others. For titanium alloys the reported activation energies (55,124, 128,141,148,149) seem to be greater than the activation energies for self diffusion in either the alpha (42) or the beta (150) phases and

certainly greater than the activation energies for grain or phase boundary diffusion. This is not consistent with the models of superplastic deformation and the scatter in the results highlights the difficulties associated with the experimental determination of Q . It has been suggested (43) that dynamic recrystallisation in two phase titanium alloys might explain the anomalously high activation energies, but no experimental evidence has been put forward in support of this.

5.2 Superplastic deformation of sheet test pieces machined from Ti-6Al-4V bar

5.2.1 Microstructure and flow stress

Most models of superplastic flow eg refs 22,28,36,38 invoke grain boundary sliding with accommodation at the grain corners by diffusion or dislocation motion. The microstructural observations in section 4.2 indicate that a model of this type applies to Ti-6Al-4V, although the precise role of each phase is not clear. In Fig 48 whole grains have moved relative to their neighbours and emerged from the test piece surface. There was no evidence of any intergranular voids, apart from in the necked regions where strain rates were locally high and the deformation non-superplastic.

For grain boundary sliding and grain rotation to occur in the aligned regions of the Ti-6Al-4V microstructure the contiguous alpha grains must slide apart as they cannot readily rotate or move in unison. A consequence of this is the break up of the contiguous alpha phase, as shown in Fig 44. Thus, the flow stress in the area of aligned microstructure where many alpha/alpha boundaries must be sheared is likely to be higher than in the more equiaxed areas where more shear along beta/beta boundaries can occur. The reason for this is the relatively low lattice and grain boundary diffusion rates in the alpha phase (see chapter 2). A further factor, tending to increase the flow stress in the aligned regions, is that the aligned grains are likely to be of a similar orientation within each string of aligned grains with relatively narrow grain boundaries and possibly even some low angle boundaries. Evidence for this is that the alpha phase texture is much more intense in the aligned regions than in the non-aligned regions (59) and the boundaries between the contiguous, aligned grains are difficult to detect optically as they etch slowly.

Quantitative analysis of the break-up of the contiguous alpha phase is not easy because measurements have to be taken from several

areas that may not have been similar before straining. However, the measurements of contiguous alpha phase aspect ratios in ref 61 and the microstructures in Fig 44 show that the aligned microstructure does break up during straining and it is assumed that those areas most strongly aligned before deformation are those which deform at the lowest rate.

For the TL and ST orientations the areas of aligned microstructure remained relatively undeformed during testing (Figs 42 and 43) compared with those areas where the two phases were homogeneously distributed. Therefore, for these two orientations the measured flow stress corresponded roughly to that of the flow stress of the equiaxed areas. For the LT orientation, where the tensile axis was parallel to the banding direction, the bands of aligned microstructure were constrained to deform at the same rate as the surrounding areas and a relatively high flow stress was measured (Fig 49). The orientation with the lowest flow stress was consistently the ST orientation. This can be explained by the observation that, although the alpha phase was most strongly aligned in the L direction, there was also some alignment in the T direction (Figs 39 and 40) resulting in flow stresses in the order $\sigma_L > \sigma_T > \sigma_S$.

The multiple necking that occurred in the ST orientation test pieces pulled at 975°C (Fig 38) appeared to be the result of strain localisation in bands where the alpha phase proportion was low at 975°C. Deformation of these areas would be favoured because sliding of beta/beta interfaces only would be required. The final beta grain size in these areas was much larger than in adjacent regions where the alpha phase retarded the beta grain growth. Presumably, this type of necking did not occur in test pieces of other orientations because the beta phase bands were not suitably oriented in relation to the tensile axis. It is shown schematically in Fig 41 that the regions of aligned and non-aligned microstructure were roughly planar in the LT plane. Thus, at 975°C the beta rich regions, would lie in planes perpendicular to the S direction.

The variation in phase proportion suggested small variations in chemical composition. Electron probe microanalysis of this material (59) has not revealed any such variations. However, analysis (151) of similar hot rolled Ti-6Al-4V bar has detected a slightly higher vanadium concentration in the bands of more heavily deformed microstructure than in other areas, possibly accounting for the origin of the banding. Therefore, it is likely that the chemical composition of the 55 mm bar used here was not uniform, although this may not be detectable throughout.

It is to be expected that the beta rich areas would also exhibit lower flow stresses than those areas low in beta phase for forming temperatures lower than 975°C. Thus, as well as grain alignment, a second factor that may contribute to the irregular superplastic strain distribution is a variation in phase proportion.

5.2.2 Strain rate sensitivity

The m values at each test temperature were lower for the high strain rate tests ($\dot{\epsilon} = 1.5 \times 10^{-3} \text{ s}^{-1}$) and Fig 36 shows that, for this strain rate, the rate of neck development was higher than for the corresponding tests at the lower strain rate of $3 \times 10^{-4} \text{ s}^{-1}$, particularly for the ST and LT orientation test pieces. The tests at 975°C resulted in failure at a low strain, typically 150%, consistent with the low strain rate sensitivities at this temperature (Fig 50). The m values increased with temperature and reached a maximum at 925°C, when the phase proportions were roughly equal. Above this temperature rapid beta grain growth caused a sharp reduction in m value.

For the TL and ST orientation tests the strain tended to be localised in the areas of non-aligned microstructure. Therefore, the m values measured on the TL and ST orientation test pieces corresponded approximately to the m values of the non-aligned microstructure, whereas for the LT orientation tests each type of microstructure deformed at the same rate and an "average" m value was recorded. Thus it appears that the strain rate sensitivity of the aligned microstructure was lower than that of the non-aligned microstructure.

5.2.3 R values

5.2.3.1 TL orientation

In the TL orientation the banding direction was across the gauge width and R values measured on the bands, such as points A1-A3 in Fig 52, were less than 1.0. It might be anticipated that the R values measured at other points along the gauge length eg B1-B4 in Fig 52 would be closer to 1.0 reflecting the more isotropic microstructures in these areas. However, this was not the case and for test temperatures in the range 800-925°C lower R values were measured in the non-banded areas. The results on superplastic anisotropy of rolled Ti-6Al-4V sheet (section 4.3) showed that R values measured after superplastic strain were influenced by the initial test piece shape. If a test piece of low

gauge length to width ratio was used then constraint by the test piece heads restricted the strain in the width direction and low R values were measured. Increasing the gauge length or reducing the width minimised this geometrical effect. In the same way, R values measured in the non-banded areas of the Ti-6Al-4V used here were influenced by the adjacent, less deformed areas. An illustration of this effect is given in Fig 66, which shows a segment of the gauge length of TL orientation test piece A9 pulled at 875°C at a strain rate of $1.5 \times 10^{-3} \text{ s}^{-1}$. The effect of microstructure in the banded region was to resist strain in the width direction and an R value of less than 1.0 was measured. To the right of this region constraint by both the test piece head and the block of strongly banded material has resulted in very low R values (0.4), even though the microstructure here was probably relatively equiaxed. To the left of the banded area the constraint becomes progressively less and the R values rise from 0.51 to 0.67. Thus, the R value measured depends strongly on the position in which the measurement is made, which partly explains the wide range of R values determined for the TL orientation for test temperatures up to 925°C. This shallow "necking" or waisting of the test piece was most pronounced for the low strain rate, 875°C test (Fig 36c) and the difference in R value between banded and non-banded areas was greatest for these test conditions (Fig 51). It is shown in Figs 36a-c that for each test temperature the final shape of the TL test pieces was much less irregular for the high strain rate tests than for the low strain rate tests. Therefore, the geometrical constraint effects invoked to explain the lower R values in the non-banded areas are likely to be less important. Fig 51b, for the high strain rate tests, does not show the trend of low R values in non-banded areas which is exhibited for the low strain rate tests (Fig 51a).

The TL test pieces pulled at 975°C showed higher R values, reflecting the low alpha phase proportion and much less directional microstructure at this temperature.

Any contribution of alpha phase texture to anisotropy above 800°C is likely to be small compared to that at room temperature and no effects of texture on R value are expected for the TL orientation because the tensile axis is parallel to the c axis and, therefore, alpha phase prism slip cannot easily occur and neither thickness strain nor width strain is preferred.

5.2.3.2 ST orientation

For this test orientation the alpha grain alignment acted to reduce the thickness strain during superplastic deformation and for test temperatures of 875°C and 925°C all the R values were greater than 1.0. However, the test piece surfaces were very irregular, as shown in Fig 37a. Thickness measurements were taken on the surface protrusions and in the depressions at random and, therefore, a wide variation in R was recorded and any trend of R with strain ϵ_1 was masked.

Results (65) on round test pieces from the Ti-6Al-4V bar pulled in the S direction show that for a test temperature of 800°C the sense of the strain anisotropy reverses for strain rates above about $4 \times 10^{-4} \text{ s}^{-1}$. For the sheet test pieces of ST orientation the texture of the bar is such that $\langle 11\bar{2}0 \rangle$ prism slip would increase the thickness strain at the expense of the width strain and reduce R values ie the effects of texture would oppose those of the banding in the microstructure. Fig 51b shows that, for a strain rate of $1.5 \times 10^{-3} \text{ s}^{-1}$ at 800°C, R values ranged from 0.9-1.1 indicating that the effects of texture were sufficient to cancel those of microstructure.

The test pieces pulled at 975°C showed relatively low elongation and incipient necking at several locations. As each neck began to develop the deformation tended towards plane strain and the thickness strain increased relative to the width strain. This reduced the local R value, as shown in Fig 38. However, for a neck to form across the test piece perpendicular to the tensile axis the Von Mises yield criterion predicts (152) that the axial stress required is $\sigma = 2\sigma_f/\sqrt{3}$, where σ_f is the flow stress. Therefore, a neck of this geometry cannot develop because the applied stress necessary exceeds that required for general deformation along the gauge length. Under these circumstances final necking occurs along a line inclined to the tensile axis and along which the strain rate can approach zero. For isotropic material this necking line makes an angle of 54.7° with the tensile axis. In both test pieces in Fig 38 the growth of necks running perpendicular to the tensile axis has been arrested at an early stage and neck development along a line inclined to the tensile axis has been favoured. This type of localised necking also occurred in the 800°C ST orientation test pieces for which the strain rate sensitivity was relatively low (Fig 50).

5.2.3.3 LT orientation

The LT orientation test pieces showed a more uniform strain distribution along the gauge length than the corresponding TL and ST test pieces. Consequently, there was less variation in R for each test temperature (Fig 51). All the R values measured for this orientation were less than 1.0 as a result of the small degree of grain alignment or grain elongation in the T direction.

The R values corresponding to the 800°C tests at a strain rate of $1.5 \times 10^{-3} \text{ s}^{-1}$ were lower than those for the other test temperatures at this strain rate. This was due to the influence of texture, which for this orientation, tended to reduce R.

5.2.3.4 Comparisons with other work

The R value results on the sheet test pieces are, for the most part, in agreement with the diametral strain ratio measurements made (62,64,65) on the round test pieces machined from the same bar. An exception to this was the round test pieces pulled in the S direction; those regions of the gauge length containing aligned microstructure were found (64,65) to deform more anisotropically than the non-aligned areas. It was not possible to discriminate between aligned and non-aligned areas in R value measurements made on the ST orientation sheet test pieces. However, R values for the different microstructures were recorded for the TL orientation tests and showed, on average, more anisotropic deformation in the non-aligned areas (Fig 51a) ie the opposite of the behaviour predicted by the round test piece results. This can be explained by the constraints present in the sheet test pieces, as discussed in section 5.2.3 and illustrated in Fig 66. Similarly, the results for the S orientation round test pieces showed (64,65) anisotropy at 970°C in the opposite sense to that indicated in Fig 51 for the ST orientation sheet test pieces at 975°C. The reason for this is the unusual necking behaviour related to the beta rich phase bands, which influenced the R value measurements sufficiently to reverse the apparent sense of anisotropy. This type of necking, approaching plane strain, can only occur in sheet test pieces. Thus, the use of sheet rather than round test pieces does affect the high temperature anisotropy in this material.

Superplastic anisotropy of rolled Ti-6Al-4V sheet of various thicknesses is discussed in section 5.3. The sheet microstructures were much less directional than that of the bar. Nevertheless, R values as

low as 0.5 were measured with the lowest R values occurring in the sheet with the highest alpha phase aspect ratio. Thus, although the rolled sheet microstructures were homogeneous ie without bands of aligned and non-aligned microstructure, the superplastic anisotropy was similar in nature to that of the bar material.

5.3 R values of Ti-6Al-4V sheet

5.3.1 Effect of sheet thickness

It is apparent from Figs 53 and 54 that superplastic strain occurred most easily in the direction in which the average alpha grain width was smallest, and that the anisotropy was most pronounced for high alpha phase aspect ratios. The thickness effect shown in Fig 53 (R decreasing with increasing sheet thickness) was therefore observed because the thinner sheets exhibited the most isotropic microstructures, presumably as a result of recrystallisation during rolling or during annealing at the superplastic temperature before straining. The effect of superplastic strain in the L direction on the LS section microstructure of the sheet of initial thickness 3.3 mm is shown in Fig 18 (compare with Fig 17a) and Fig 19 shows that the alpha phase aspect ratio decreased with superplastic strain. This is reflected in the increase of R with strain (Fig 53).

This behaviour is consistent with the superplastic strain anisotropy shown by heavily banded Ti-6Al-4V bar. It is significant that such anisotropy can occur in material processed for isotropic superplasticity and emphasises the need for sheet production techniques capable of achieving equiaxed microstructures.

5.3.2 Effect of test piece geometry

The curves in Fig 55 show that the constraints due to the test piece heads were more pronounced for larger gauge widths (Fig 55a) and smaller gauge lengths (Fig 55b). To ensure that geometrical constraints are minimised and natural width and thickness strains are measured Fig 55a indicates that the length to width ratio of the gauge length should be 1.25:1 or greater (corresponding to the test piece of gauge length 10 mm and gauge width 8 mm in Fig 55a) and according to Fig 55b it should be at least 1.6:1 (corresponding to the test piece of gauge length 25 mm and gauge width 16 mm in Fig 55b). Therefore, for the 3.3 mm sheet used in this investigation at 925°C and a strain rate of $3.1 \times 10^{-4} \text{ s}^{-1}$ a test piece aspect ratio of 1.6:1 or greater is required for the determination

of R as a unique function of strain. The limiting superplastic strain to which meaningful R values can be measured depends on the strain at which the variation in strain rate along the gauge length becomes significant. For specimen 9 in Fig 56d this corresponded to a true strain of about 2.0 (639%), although this in turn depends on the strain rate sensitivity of the material and the machining tolerance of the test piece as any waisting or irregularity before testing will cause some localisation of strain.

For material of lower strain rate sensitivity than the Ti-6Al-4V used in this work the influence of the test piece head would be greater and the tapering of the gauge length more pronounced as the tendency for the strain to localise in the narrowest region of the gauge length would be stronger. In such cases the use of a high aspect ratio test piece (large length to width ratio) becomes increasingly important. This is illustrated by the R value measurements made on the Ti-15V-3Cr-3Al-3Sn (Fig 58), which indicated some anisotropy despite the fact that the starting microstructure in this alloy appeared to be completely isotropic (Fig 34a). The deviation of R from unity was therefore due to geometrical constraint even though the gauge aspect ratio was sufficiently high (1.6:1) to avoid significant end effects in the Ti-6Al-4V sheet test pieces. It can be seen from Fig 33 that the strain variation along the gauge length of a typical Ti-15V-3Cr-3Al-3Sn test piece was considerable and that the centre of the gauge length was not free from the influence of the test piece heads.

The standard procedures (153-155) for room temperature R value measurement specify test pieces of high aspect ratio, typically 10:1, and of length approximately 200 mm. This aspect ratio is sufficient to minimise the geometrical effects noticed during superplastic R value measurement (Fig 55), but in practice these standard test pieces would be too long for uniaxial superplastic testing when strains of several hundred percent may be required (the hot zone of a laboratory furnace being typically 100 mm long). The test pieces used in other work on the superplasticity of Ti-6Al-4V eg refs 55,124,148,156 have gauge aspect ratios of 1.6:1 or greater and therefore, based on the criterion established in this investigation, would have been suitable for R value determinations.

The choice of a standard test piece for the measurement of R values after superplastic strain is therefore a matter of compromise; a high aspect ratio is required in order to minimise end effects, but there are practical restrictions on the gauge length and width.

The divergence of the two sets of curves in Fig 57 at $\epsilon_1 < 1.5$ arose because the width and thickness strain measurements on the failed specimen were made close to the end of the gauge length. Further along the gauge length ie at higher ϵ_1 the curves converged as the constraint imposed by the test piece head became less pronounced, but for $\epsilon_1 > 2.0$ the curves diverge again. It is likely that as the shallow neck developed during testing the local strain rate increased such that deformation by slip became significant (61,64,65). It was shown in Fig 15 that the texture intensity of the alpha phase in Ti-6Al-4V decreases slowly with superplastic strain. Therefore, deformation of Ti-6Al-4V at the superplastic temperature by slip even following a certain amount of superplastic strain can be influenced by the initial alpha texture. The 3.3 mm sheet had a basal edge texture (Fig 13) and therefore through-thickness slip was preferred in L orientation test pieces. This is consistent with the behaviour shown in Fig 57 for $\epsilon_1 > 2.0$. It is expected that if these high strains could be achieved in an untapered test piece eg by remachining a parallel gauge length at $\epsilon_1 = 2.0$ then ϵ_w and ϵ_t would lie on the extrapolated (dashed) curve in Fig 57.

Another feature of the curves in Figs 55 is that anisotropic straining only occurred for $\epsilon_1 < 1.0$. At higher strains ϵ_w and ϵ_t lie parallel to the line depicting isotropic behaviour. The gradient $d\epsilon_w/d\epsilon_1$ for the 4 mm gauge width in Fig 55a increased from 0.41 at $\epsilon_1 = 0.5$ to close to 0.5 (isotropy) at $\epsilon_1 = 2.0$. This behaviour, representing R value increasing with strain, is related to the reduction in alpha phase aspect ratio on the TS section (as shown in Figs 17-19 for the LS section).

The definition of R based on total strains, $R = \epsilon_w/\epsilon_t$, means that R values measured after high superplastic strains are influenced by the anisotropic strain which is accumulated for ϵ_1 less than about 1.0. Consequently, even when the microstructure has become equiaxed and the material behaves isotropically this is not reflected in the R value which remains less than 1.0. An alternative definition (106) of R using incremental strains, $R' = d\epsilon_w/d\epsilon_t$, provides a more useful description of the normal anisotropy in those cases where R changes with strain.

R values calculated from Fig 55a (using data from the 4 mm gauge width test pieces) are plotted against strain in Fig 67. Approximate values of R' calculated from R are also given. The curve for R' is initially steeper than that for R and approaches $R' = 1.0$ which indicates the nearly isotropic behaviour of the material at high strain. Fig 67 also includes the average R' value $\bar{R}' = (R'_L + R'_T)/2$.

The superplastic flow stress data (section 4.1) for each of the alloys indicate very little anisotropy in the plane of the sheet and the L and T orientation data for the IMI 550 sheet are almost identical (Fig 20). The flow stress in the two orthogonal directions in the plane of the sheet σ_L and σ_T can be related (106) to the two corresponding R values R'_L and R'_T as follows:

$$\frac{\sigma_L}{\sigma_T} = \sqrt{\frac{R'_L(1 + R'_T)}{R'_T(1 + R'_L)}}$$

Substitution of $R'_L = 0.6$ and $R'_T = 0.5$ from Fig 67, which gives data for the Ti-6Al-4V sheet, produces $\sigma_L/\sigma_T = 1.06$. Thus, even though the microstructure is measurably anisotropic in the plane of the sheet (Fig 54), the superplastic flow stress is expected to be only marginally greater in one direction. The IMI 550, for which the flow stress at 900°C was isotropic in the plane of the sheet (Fig 20), also showed very little variation in R between the L and the T directions (Fig 58).

5.4 Application of uniaxial data to hemisphere forming

Plastic anisotropy in sheet metal can, in certain circumstance, improve cold formability (106,157,158) and, along with the strain hardening exponent n , the plastic strain ratio R is often measured to determine cold forming characteristics. For cold sheet forming the normal anisotropy is particularly pronounced for strongly textured hcp alloys (105, 107) where thinning of the sheet can be inhibited by appropriate orientation of the slip systems. Under these conditions "texture hardening" (159) occurs and critical thickness strains for failure (107) can only be obtained at relatively high total strains.

Similarly, superplastic formability can be characterised by R values, strain hardening and strain rate sensitivity. However, under superplastic conditions the strain is accommodated largely by grain boundary sliding and slip and crystallographic texture may be of much

less importance. Consequently, flow stresses are mainly dependent on grain size and shape and anisotropic flow during forming is due to directionality in the microstructure. The R value gives an indication of the through-thickness strength of the sheet relative to the in-plane strength. Biaxial stretching is equivalent to uniaxial thinning. Therefore, the main effect of R on bulge forming operations is to influence the gas pressure required for deformation.

This section describes the application of the strain hardening, strain rate sensitivity and R value data determined uniaxially on 3.3 mm Ti-6Al-4V sheet to the forming of a hemisphere using the concepts of equivalent stress and equivalent strain. The influence of each of these parameters is considered and an optimum pressure cycle is given for a constant strain rate is at the pole of the dome. Although the hemisphere has several practical applications and has already been the subject of analysis (160-165) the effects of superplastic anisotropy have so far not been considered.

A statement of the equivalent stress function σ_e derived (106) from Hill's general analysis (166) for plastically anisotropic material is:

$$\sigma_e = \sigma_{f1} = \left[\frac{R_1(\sigma_1 - \sigma_2)^2}{1 + R_1} + \frac{R_1(\sigma_2 - \sigma_3)^2}{R_2(1 + R_1)} + \frac{(\sigma_3 - \sigma_1)^2}{1 + R_1} \right]^{\frac{1}{2}} \quad (2)$$

where σ_1, σ_2 , and σ_3 are the principal stresses with σ_3 acting through the thickness, σ_{f1} is the flow stress in the X_1 direction and R_1 and R_2 measured along the X_1 and X_2 axes respectively are:

$$R_1 = \frac{d\epsilon_2}{d\epsilon_3} = R'_L \quad ; \quad R_2 = \frac{d\epsilon_1}{d\epsilon_3} = R'_T$$

If thin shell properties are assumed (162) then σ_3 is small and can be ignored. At the pole of a hemisphere $\sigma_1 = \sigma_2$ and equation (2) becomes:

$$\sigma_e = \sigma_L = \sigma_1 \left[\frac{R'_L + R'_T}{R'_T(1 + R'_L)} \right]^{\frac{1}{2}} \quad (3)$$

where σ_L is the flow stress in the L direction.

The tangential stress at the pole of a hemisphere is given by:

$$\sigma = \frac{Pr}{2t} \quad (4)$$

where P is the pressure, r is the radius of curvature and t is the thickness. Combining (3) and (4):

$$P = \frac{2t\sigma_L}{r} \left[\frac{R'_L + R'_T}{R'_T(1 + R'_L)} \right]^{-\frac{1}{2}} \quad (5)$$

where each of the variables is a function of strain. The experimentally determined values of R'_L , R'_T (Fig 67) and flow stress, σ_L (Fig 4) of the 3.3 mm thick sheet were used to evaluate equation (5) for various stages of forming of the hemisphere. The calculation was based on a hemisphere diameter of 200 mm and the radius of curvature, r , was calculated for each strain increment by assuming that the bulge surface formed part of a sphere. The polar thickness at each stage of forming was estimated by interpolating the results of Cornfield and Johnson (162) who calculated bulge thickness profiles for various strain rate sensitivities. An m value of 0.67 (from Fig 7) was used.

The strain hardening, based on equivalent strain at the pole ϵ_e was calculated for each thickness strain increment as follows:

The work done for a small strain increment is given by the sum of the work done in each of the principal directions ie

$$\sigma_e d\epsilon_e = \sigma_1 d\epsilon_1 + \sigma_2 d\epsilon_2 + \sigma_3 d\epsilon_3 = 2\sigma_1 d\epsilon_1 = -\sigma_1 d\epsilon_3$$

Substituting in equation (3) and assuming proportional straining (106):

$$\epsilon_e = - \left[\frac{R'_L(1 + R'_T)}{R'_L + R'_T} \right]^{\frac{1}{2}} \epsilon_3 \quad (6)$$

where ϵ_3 is the thickness strain.

The strain increments were related to forming time by imposing a condition of constant strain rate of $3.1 \times 10^{-4} \text{ s}^{-1}$ at the pole. This strain rate corresponded to maximum strain rate sensitivity during uniaxial testing.

Fig 68 shows four pressure cycles calculated for the forming of a hemisphere. For each curve different material properties were assumed as follows:

- (1) Uniform thinning was assumed, a constant flow stress of 9 MPa and an R value of 1.0 (isotropic deformation) were used. This type of calculation is probably typical of many used commercially.
- (2) A strain rate sensitivity of $m = 0.67$ was imposed. A constant flow stress of 9 MPa and an R value of 1.0 were assumed.
- (3) A strain rate sensitivity of $m = 0.67$ was used and strain hardening was included. The R value used was 1.0.
- (4) As in (3) except that the R values R'_L and R'_T were assumed to vary with strain. This calculation makes full use of the uniaxial experimental results.

Compared with the uniform thinning case the effect of including the strain rate sensitivity was to extend the forming time and reduce the maximum pressure (Fig 68), curve 2). Introducing strain hardening caused a rise in the maximum pressure and displaced the peak to a longer time (Fig 68, curve 3). Inclusion of the experimentally determined R values in the calculation caused a reduction in pressure (Fig 68, curve (4)). The choice of a fixed flow stress of 9 MPa for two of the curves resulted in an overestimation of pressure for the first half of the cycle and an underestimation for the second half when compared with the curves incorporating strain hardening. It is significant that for times up to 20 minutes the greatest discrepancy in pressure (up to 40%) was between curve 1 for uniform thinning and curve 4 incorporating all the measured variables.

The influence of strain rate sensitivity on the pressure cycle for the forming of the hemisphere was further investigated by assuming m values of 0.5 and 1.0 in the calculation including strain hardening and anisotropy. The results are given in Fig 69. All the calculated pressure cycles lie close to the curve drawn through the points for $m = 0.67$. The only significant difference between the pressure cycles is

that forming time increases with decreasing m value. The optimum forming cycles for the range of m values typical of Ti-6Al-4V are, therefore, approximately the same despite the fact that the final thickness profiles, as calculated by Cornfield and Johnson (162) vary significantly. In practice the strain rate sensitivity of the sheet may be more significant factor in minimising thinning at the pole than adherence to the optimum pressure cycle, although the two factors are related by the sensitivity of m to strain rate variations.

The other set of points included in Fig 69 are those calculated assuming an average R value $\bar{R}' = (R'_L + R'_T)/2$. These points also lie close to the curve and indicate that a single R value (varying with strain) can be used to describe the material. Traditionally R is also measured at 45° to the rolling direction and a mean R value defined (154) :

$$\bar{R} = (R_0 + R_{45} + R_{90})/4.$$

A simpler definition suffices for R values after superplastic strain because the normal anisotropy does not vary substantially between different directions in the plane of the sheet.

The pressure calculations have been based on two main assumptions:

(1) The in-plate stresses σ_1 and σ_2 are equal. In practice, these two stresses would be slightly different as a result of the planar anisotropy in the sheet, as discussed in section 5.3. This means that the bulge profile would not be symmetrical about the pole. However, Fig 69 shows that using the average R value $\bar{R}'$, rather than R'_L and R'_T in the calculation does not significantly affect the pressure cycle. Thus, the effects of planar anisotropy are minimal.

(2) Strain rate sensitivity does not depend on strain. Experiment shows (Fig 10) that this is approximately true for strains up to 1.0 and Fig 69 indicates that varying m between 0.5 and 1.0 does not substantially affect the calculated pressures.

In view of the assumptions that have been made the calculated pressure cycles can only be considered as approximations and strictly only apply to the material from which the uniaxial data was derived at 925°C . Nevertheless they serve to illustrate the underlying principles and demonstrate the influence of the various material parameters.

5.5 Analysis of necking of Ti-6Al-4V during superplastic deformation

The computer analysis of neck development has allowed the material

properties inside and outside the neck to be estimated separately for each stage of deformation. Such results could not be achieved directly by experiment.

It was noticed that the neck strain rate decreased slightly during the first few percent of deformation (Fig 61). This was the case for each of the neck sizes examined and occurred because, in the early stages of deformation, the strain in the neck was similar to that in the uniform region and a relatively high neck strain rate was required to balance the load. Further deformation led to greater strain hardening in the neck compared with the uniform region and a small reduction in neck strain rate occurred.

The two terms on the right hand side of equation (1) represent the contributions of strain and strain rate changes to the overall increase in flow stress with strain. Thus, Fig 64 illustrates the relative importance of strain rate sensitivity and strain hardening in resisting neck development. For the first part of the deformation, when neck strain rate changes were minimal (Fig 61), the strain hardening term is largest, but beyond about 85% uniform strain the strain rate sensitivity is dominant. However, strain hardening also occurs in the uniform region of the gauge length and for much of the deformation this almost exactly cancels out the strain hardening in the neck (Fig 64). The uniform strain at failure or "limiting strain" is therefore strongly influenced by the strain rate sensitivity and particularly by the residual m value after high strains. The initial m value is likely to be of little consequence by comparison. The practical significance of this for superplastic forming is that an alloy, for which grain coarsening is relatively slow, may well demonstrate more superplastic ductility than another that exhibits higher m values at low strains, but is less resistant to grain coarsening. Similarly, a reduction in forming temperature may reduce the initial strain rate sensitivity, but does not necessarily reduce the limiting strain. Other authors (167-172) have remarked on the strong influence of rate sensitivity in resisting neck development particularly in the "post-uniform" stage. The results presented here apply only to superplastic Ti-6Al-4V with a relatively large pre-existing defect in the gauge length, but clarify the significance of m and n .

Ghosh (172) carried out a numerical analysis of neck development incorporating strain and strain hardening based on the relation $\sigma = K\epsilon^n \dot{\epsilon}^m$. The results of this work are not directly comparable with those presented here because Ghosh only investigated n values as high as 0.2. However, discrepancies would be expected because Ghosh assumed constant values of m and n in his analyses which would clearly be a gross approximation for superplastic Ti-6Al-4V, as shown by Fig 63. Arieli and Mukherjee (173) suggest that total superplastic elongations are more closely related to the terminal strain rate sensitivity than to the initial value. This is confirmed by the present work as discussed above.

The point of instability during tensile deformation has been the subject of some discussion (167-172,174-176). According to Hart (167) instability is reached when $\gamma + m = 1$, where γ is a strain hardening parameter $\frac{1}{\epsilon} \left(\frac{\partial \ln \sigma}{\partial \ln \epsilon} \right)$. The criterion for instability in this case is $(\delta \dot{A} / \delta A)_p > 0$. Where δA is the variation in cross section area between the necked and the uniform regions and the condition of constant load p simply reflects that each section of the gauge length bears the same load at any one time. Duncombe (168,169) chose the point at which the variation in strain rate $\delta(\dot{A}/A)$ increased as the point of instability and Ghosh (172) chose the point of maximum load, which for constant strain rate testing, is equivalent to the Considere criterion ($\gamma = 1$). Each of these definitions can be applied to superplastic deformation as follows:

Hart (167) discussed the point of instability by considering small variations in strain and strain rate along the gauge length associated with small inhomogeneities or incipient necks. However, it was assumed that, at any instant, the values of γ and m did not vary along the gauge length. Therefore, as a neck develops the stability criterion $\delta \dot{A} / \delta A = 0$ will be represented less and less accurately by the expression $\gamma + m = 1$, whether or not the deformation is considered stable. For superplastic materials, where a long period of neck development precedes failure or where a large defect exists prior to deformation, this inaccuracy is likely to be significant. The value of $\delta \dot{A} / \delta A$ is shown in Fig 70 as a function of strain along with the value of $\gamma + m$ (for the uniform region). The indicated points of instability occur at significantly different amounts of strain (50% and 150% respectively). Furthermore, the value of $\gamma + m$ decreases monotonically

towards instability whereas the curve for $\dot{\delta A}/\delta A$ indicates a region of increasing stability up to 15% uniform strain. The sign of δA will not normally change during deformation and the $\dot{\delta A}/\delta A = 0$ criterion can be more simply expressed as the point at which the rate of change of area is the same in the uniform region and in the neck. Thus, Hart's criterion can be applied provided that the general form $\dot{\delta A}/\delta A = 0$ is used rather than the derived form $\gamma + m = 1$.

The maximum load occurs when $\gamma = 1$. It also follows, from the definition of n that $\epsilon = n$ at this point. Examination of Figs 61 and 63 shows that at maximum load $\epsilon \approx n \approx 0.52$. The value of n is often assumed to be constant during room temperature testing, but this is not necessary in order that the equality above is satisfied at maximum load. However, it is important to note that the condition of constant strain rate in the uniform gauge section implies that maximum load would occur at the same uniform strain regardless of the size of the pre-existing defect.

Applying the definition of instability $\delta(\dot{A}/A) > 0$ implies that deformation is unstable from the outset in the present case. However, neck strain rate decreased during the early stages of deformation. Therefore, choosing the point at which strain rate variation ($\dot{\delta A}/A$) increases (about 10% uniform strain) seems more sensible in this case. Hutchinson and Obrecht (174) used the area variation δA to monitor neck development in rate sensitive materials. For the case presented here δA decreased with strain and reached a minimum at about 73% uniform strain. Thus, increasing δA could also be used to define instability.

Therefore, the definition of tensile instability is very much open to interpretation and for the case discussed here a wide range of values of instability strain are predicted, as summarised in Fig 70. None of the definitions can be applied without detailed knowledge of the neck development except for the maximum load criterion, which takes no account of the defect size when constant strain rate in the uniform region is assumed.

The imaginary specimen considered here is of infinite length with the neck developing in a pre-existing defect. Real specimens may contain such a defect, but usually necking takes place at the centre of the gauge length where the material is least constrained. Tensile failure is also likely to be influenced by cavitation, even for the superplastic

deformation of titanium alloys, and localised rather than diffuse necking will probably occur at some stage depending on the specimen shape and isotropy. No account has been taken of stress components on the neck other than the axial component ie the neck radius is considered to be infinite. It is also assumed that a constant strain rate is maintained in the uniform region of the specimen. However, in practice, if the average strain rate is maintained constant over the gauge length such that $\dot{\epsilon} = V/l$, where V is the extension rate and l is the instantaneous gauge length, then the strain rate in the uniform region will reduce as the neck develops.

A further assumption is that the flow stress of the material can be described by the curves in Fig 4 regardless of its strain history. This is roughly equivalent to assuming that the effects of static annealing on grain size and hence on flow stress are negligible compared with the effects of strain. Other work (138) shows that this assumption is reasonable, particularly as no strain rates lower than $3 \times 10^{-4} \text{ s}^{-1}$ are considered.

6 CONCLUSIONS

- 1 Superplastic deformation of the alpha/beta titanium alloys Ti-6Al-4V, IMI 550 and Ti-8Al-1Mo-1V occurred largely by grain boundary sliding. This produced random grain rotations leading to reductions in alpha and beta phase texture intensities. Observations of grain emergence from test piece surfaces were made.
- 2 The highest m values recorded for the Ti-6Al-4V and Ti-8Al-1Mo-1V alloys were at temperatures corresponding to equal phase proportions. At higher temperatures the m value of the Ti-6Al-4V was sharply reduced as a result of rapid grain coarsening.
- 3 The sheet test pieces machined from Ti-6Al-4V bar exhibited lower strain rate sensitivity and higher flow stress in areas of microstructure containing contiguous, aligned alpha grains than in non-aligned regions. Therefore, the R values were influenced by directionality in the microstructure and hence by test piece orientation. The sense of the superplastic strain anisotropy for each orientation was consistent with the measured order of flow stress at 875-975°C; $\sigma_L > \sigma_T > \sigma_S$.
- 4 The R values of the rolled Ti-6Al-4V sheets were influenced by directionality in the microstructure. The lowest R values were measured when the alpha phase aspect ratio on the section perpendicular to the tensile axis was highest. The R values of Ti-6Al-4V sheet increased with superplastic strain as the microstructure became more equiaxed.
- 5 The R values of the Ti-6Al-4V rolled sheet test pieces were also influenced by the effect of the test piece heads on the width contraction of the gauge length. This effect was reduced by increasing the length or reducing the width of the gauge section. A gauge length to width ratio of 1.6:1 was apparently sufficient for the measurement of R values without significant test piece end effects.
- 6 The only observed influence of crystallographic texture on R value was for the Ti-6Al-4V sheet test pieces machined from bar and occurred at the lowest test temperature (800°C) and at the highest strain rate ($1.5 \times 10^{-3} \text{ s}^{-1}$).
- 7 During superplastic deformation of the beta alloy Ti-15V-3Cr-3Al-3Sn subgrains developed and sliding on subgrain boundaries occurred. This process refined the grain size and led to an increase in m value with

strain. The superplastic flow stress of the Ti-15V-3Cr-3Al-3Sn was high and the m value low by comparison with the alpha/beta alloys. This was a result of the large initial grain size. The texture intensity of Ti-15V-3Cr-3Al-3Sn was not reduced by superplastic strain.

8 Annealing at the forming temperature caused recrystallisation and a reduction in room temperature strength in Ti-6Al-4V, IMI 550 and Ti-8Al-1Mo-1V. Superplastic strain generally caused a further slight loss of strength in these alloys. The room temperature 0.2% proof stress and tensile strength of IMI 550 after superplastic deformation was raised by increasing the cooling rate from the forming temperature and by ageing. Room temperature strengths up to 18% greater than those of as-formed Ti-6Al-4V were achieved in IMI 550 by heat treatment after forming. Superplastic deformation reduced the room temperature strength of Ti-15V-3Cr-3Al-3Sn.

9 The calculated pressure cycles for the forming of a hemisphere with a constant strain rate at the pole were significantly altered by the inclusion of uniaxial experimental strain hardening and R value data. Variation of strain rate sensitivity in the range 0.5-1.0 produced only minimal changes in the optimum pressure cycle.

10 The analysis of necking in superplastic Ti-6Al-4V showed that, for the case examined, strain hardening controlled the neck development during the early stages of deformation and strain rate hardening became the most significant factor at a later stage. The application of different definitions of instability produced widely differing values of uniform strain at the point of instability.

11 This thesis has paid particular attention to the causes of anisotropy of superplastic deformation in both sheet and bar material. The significance of microstructural directionality was established and it was shown that, for the alpha/beta alloys, crystallographic texture did not cause anisotropy under superplastic conditions.

12 The rolled sheet alloys Ti-6Al-4V, IMI 550 and Ti-8Al-1Mo-1V and Ti-15V-3Cr-3Al-3Sn were assessed for superplastic formability and it was shown that the alpha/beta alloys were superior to the beta alloys in this respect. The IMI 550 exhibited high strain rate sensitivity and relatively high room temperature strength after forming, especially after post-forming heat treatment. Therefore, this alloy appears to be particularly attractive for superplastic forming operations.

7 SUGGESTIONS FOR FURTHER WORK

- 1 More detailed study of the role of each of the phases in the superplastic deformation of two phase alloys is necessary. Light microscopy is restricted to examination of the room temperature structure after forming. Therefore, more in-situ observations of the deformation, such as those by Ma and Hammond (44) using the photoemission electron microscope (PEEM) are required.
- 2 A detailed model of superplastic deformation in two phase materials, leading to a new constitutive equation, is required. More accurate grain and phase boundary diffusion and high temperature shear modulus data would be needed in order to reliably predict the superplastic strain rates of alpha/beta titanium alloys.
- 3 The IMI 550 alloy appears to be an attractive candidate for sheet forming operations. However, more post-forming room temperature mechanical property data, such as fatigue resistance and fracture toughness, is required to confirm this.
- 4 It would be desirable to carry out a series of biaxial forming trials of hemispheres and other shapes to further investigate the application of uniaxial flow stress, m and R value data to gas pressure forming. The purpose of this would be to develop a technique for predicting the optimum forming pressure cycle for any shape.
- 5 The analysis of neck development has concentrated on the case of constant strain rate in the uniform section and a neck of infinite radius. In order to make the results more representative of real specimens it would be desirable to allow the uniform strain rate to decrease such that the average strain rate rather than the uniform strain rate was maintained constant as discussed above. Another important case to investigate is that of uniform extension rate, in which case an imaginary gauge length would need to be specified. To incorporate the effect of a finite neck radius the specimen length would have to be divided into segments and the equivalent stress and strain calculated for each segment for each strain increment as in the approach adopted by Ghosh (171). However, superplastic failures do not normally show substantial localised necking and this elaborate method may not be necessary.

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Table 1

EFFECT OF SUPERPLASTIC STRAIN AT 925°C AND $3 \times 10^{-4} \text{s}^{-1}$ ON THE TENSILE PROPERTIES OF Ti-6Al-4V SHEET. INITIAL SHEET THICKNESS 3.3mm

Orien- tation	SP Strain %	0.1PS MPa	0.2PS MPa	0.5PS MPa	TS MPa	E GPa	E_u % on 20mm	E_t % on 24mm
700°C 2 hrs, furnace cooled (as-received)								
T	-	1053	1051	1043	1090	124	8.8	16.3
L	-	938	933	935	1003	106	-	17.9
Annealed at the forming temp 925°C $\frac{1}{2}$ hr, cooled 25°C min ⁻¹								
T	-	945	935	928	1025	127	10.8	20.0
L	-	825	816	818	914	107	9.8	20.0
Superplastically formed, cooled 25°C min ⁻¹							E_u % on 10mm	E_t % on 14mm
T	32	917	917	921	1023	129	10.4	18.5
T	58	918	916	916	1003	131	11.0	-
T	128	919	919	922	987	121	8.5	11.5
L	18	847	845	847	938	105	9.2	-
L	153	807	809	814	910	95	10.3	20.0

PS $\equiv$ proof stress, TS $\equiv$ tensile strength, E $\equiv$ Young's modulus,
 E_u $\equiv$ uniform elongation, E_t $\equiv$ total elongation

Table 2

EFFECT OF HEAT TREATMENT ON THE TENSILE PROPERTIES OF IMI 550 SHEET

Heat Treatment	Orien- tation	0.1PS MPa	0.2PS MPa	0.5PS MPa	TS MPa	E GPa	E _u % on 20mm	E _t % on 25mm
700°C 2hr, air cooled (as received)	T	1034	1052	1060	1060	123	5.8	14.3
	L	989	997	1011	1079	113	4.2	13.5
900°C ½hr, cooled 25°C min ⁻¹	T	1025	1033	1034	1034	112	5.6	11.9
	L	953	948	948	1014	107	5.8	18.8
900°C ½hr, cooled 25°C min ⁻¹ aged 500°C 24hrs	T	1096	1104	1111	1121	121	5.6	12.4
	L	1033	1028	1033	1118	111	4.9	14.5
900°C ½hr, cooled 150°C min ⁻¹	T	1062	1088	1109	1125	118	4.3	13.2
	L	997	1013	1027	1116	108	5.0	15.0
900°C ½hr, cooled 150°C min ⁻¹ aged 500°C 24hrs	T	1129	1142	1155	1174	122	4.8	12.5
	L	1070	1072	1082	1175	114	4.9	13.6
900°C 3½hrs, cooled 25°C min ⁻¹	T	1002	1016	1025	1044	121	6.0	17.8
	L	922	922	928	1017	108	6.2	18.1
900°C 3½hrs, cooled 150°C min ⁻¹	T	1023	1055	1082	1117	119	5.4	15.9
	L	963	984	1000	1096	110	5.3	18.1

Each result is the average of two tests

Table 3

EFFECT OF SUPERPLASTIC STRAIN AT 900°C AND $3 \times 10^{-4} \text{s}^{-1}$ ON THE TENSILE PROPERTIES OF IMI 550 SHEET

Orien- tation	SP strain %	0.1PS MPa	0.2PS MPa	0.5PS MPa	TS MPa	E GPa	E_u % on 10mm	E_t % on 21mm
L	44	809	821	838	902	104	3.2	5.8
L	91	899	907	917	994	105	3.9	7.6
L	123	931	935	941	1027	104	5.2	8.7
L	190	832	841	848	926	98	5.3	7.5
T	37	961	959	959	1020	112	5.2	8.8
T	82	941	947	959	1001	111	3.5	4.8
T	120	926	921	922	978	105	4.6	7.8
T	169	960	965	972	1003	121	2.7	5.9
T	172	963	974	963	1019	113	6.0	7.5

Table 4

EFFECT OF POST-FORMING HEAT TREATMENT ON THE TENSILE PROPERTIES OF
IMI 550 SHEET (L ORIENTATION)

Post-forming heat treatment	SP strain %	0.1PS MPa	0.2PS MPa	0.5PS MPa	TS MPa	E GPa	E _u % on 10mm	E _t % on 12mm
Cooled 25°C min ⁻¹	190	832	841	848	926	98	5.3	7.5
Cooled 25°C min ⁻¹ aged 500°C 24 hrs	172	983	995	1010	1091	108	5.5	11.0
" " "	171	942	945	948	1024	108	5.4	9.3
900°C ½hr, cooled 150°C min ⁻¹	181	907	923	957	1035	98	3.0	10.8
900°C ½hr, cooled 150°C min ⁻¹ , aged 24hrs 500°C	183	1004	1020	1040	1127	108	5.4	9.5
" " "	191	997	1017	1037	1123	112	3.0	9.0

Table 5

EFFECT OF SUPERPLASTIC STRAIN AT 940, 970 AND 1010°C AND $3 \times 10^{-4} \text{ s}^{-1}$
ON THE TENSILE PROPERTIES OF Ti-8Al-1Mo-1V SHEET

Orien- tation	Forming/ Annealing temp °C	SP strain %	0.1PS MPa	0.2PS MPa	0.5PS MPa	TS MPa	E GPa	E_u % on 10mm	E_t % on 15mm
790°C 8hrs, furnace cooled, 790°C $\frac{1}{4}$ hr, air cooled (as-received)									
T	-	-	868	877	882	975	125	9.7	17.2
Annealed at the forming temp $\frac{1}{2}$ hr, cooled 25°C min ⁻¹									
T	940	-	831	836	838	939	121	10.2	18.2
T	970	-	826	826	826	931	127	10.7	19.1
T	1010	-	823	823	823	921	125	11.7	22.1
Superplastically formed, cooled 25°C min								E_u % on 20mm	E_t % on 25mm
T	940	187	821	842	863	877	126	5.2	15.4
T	940	267	784	808	829	908	119	6.2	12.0
L	940	221	787	806	850	913	116	8.2	15.8
T	970	210	773	778	789	844	120	8.0	11.8
T	1010	184	783	795	809	881	122	7.7	11.9

Table 6

EFFECT OF SUPERPLASTIC STRAIN AT 810, 860 AND 910°C AND $3 \times 10^{-4} \text{ s}^{-1}$
ON THE TENSILE PROPERTIES OF Ti-15V-3Cr-3Al-3Sn SHEET

Orien- tation	Forming temp °C	SP strain %	0.1PS MPa	0.2PS MPa	0.5PS MPa	TS MPa	E GPa	E _u % on 20mm	E _t % on 25mm
T*	as- received	-	787	787	785	802	89	-	23.3
L*	" "	-	754	762	762	779	81	-	25.9
T	810	340	690	694	696	724	76	2.0	10.3
L		348	628	628	632	648	74	2.5	6.6
T	860	316	657	660	663	698	77	3.4	9.7
L		275	639	639	639	667	72	4.7	10.5
T	910	310	636	638	641	674	74	2.2	6.4
L		272	631	643	648	681	71	4.0	7.7

* Average of two tests

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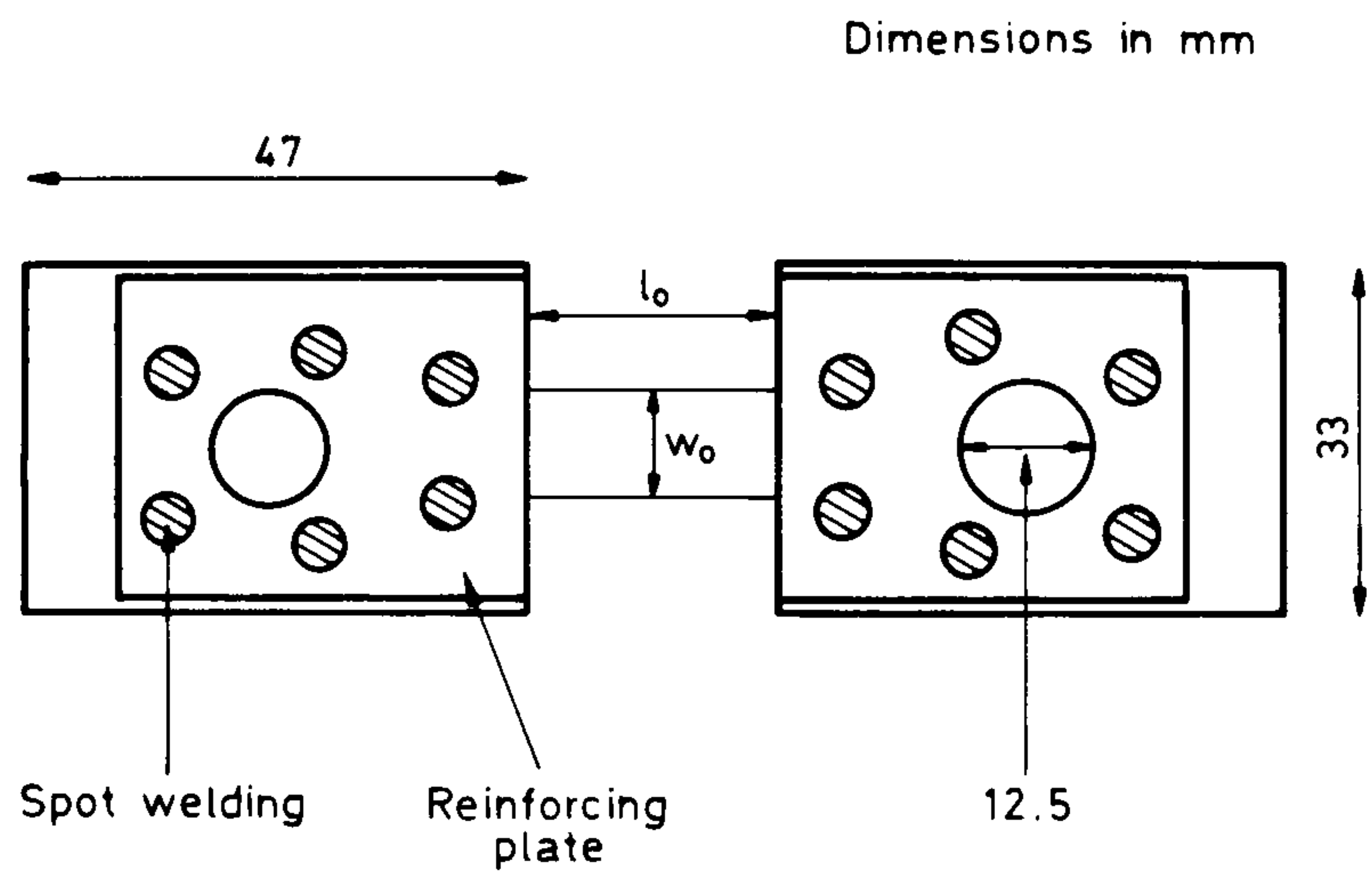
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		l_0 mm			
		2	5	10	25
w_0 mm		-	-	4	-
		-	-	8	-
		16	16	16	16
		-	-	19.5	-

Fig 1 Test piece dimensions

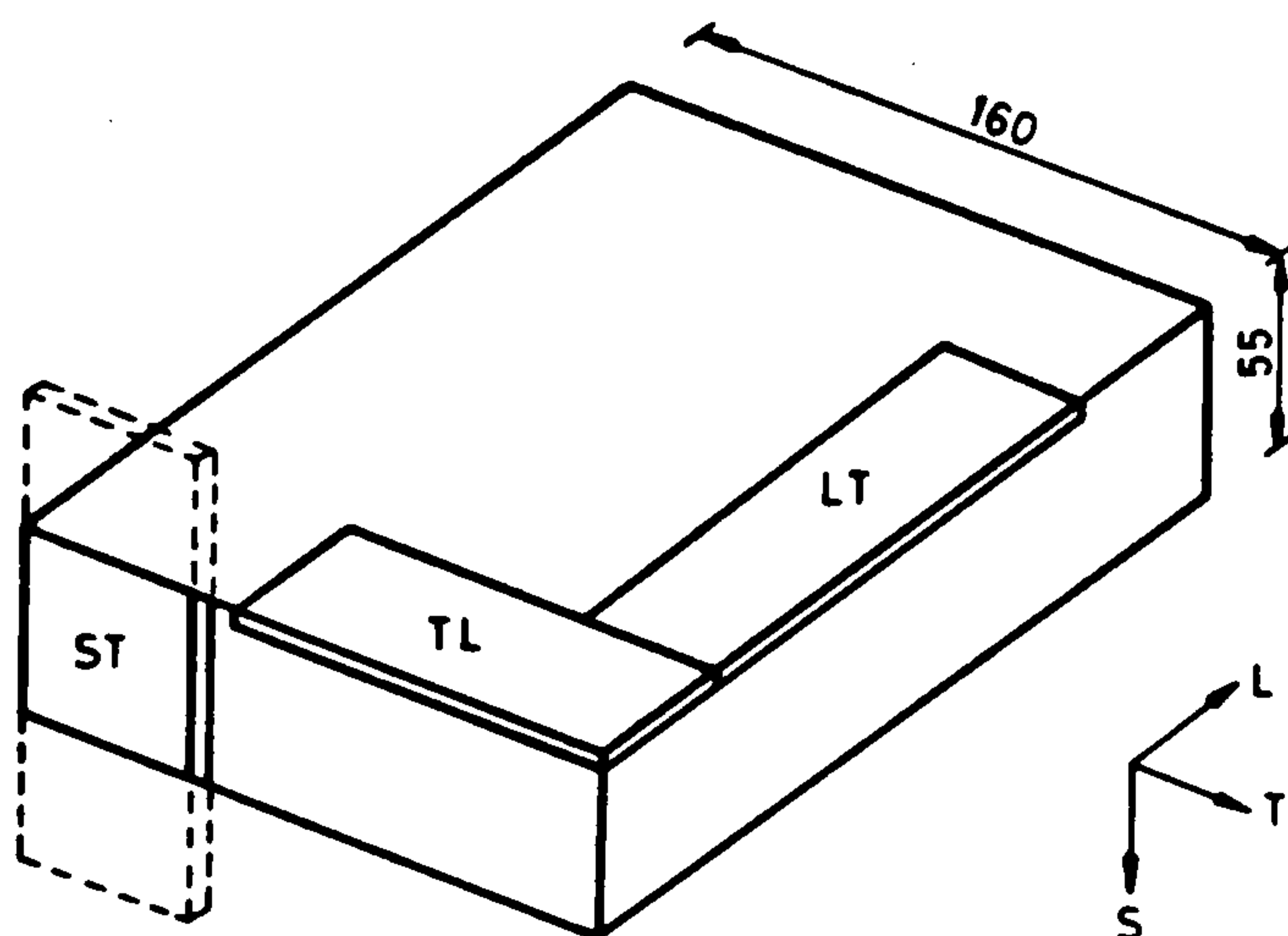
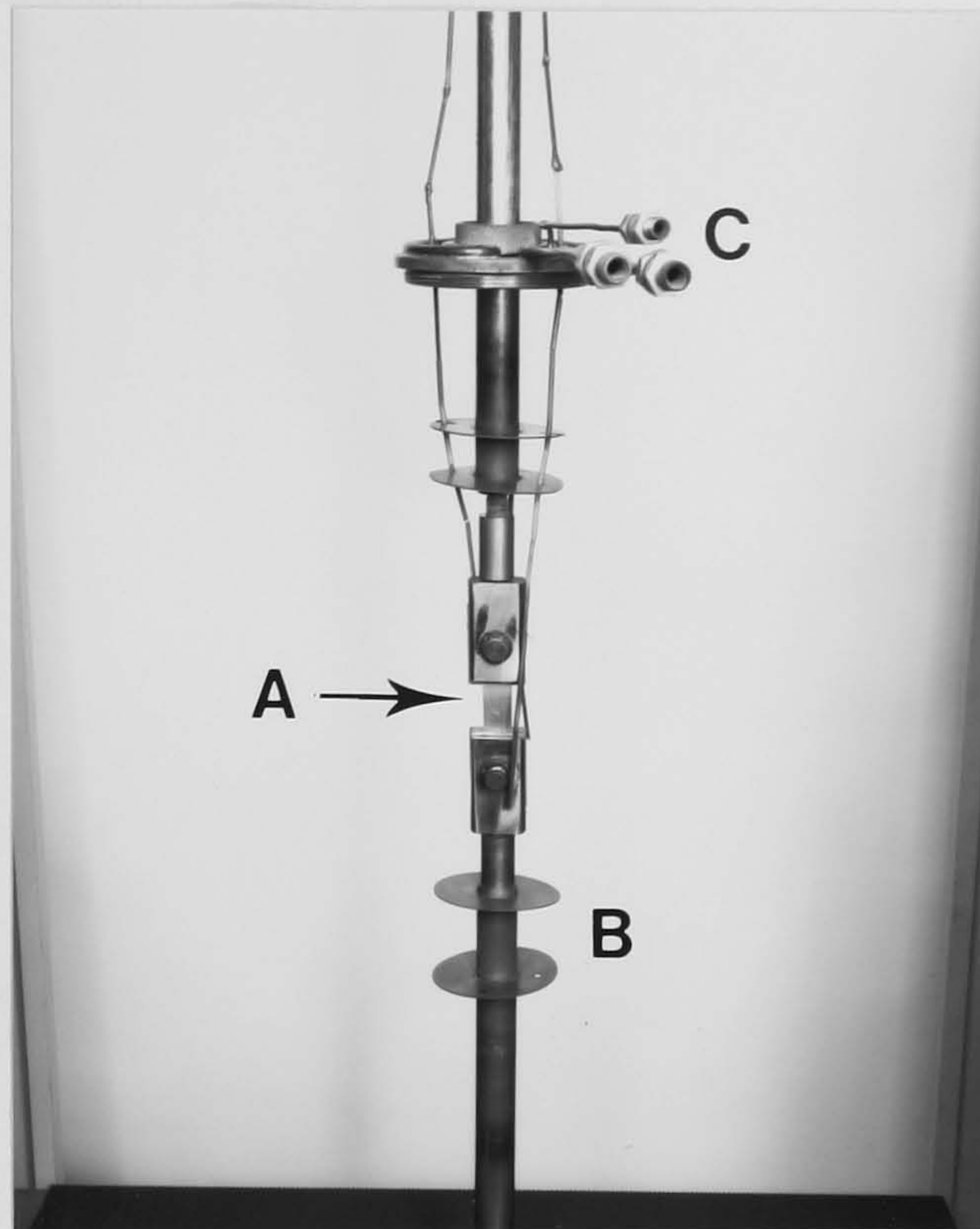
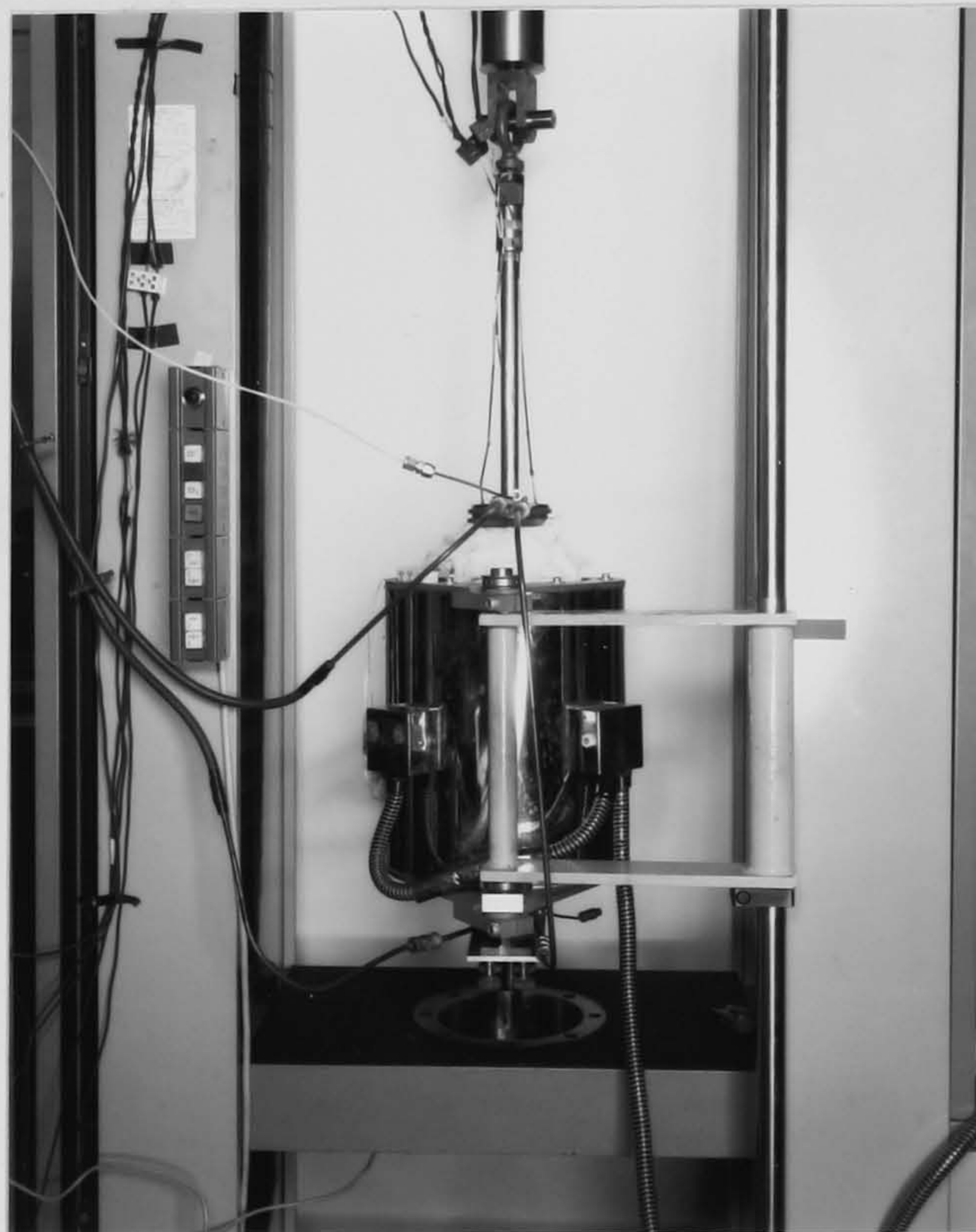


Fig 2 Cutting diagram for the edge textured Ti-6Al-4V bar



a) Quartz tube removed
showing:
A test piece
B heat shields
C water and Ar
connections



b) Cooling pipes
connected and
furnace closed

Fig 3 Apparatus for superplastic tensile testing

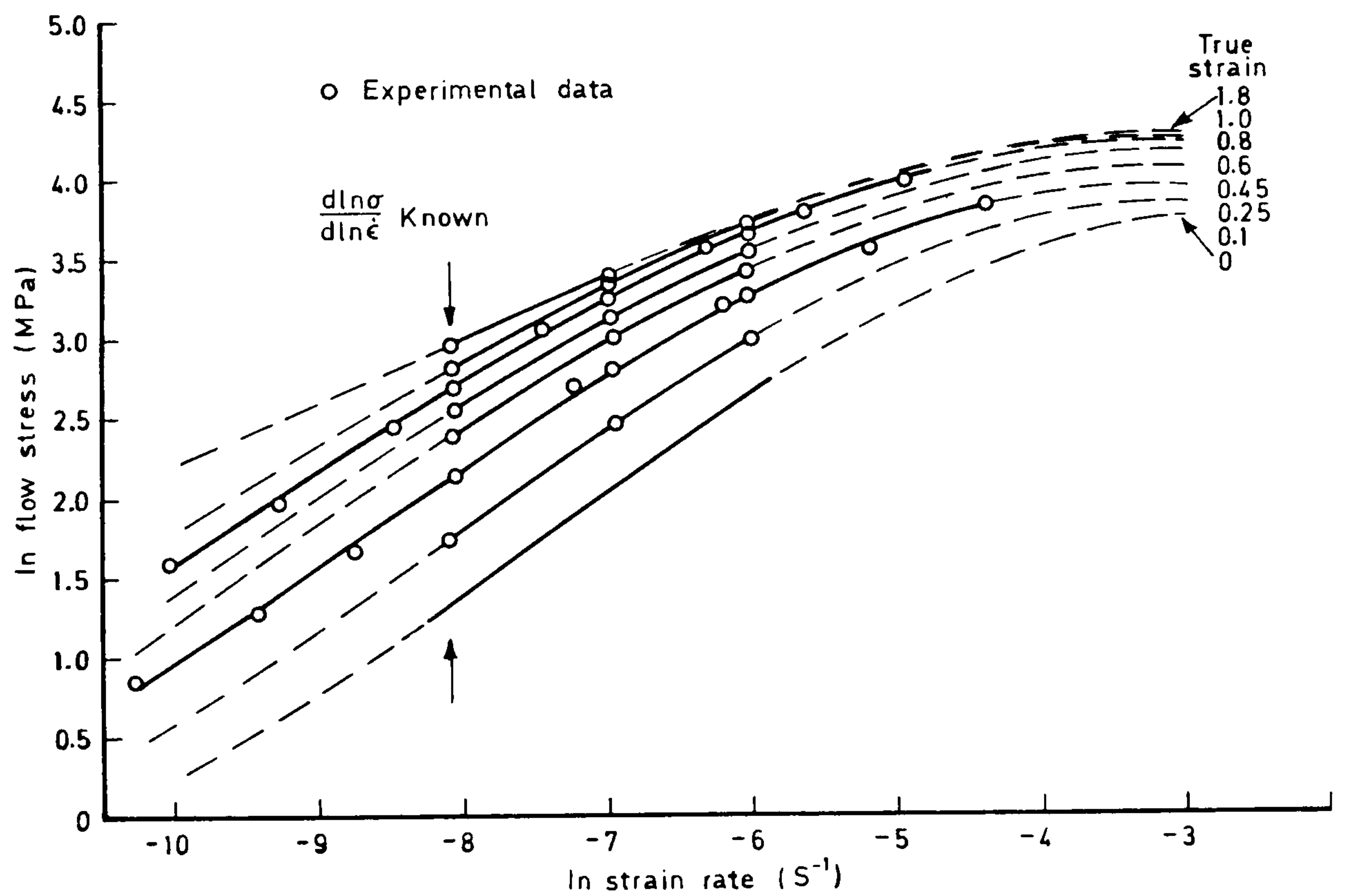


Fig 4 Flow stress data for 3.3mm Ti-6Al-4V sheet (pulled at 925 °C and $3 \times 10^{-4} s^{-1}$ in the L direction) used for the computer modelling of neck development

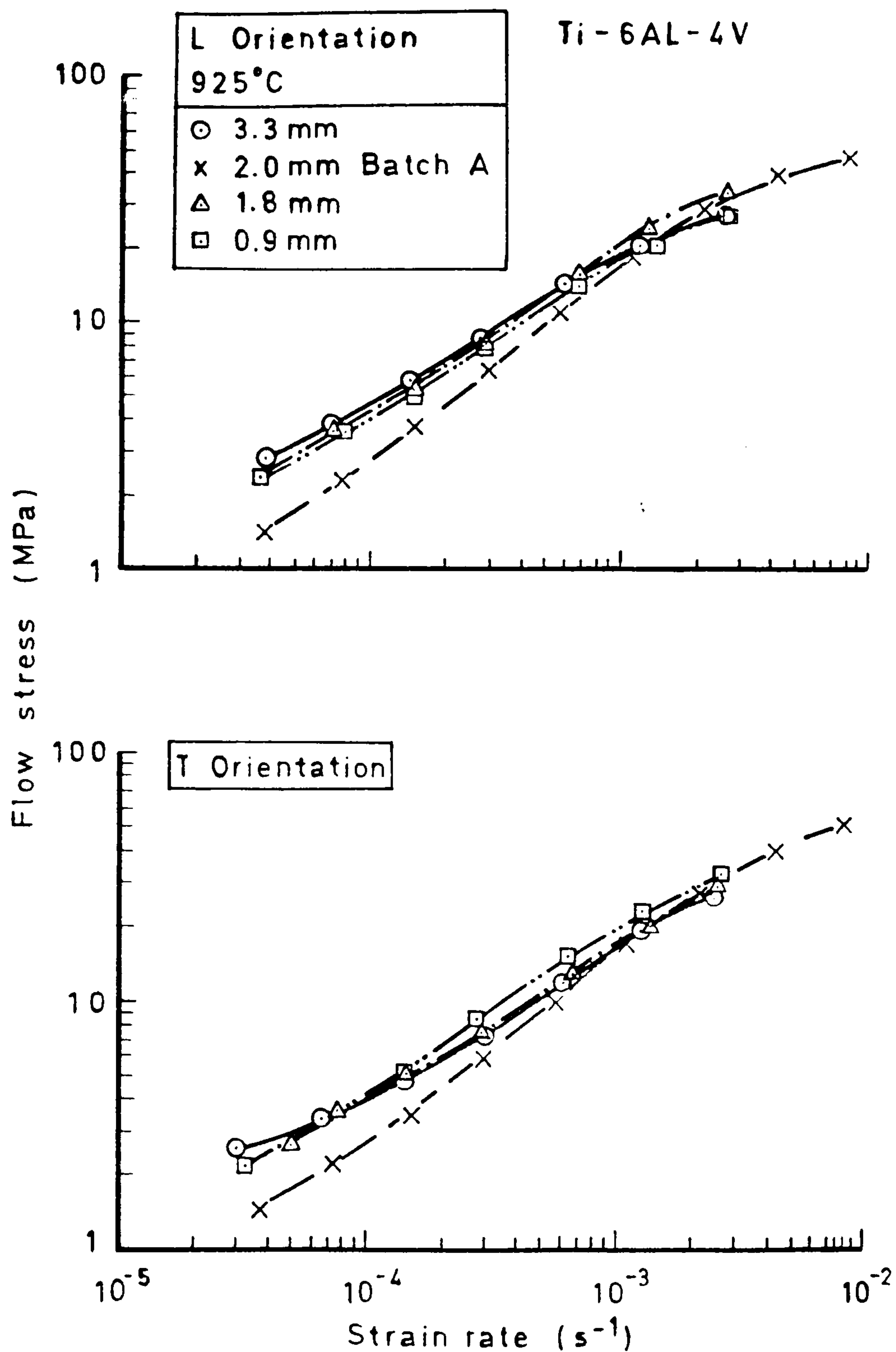


Fig 5 Effect of strain rate on flow stress of Ti-6Al-4V sheet of initial thickness 0.9-3.3mm pulled at 925 °C

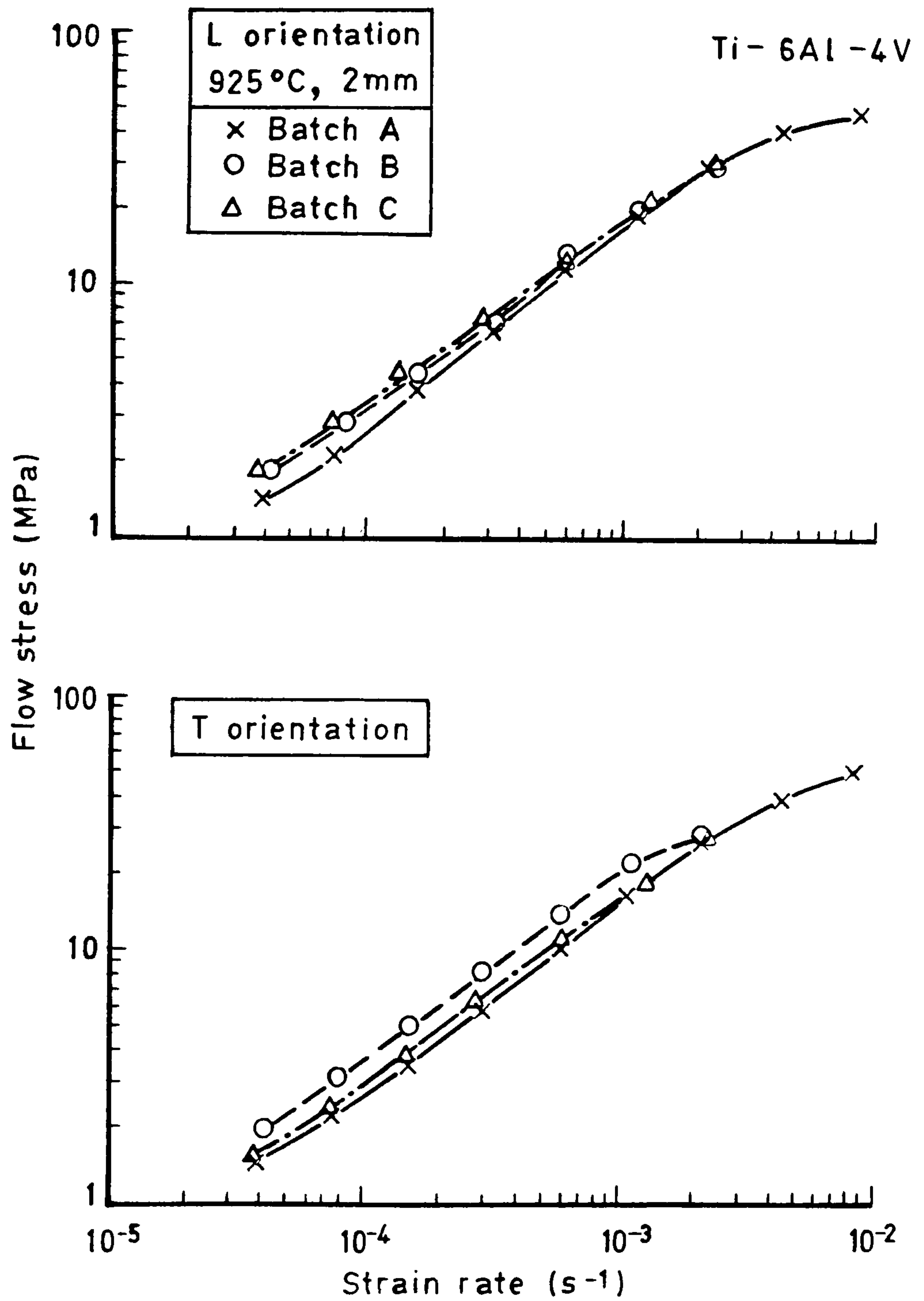


Fig 6 Effect of strain rate on flow stress of Ti-6Al-4V sheet of initial thickness 2.0mm pulled at 925 °C

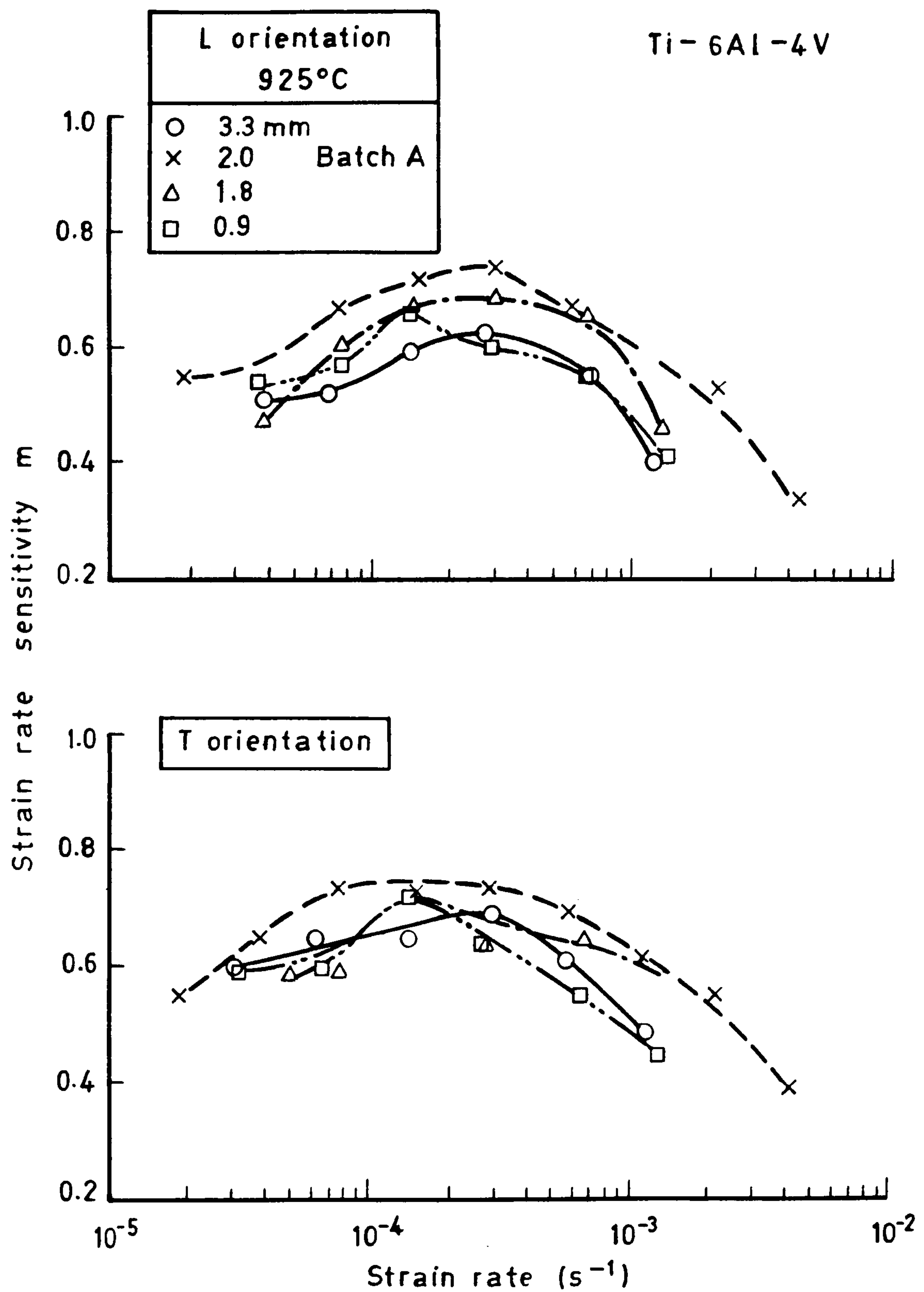


Fig 7 Effect of strain rate on m value of Ti-6Al-4V sheet of initial thickness 0.9-3.3mm pulled at 925 °C

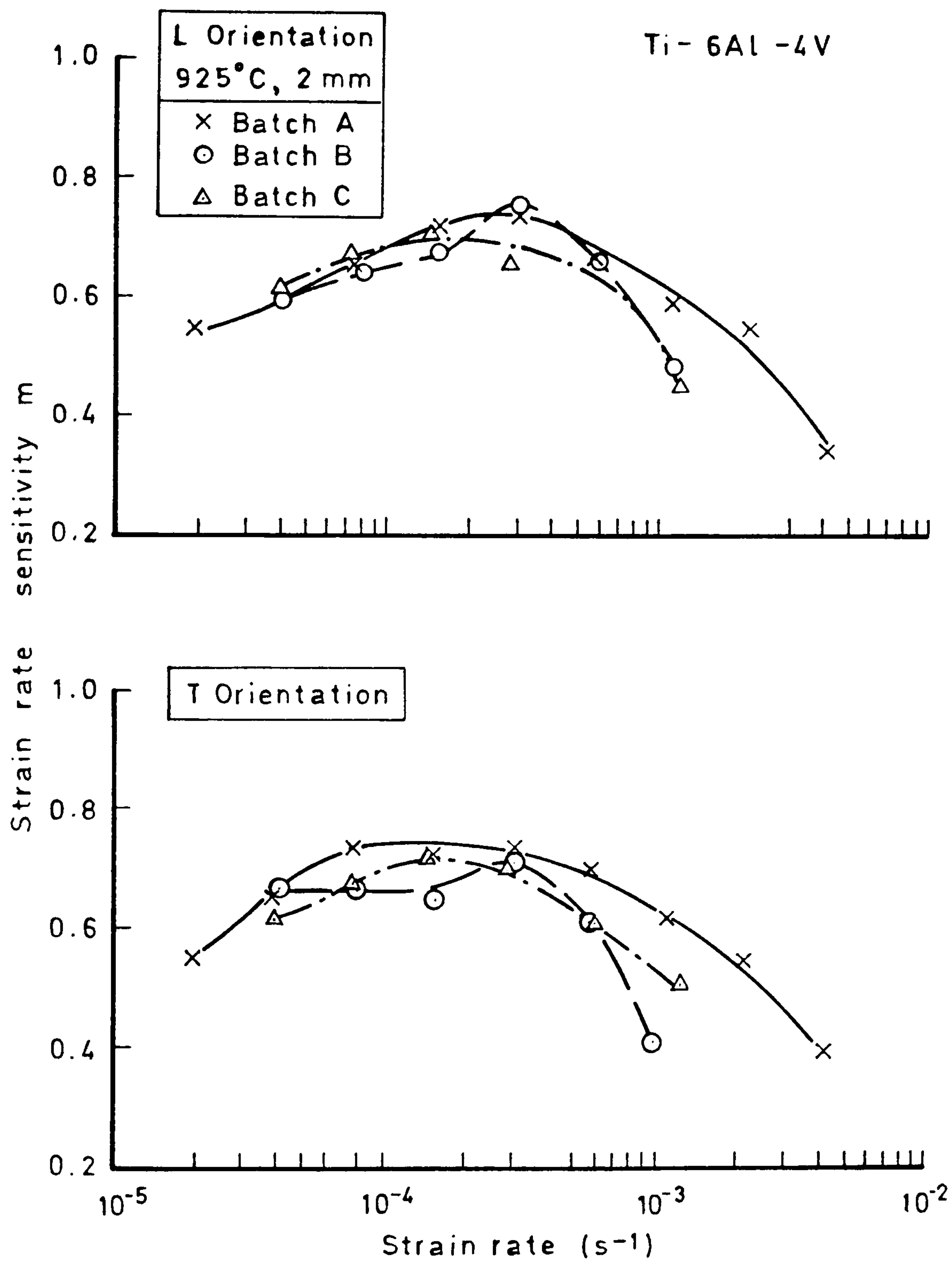


Fig 8 Effect of strain rate on m value of Ti-6Al-4V sheet of initial thickness 2.0mm pulled at 925 °C

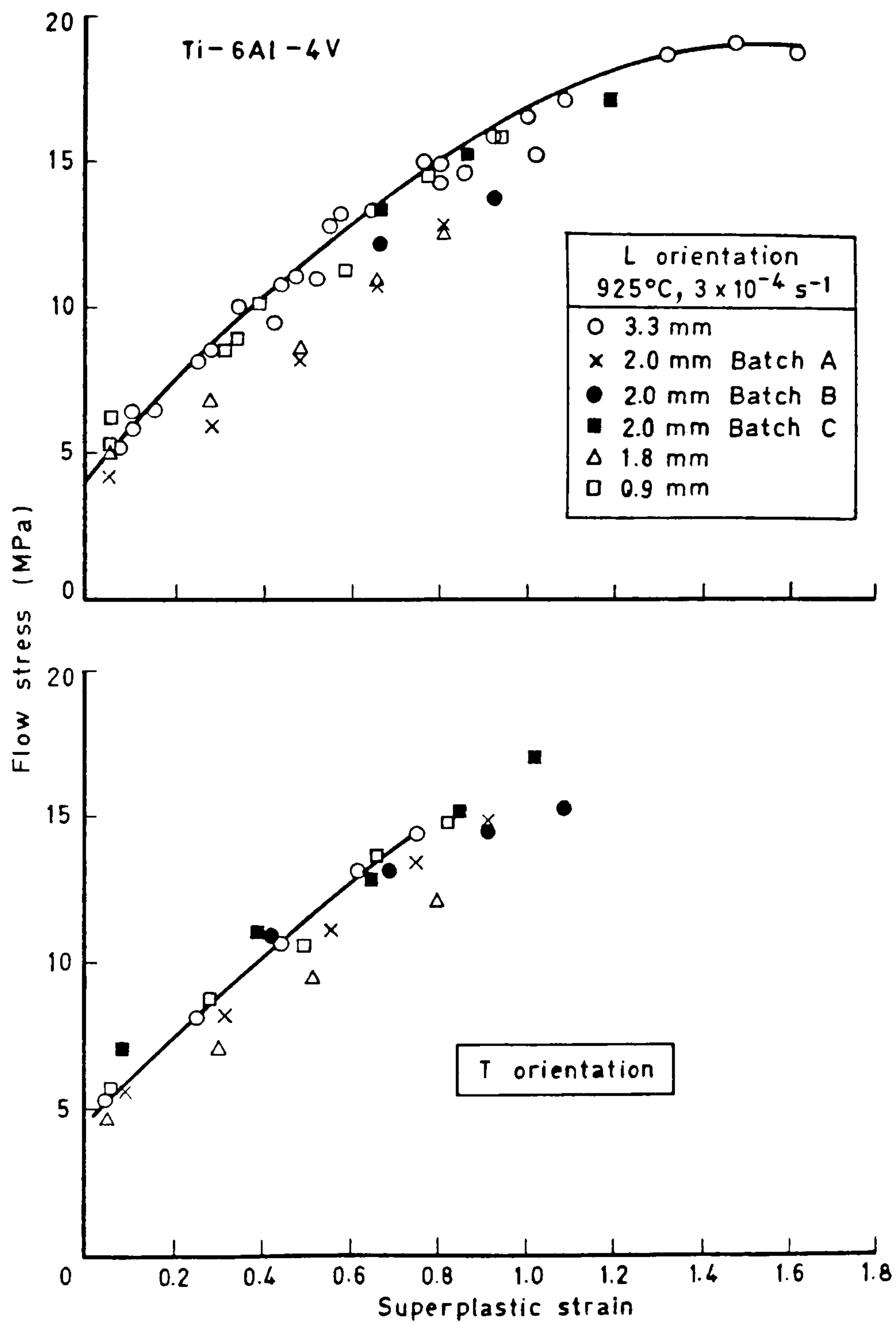


Fig 9 Effect of strain at 925°C and $3 \times 10^{-4} \text{s}^{-1}$ on flow stress of Ti-6Al-4V sheet

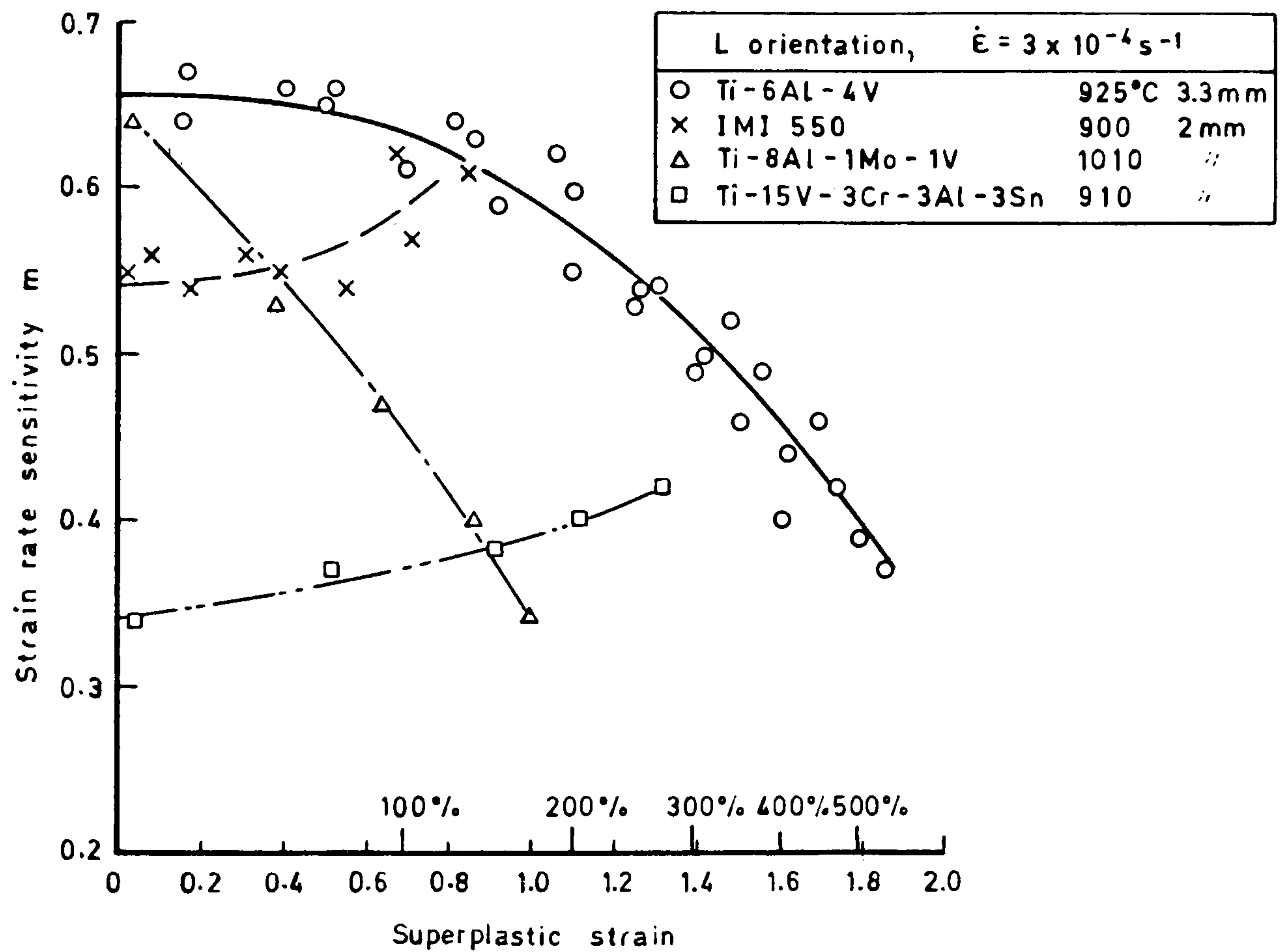


Fig 10 Effect of superplastic strain at $3 \times 10^{-4} \text{ s}^{-1}$
on m value of titanium sheet alloys

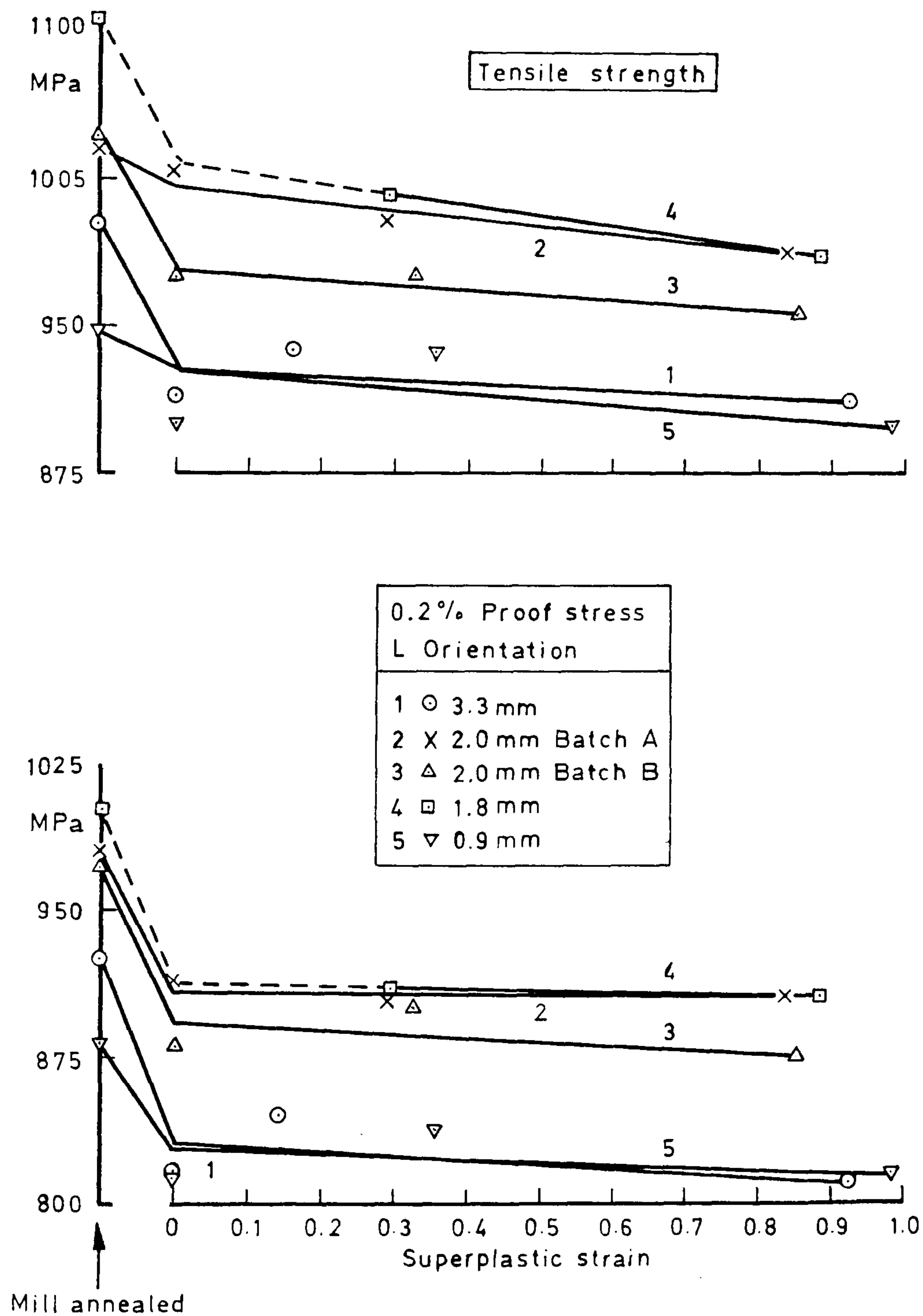


Fig 11 Effect of superplastic strain at 925°C and $3 \times 10^{-4}\text{s}^{-1}$ on room temperature strength of Ti-6Al-4V sheet (L orientation)

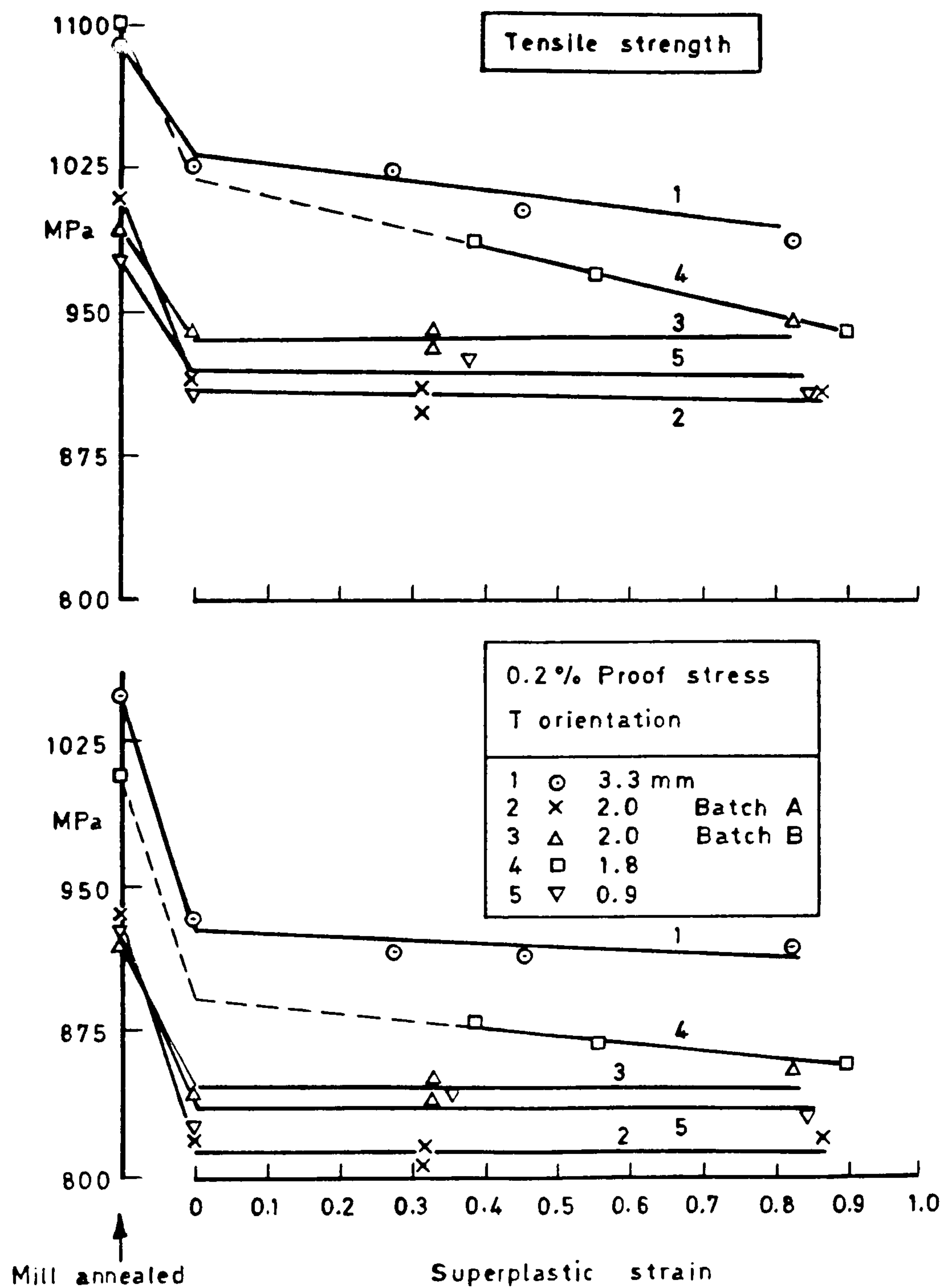
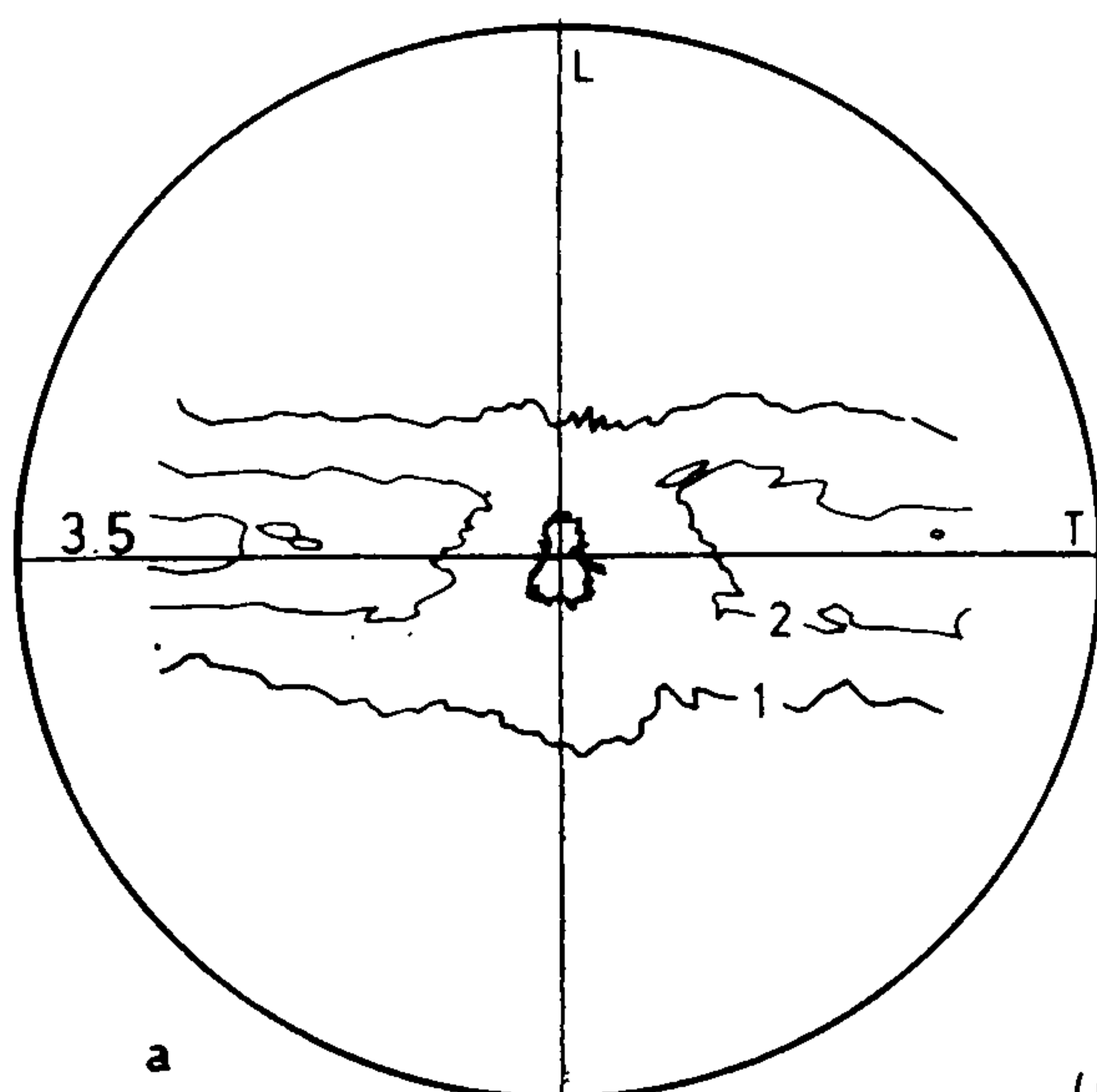


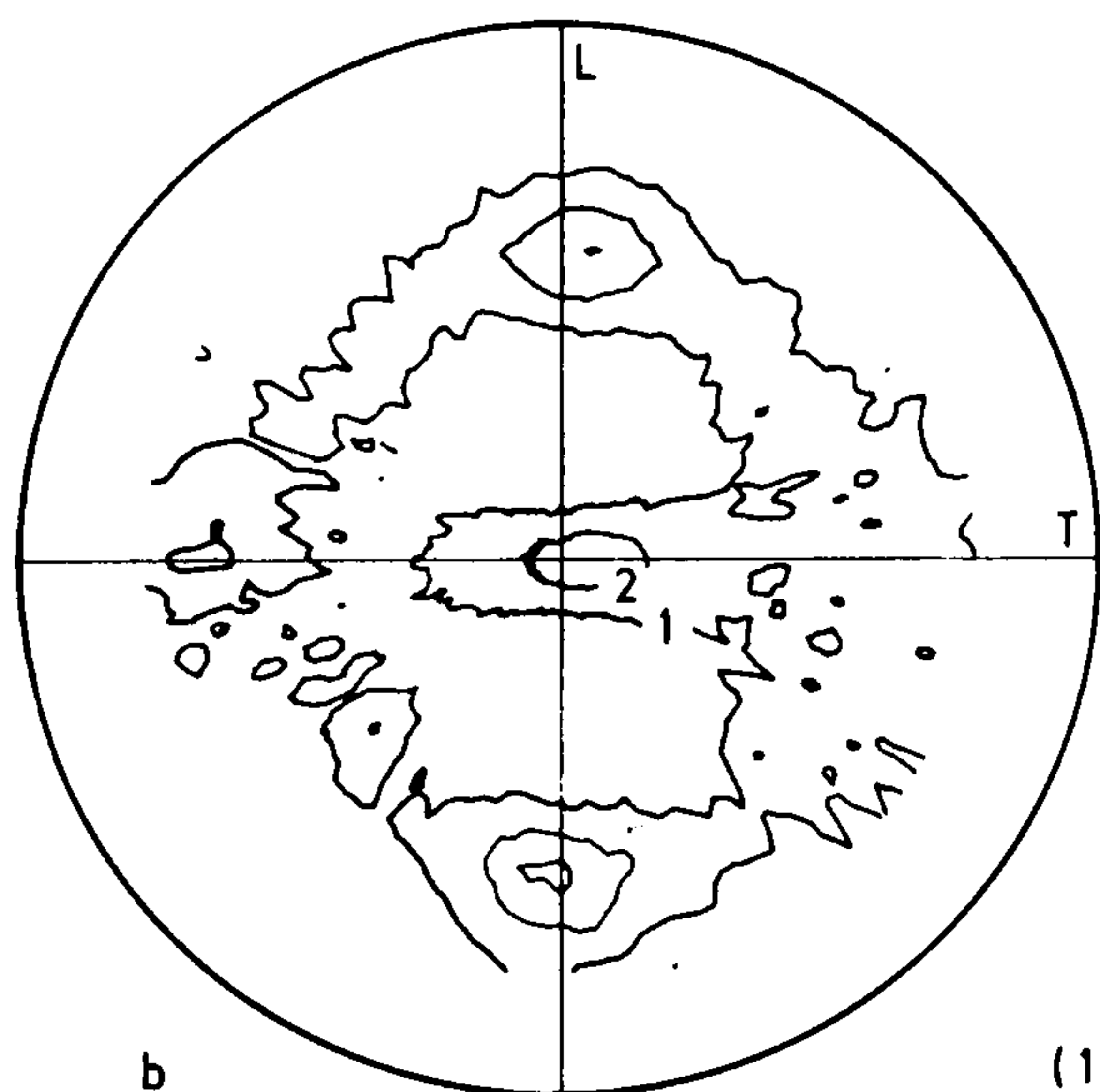
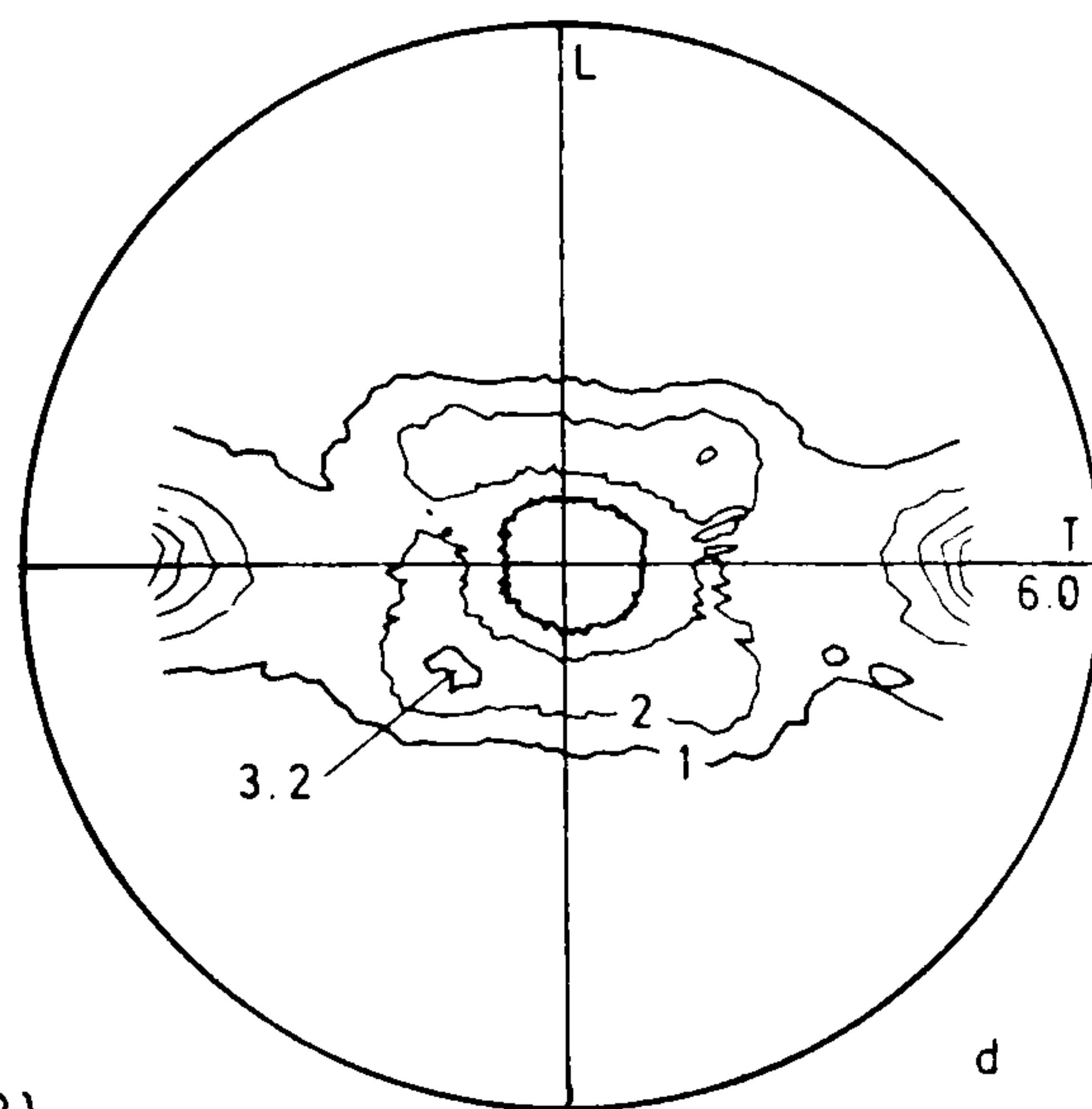
Fig 12 Effect of superplastic strain at 925 °C and $3 \times 10^{-4} \text{s}^{-1}$ on room temperature strength of Ti-6Al-4V sheet (T orientation)

Sheet surface
Contour intervals 1 x random

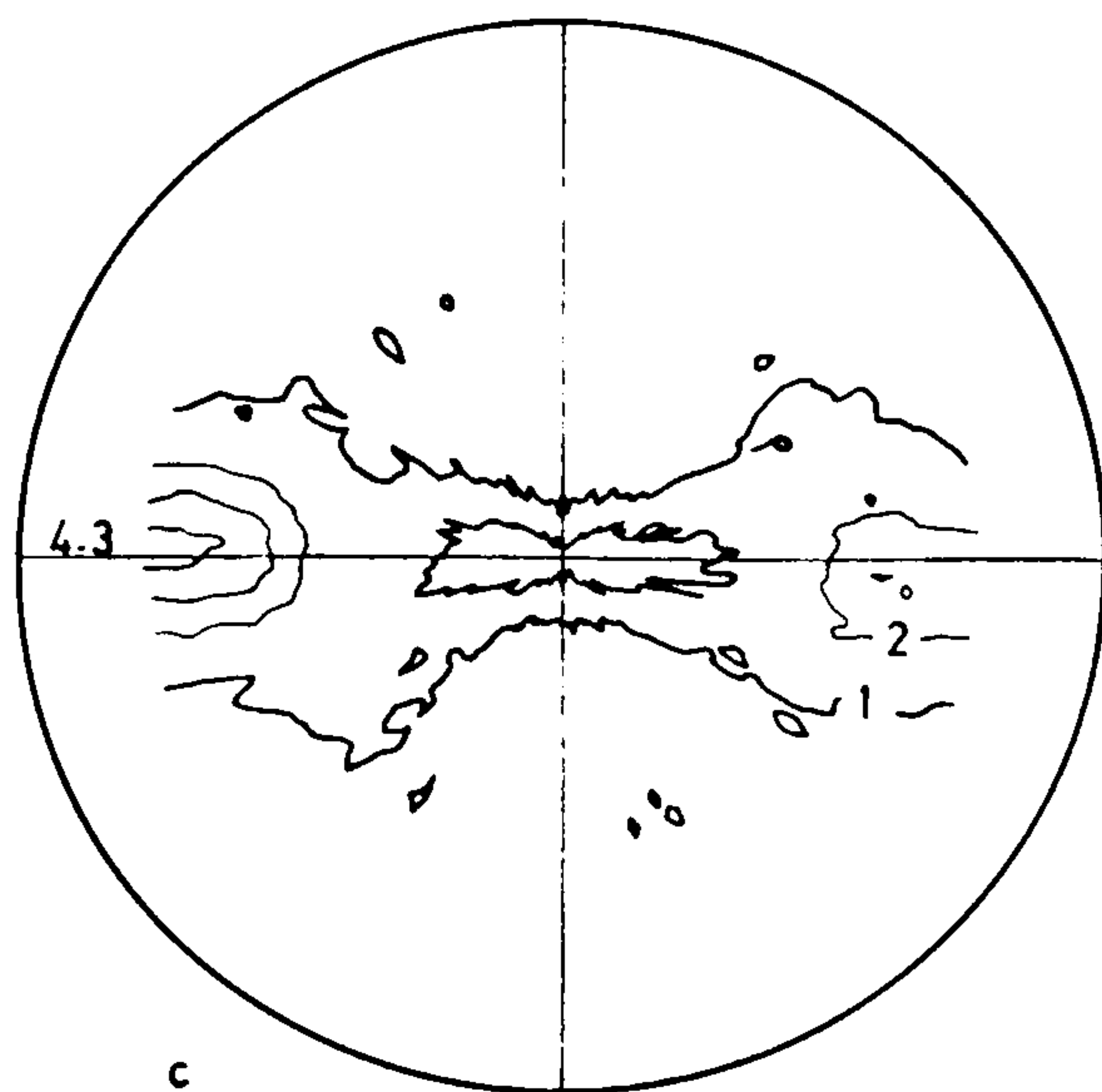
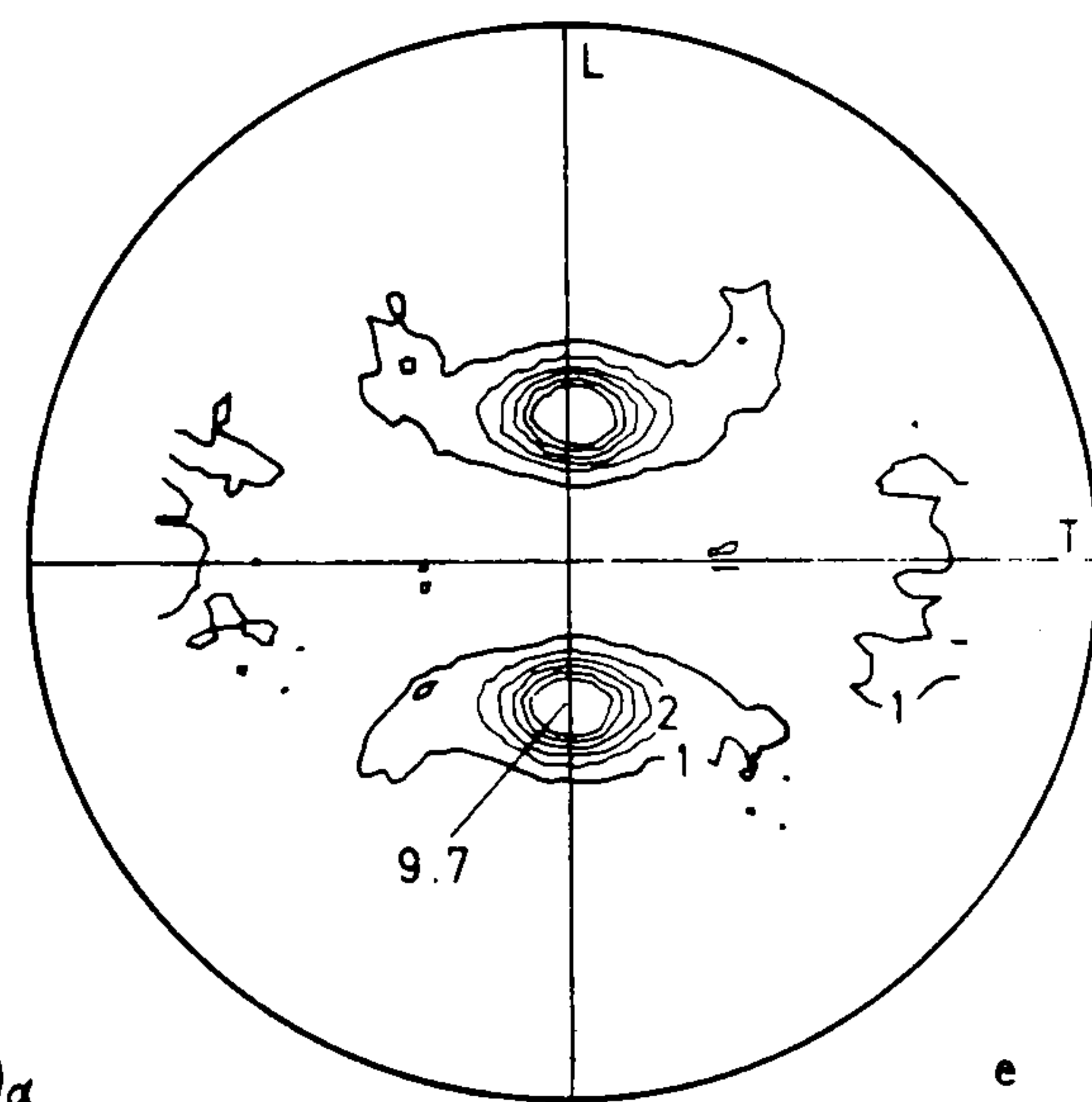
Sheet centre
Contour intervals 1 x random



$(0002)_\alpha$



$(10\bar{1}0)_\alpha$



$(110)_\beta$

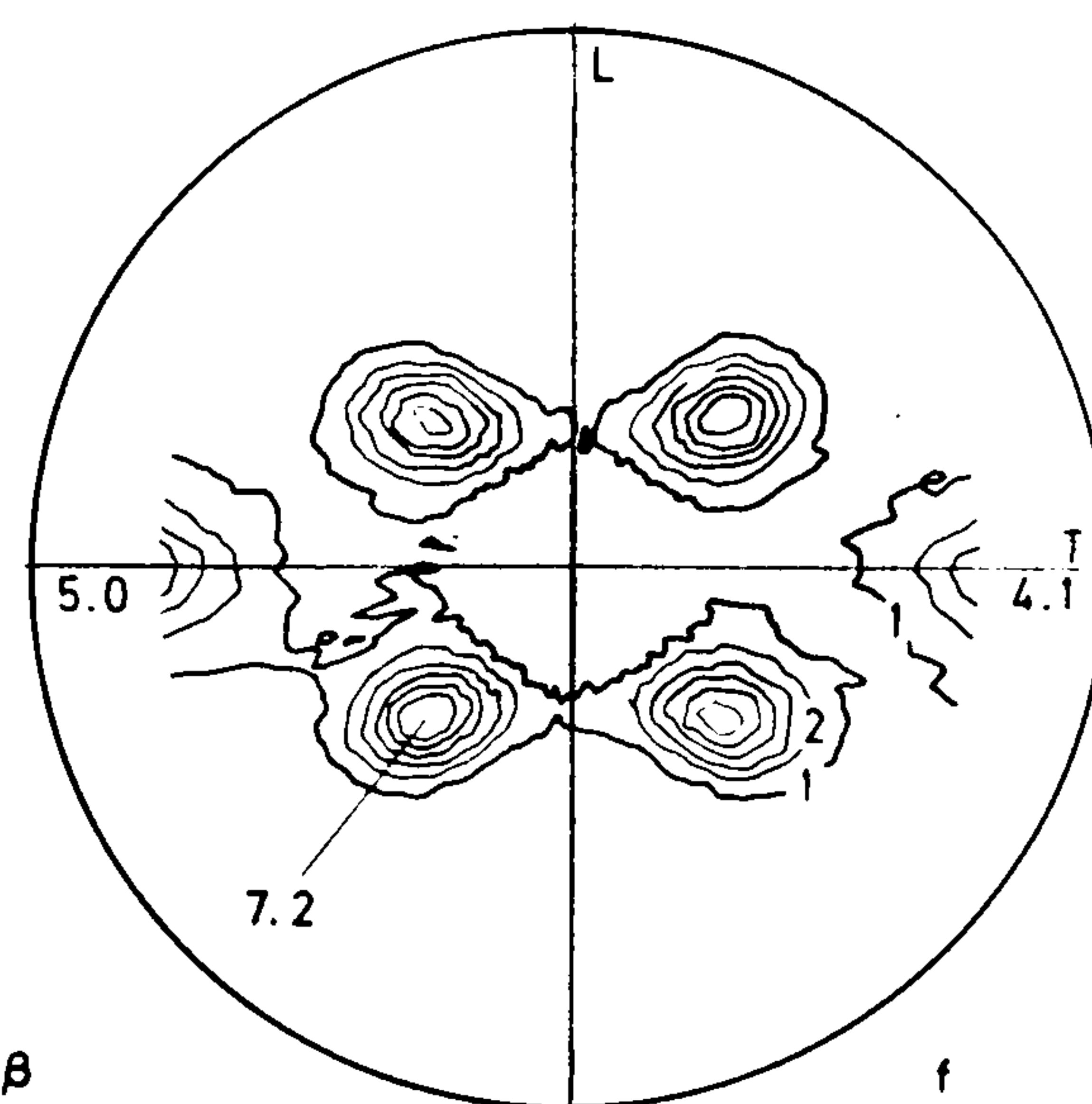


Fig 13 Alpha and beta phase textures of 3.3mm Ti-6Al-4V sheet
in the mill annealed condition (as-received)

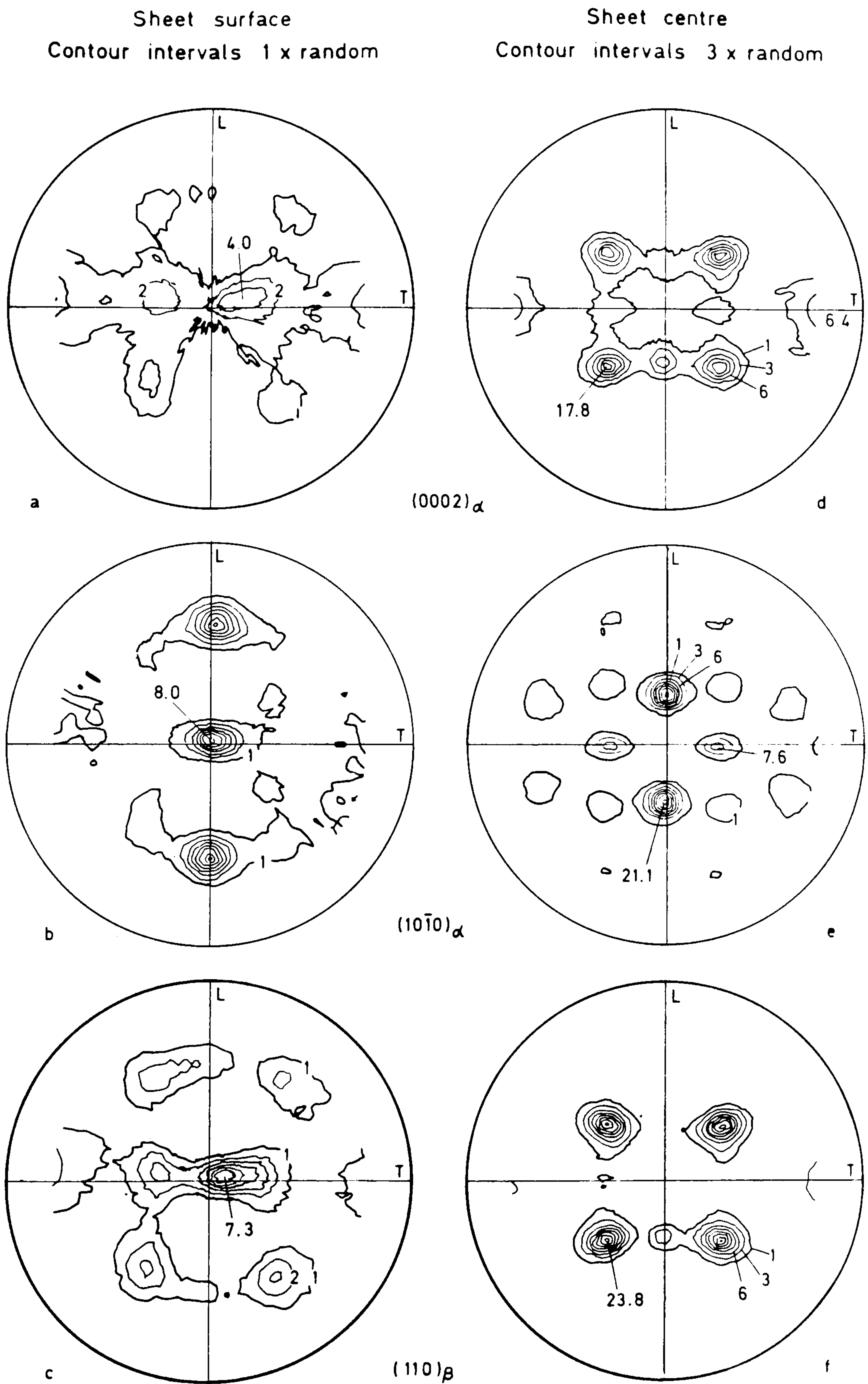


Fig 14 Alpha and beta phase textures of 3.3mm Ti-6Al-4V sheet
after annealing at 925°C for 1hr

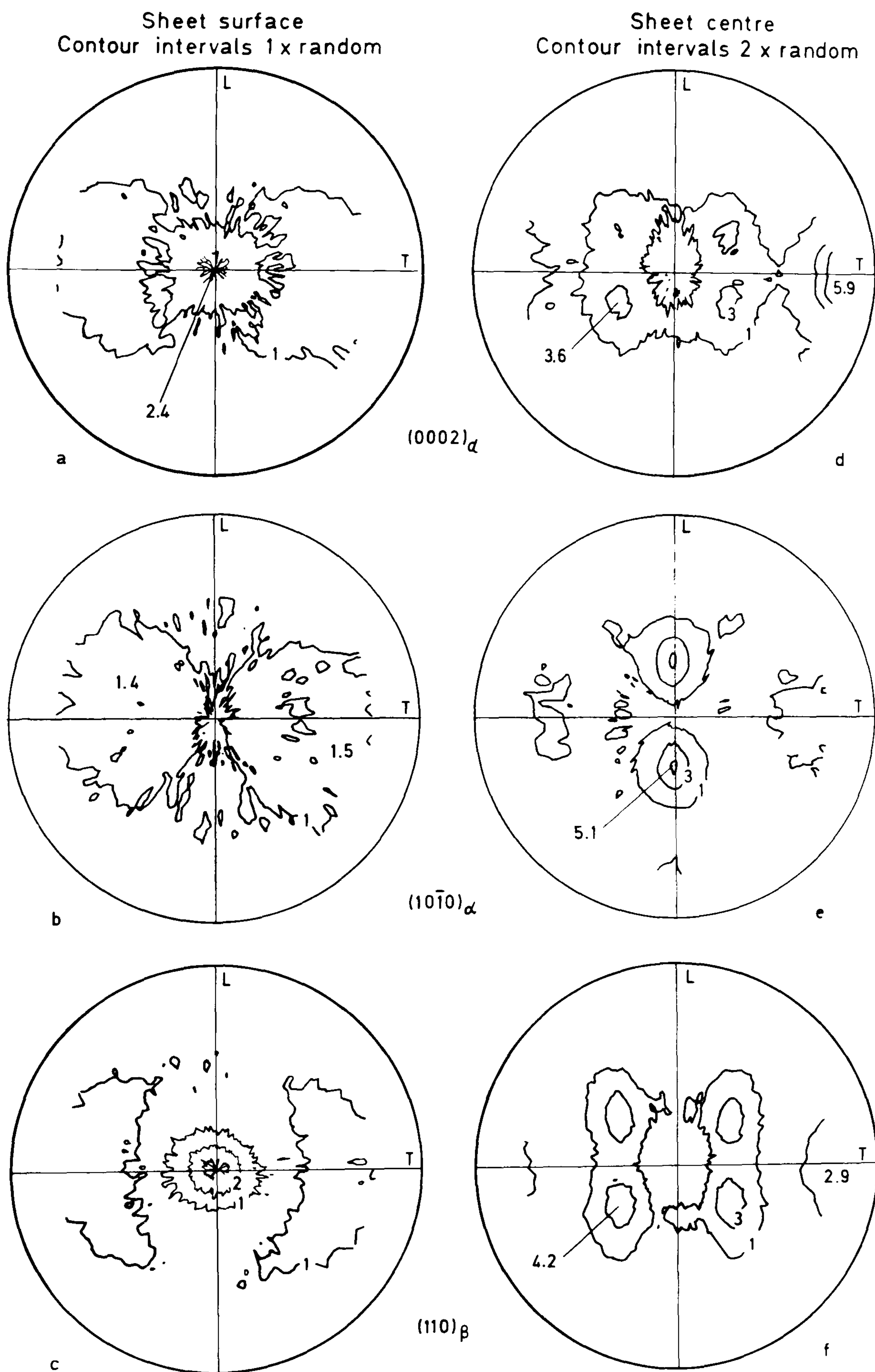
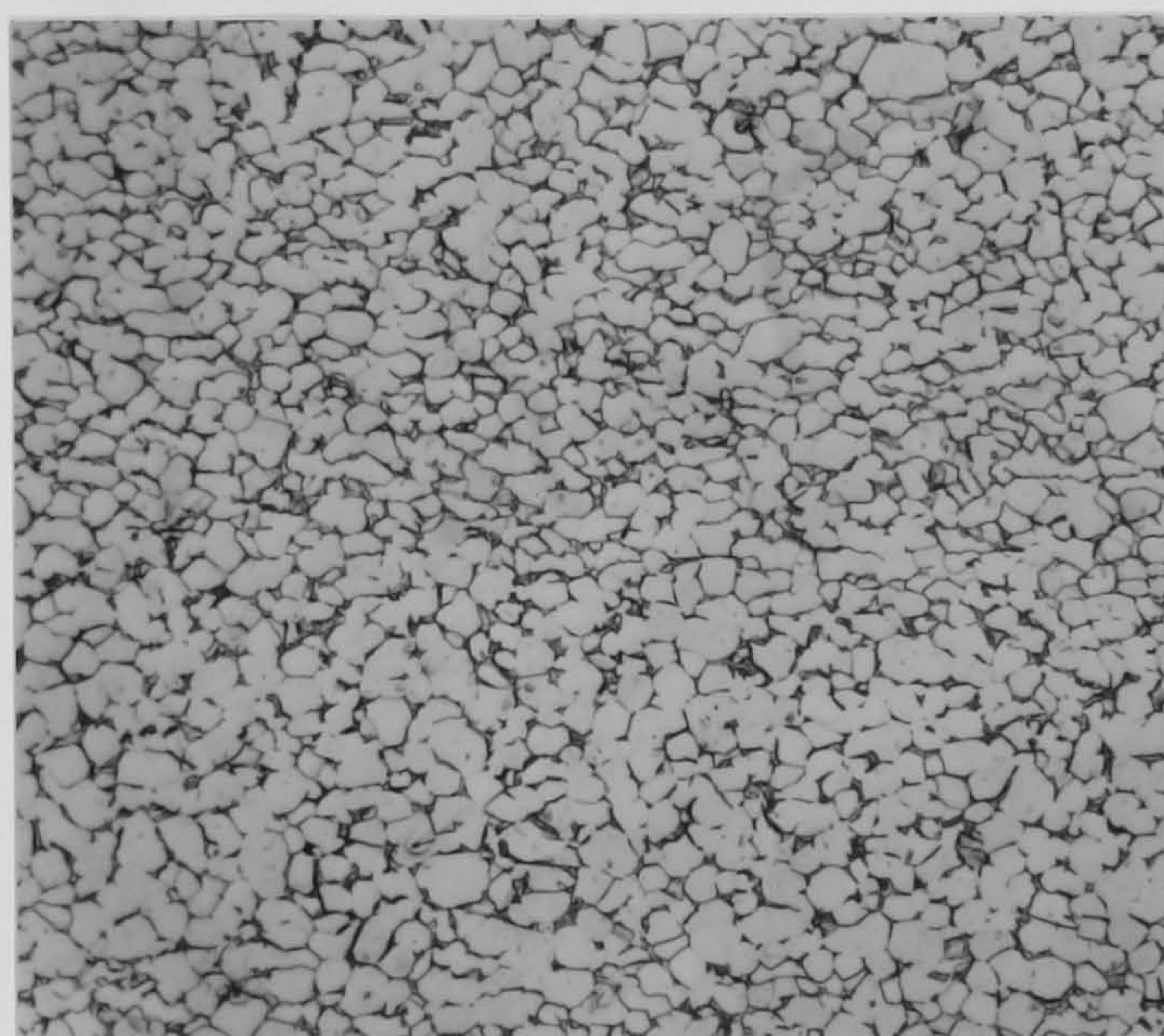


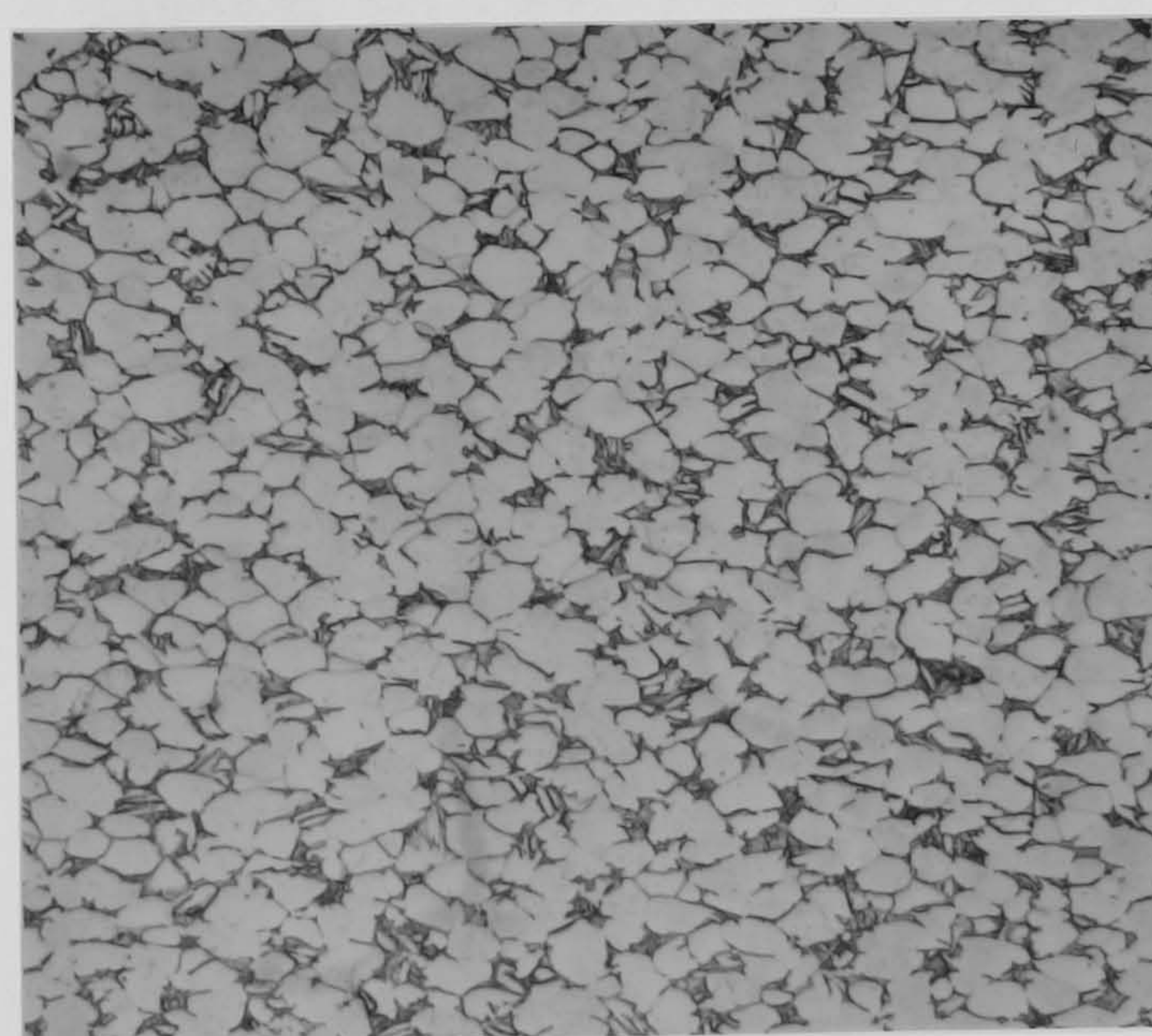
Fig 15 Alpha and beta phase textures of 3.3mm Ti-6Al-4V sheet
after 200% superplastic strain at 925°C and $3 \times 10^{-4} \text{s}^{-1}$



a) As-received

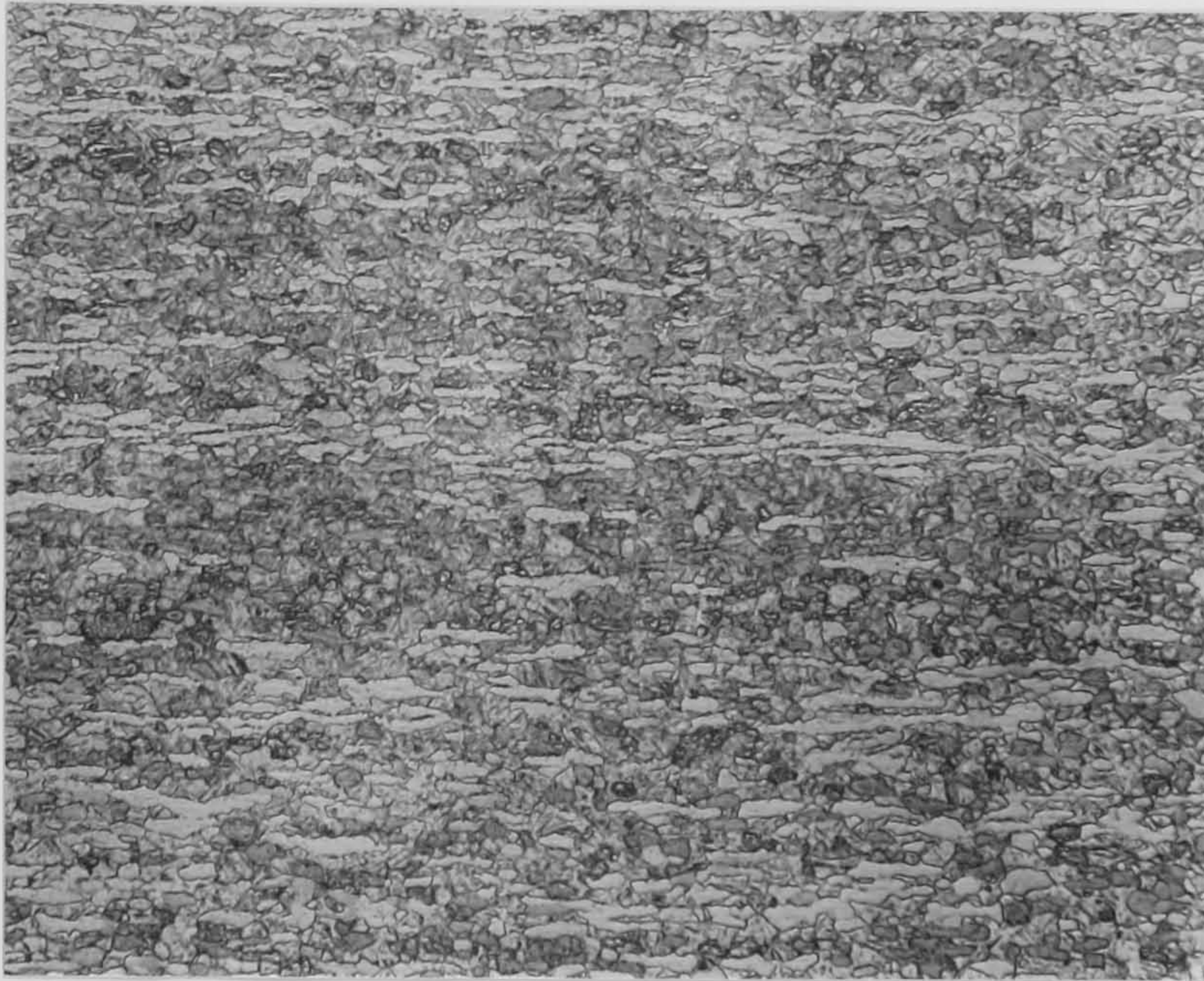


b) Annealed 925 °C 1hr



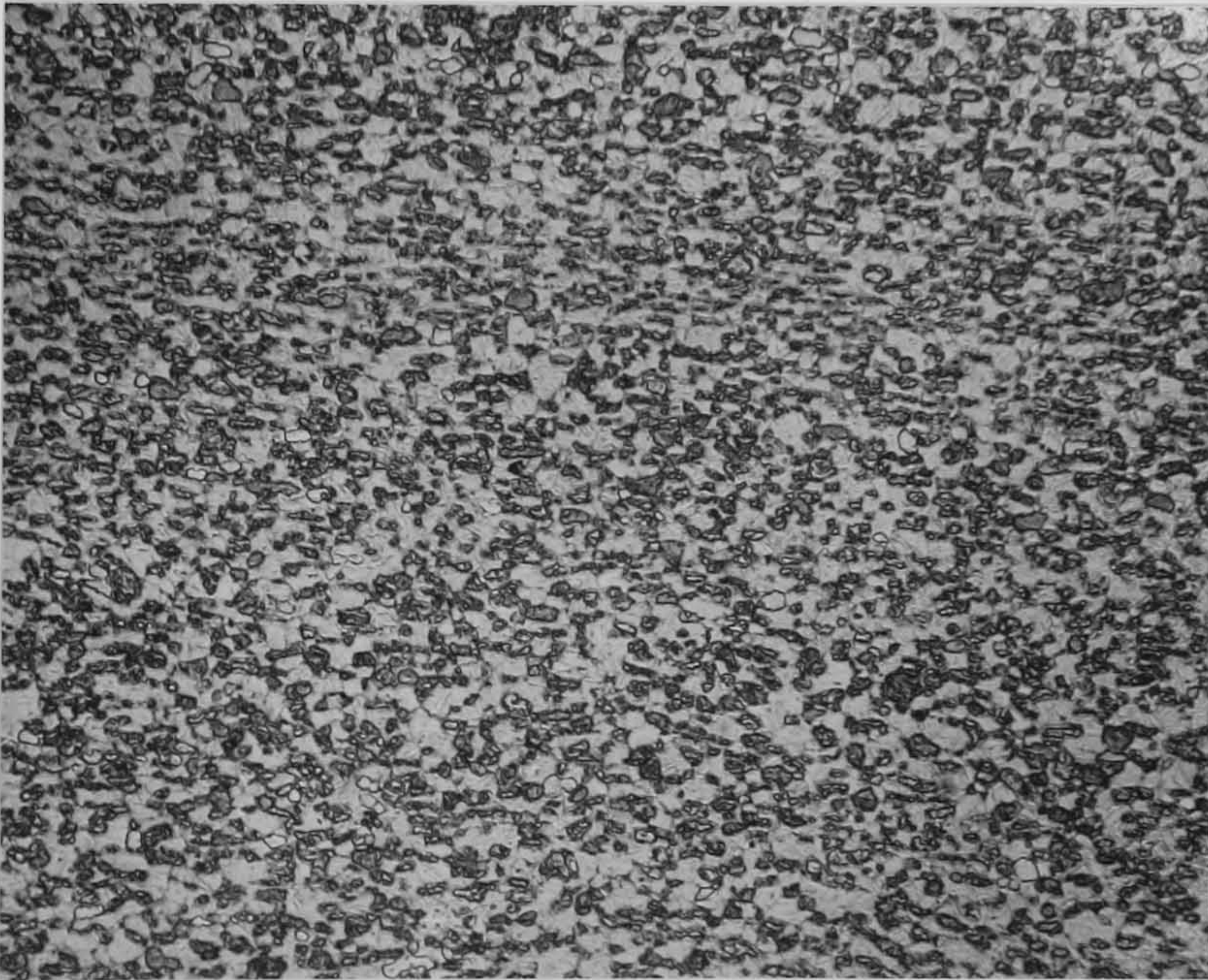
c) After 150% superplastic strain at 925 °C and $3 \times 10^{-4} \text{ s}^{-1}$

Fig 16 Microstructure of 2.0mm Ti-6Al-4V sheet (batch A)



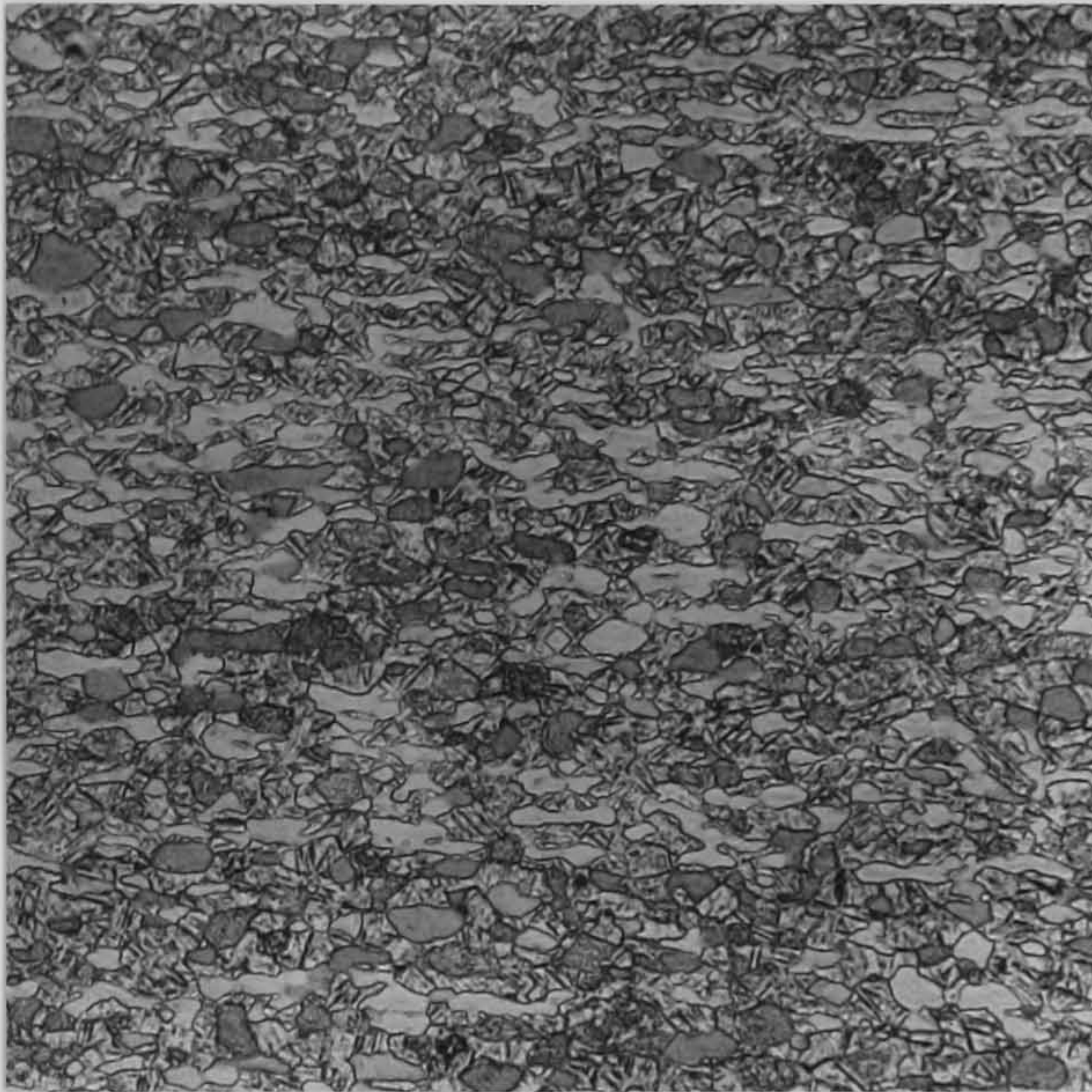
a)

20 μm

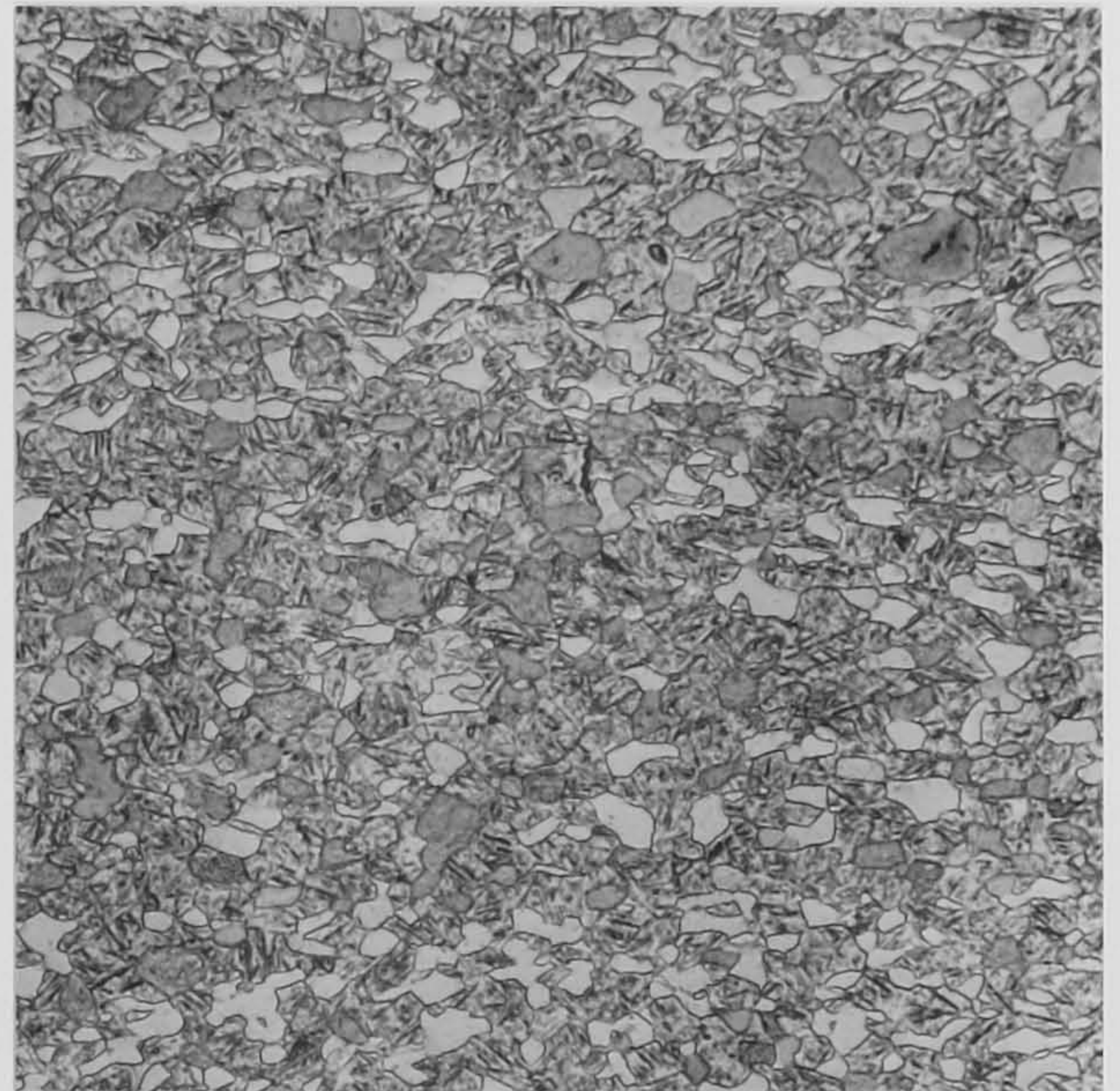


b)

Fig 17 Microstructure of mill annealed 3.3mm Ti-6Al-4V
sheet quenched from 925 °C



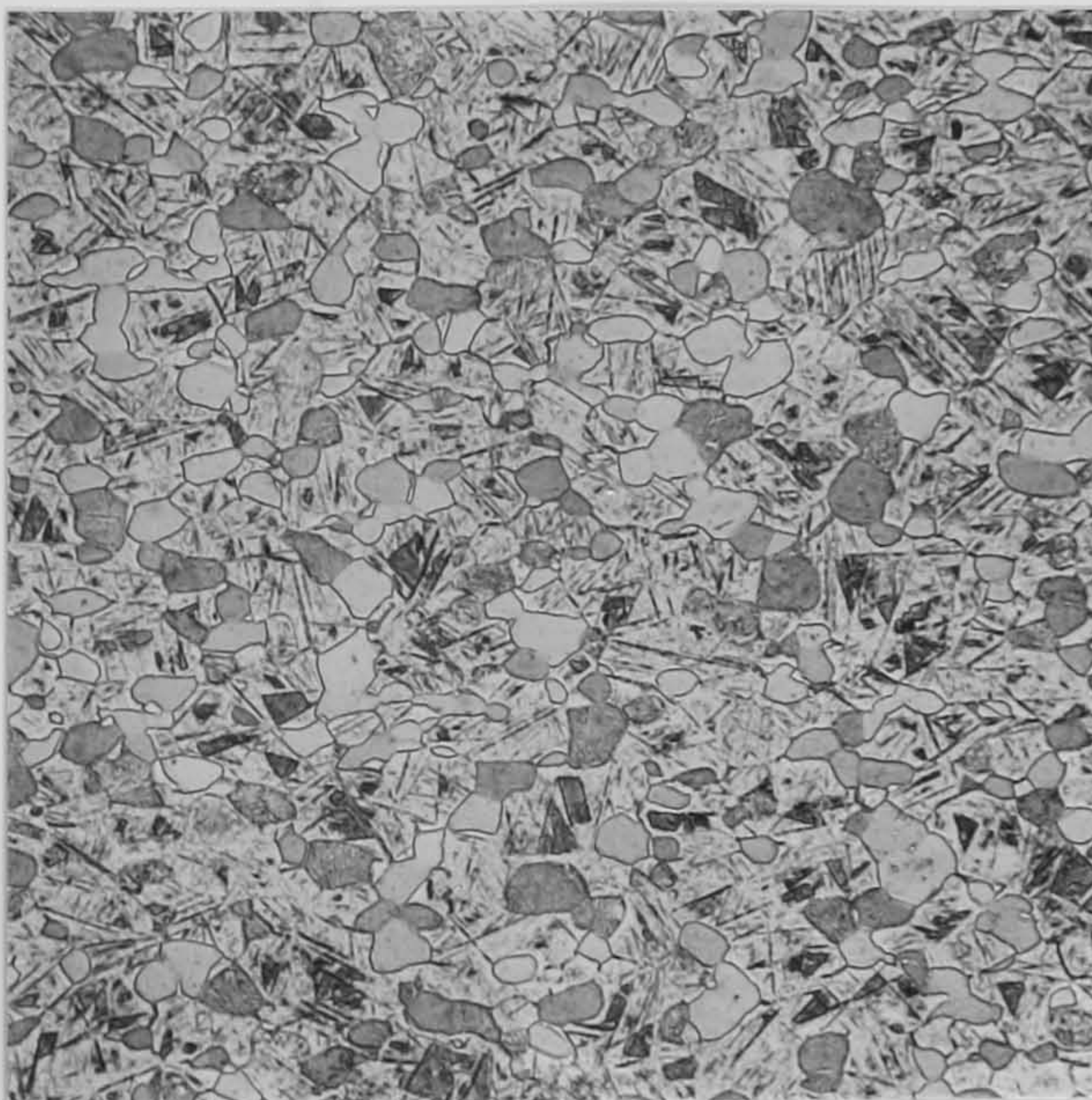
a) 50% strain



b) 140% strain



20 μm



c) 435% strain



d) 735% strain

Fig 18 Microstructure of 3.3mm Ti-6Al-4V sheet after superplastic strain at 925°C and $3 \times 10^{-4} \text{s}^{-1}$. Material reheated to 925°C and quenched

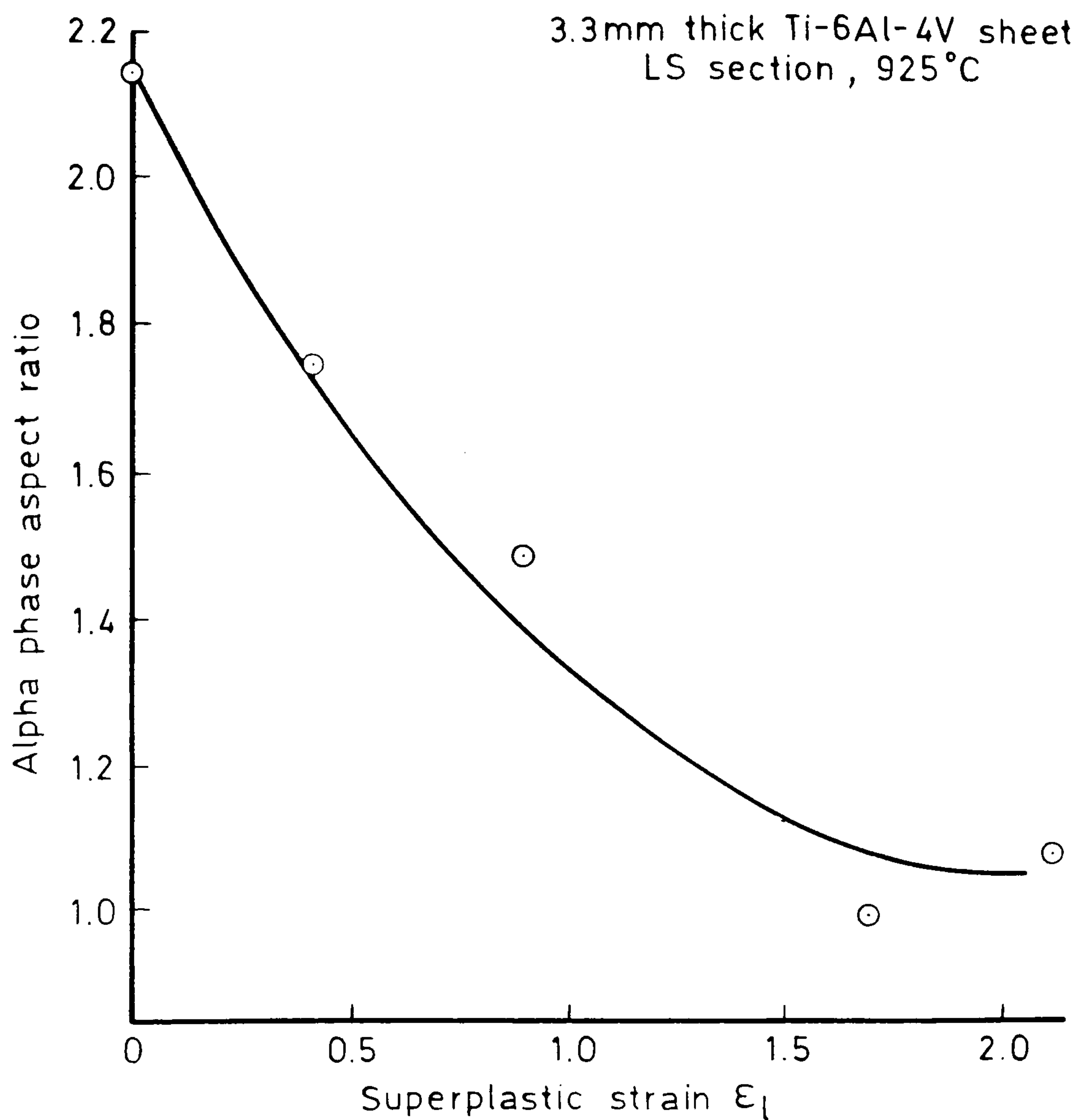


Fig 19 Effect of strain at 925 °C and $3 \times 10^{-4} \text{ s}^{-1}$ on alpha phase aspect ratio of 3.3mm Ti-6Al-4V sheet. Material reheated to 925 °C and quenched.

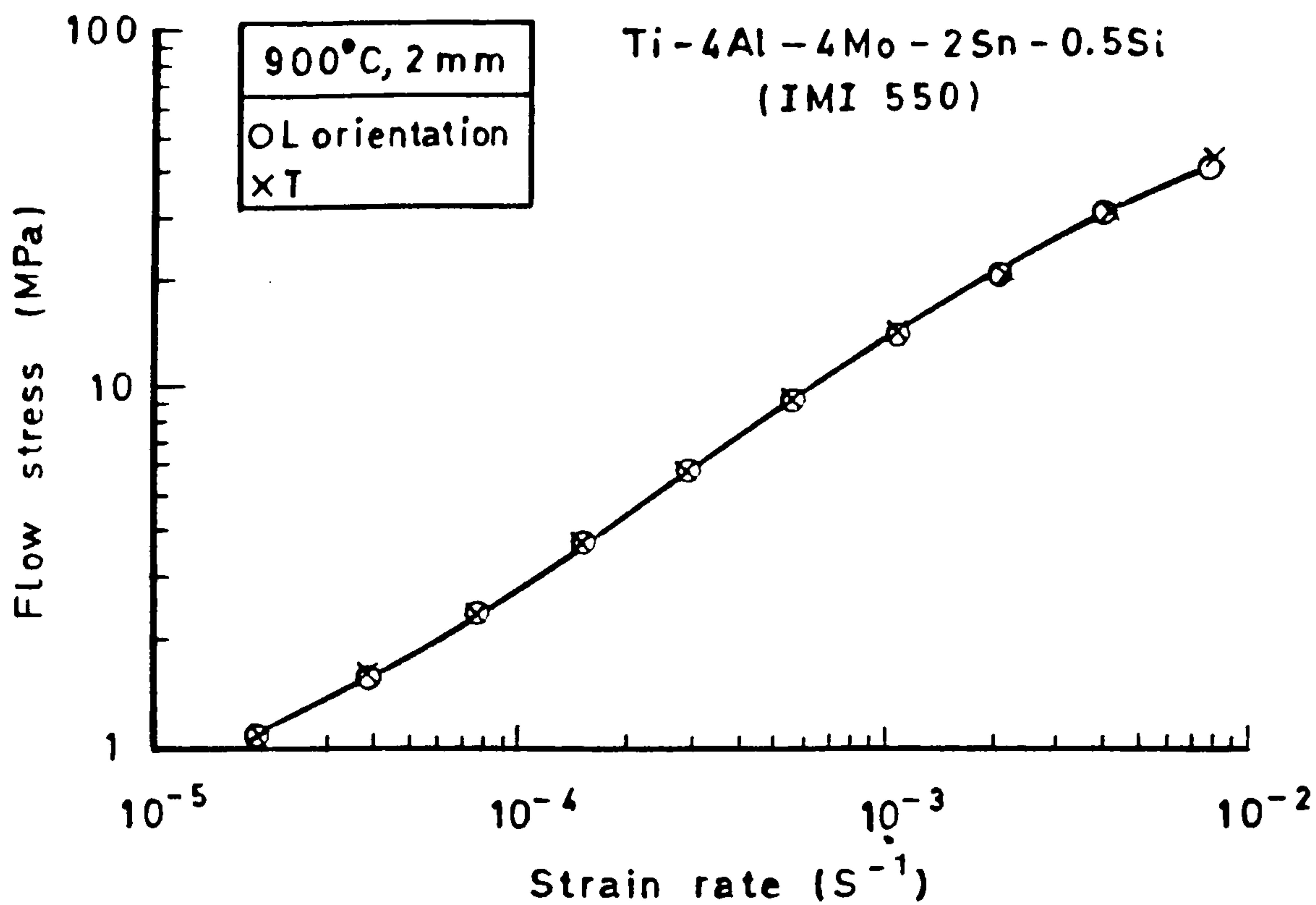


Fig 20 Effect of strain rate on flow stress of IMI 550 sheet pulled at 900 °C

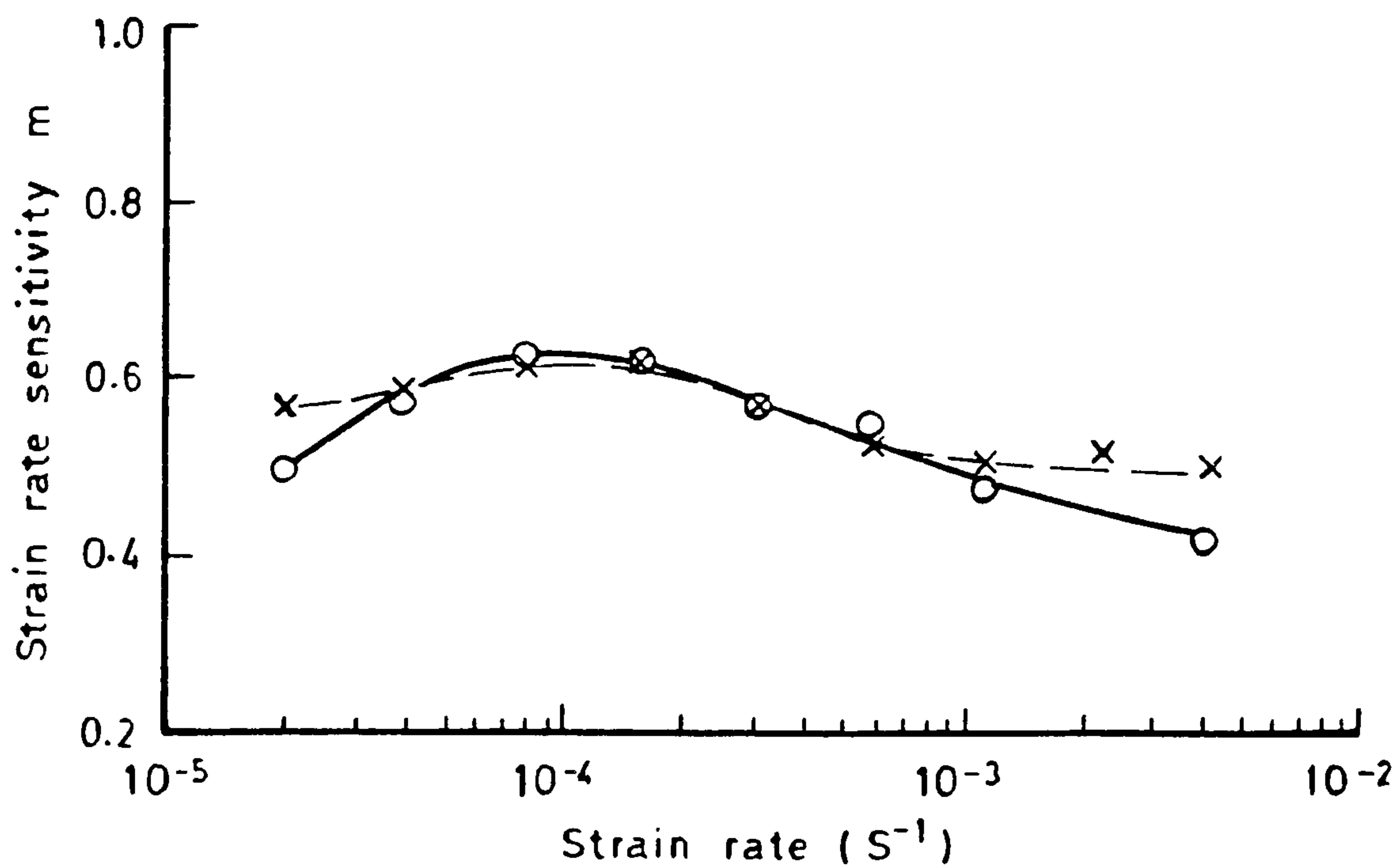


Fig 21 Effect of strain rate on m value of IMI 550 sheet pulled at 900 °C

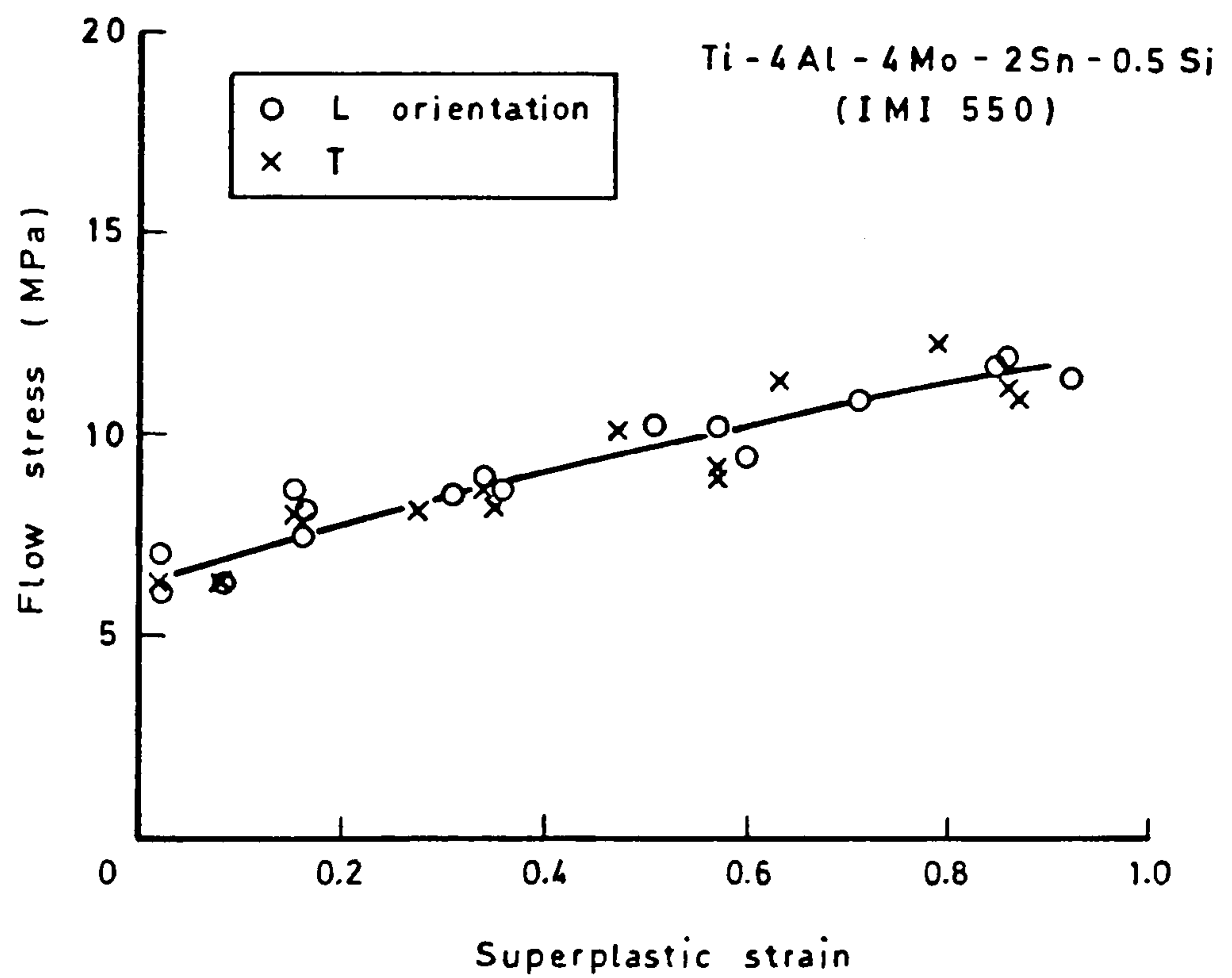


Fig 22 Effect of strain at 900°C and $3 \times 10^{-4}\text{s}^{-1}$
on flow stress of IMI 550 sheet

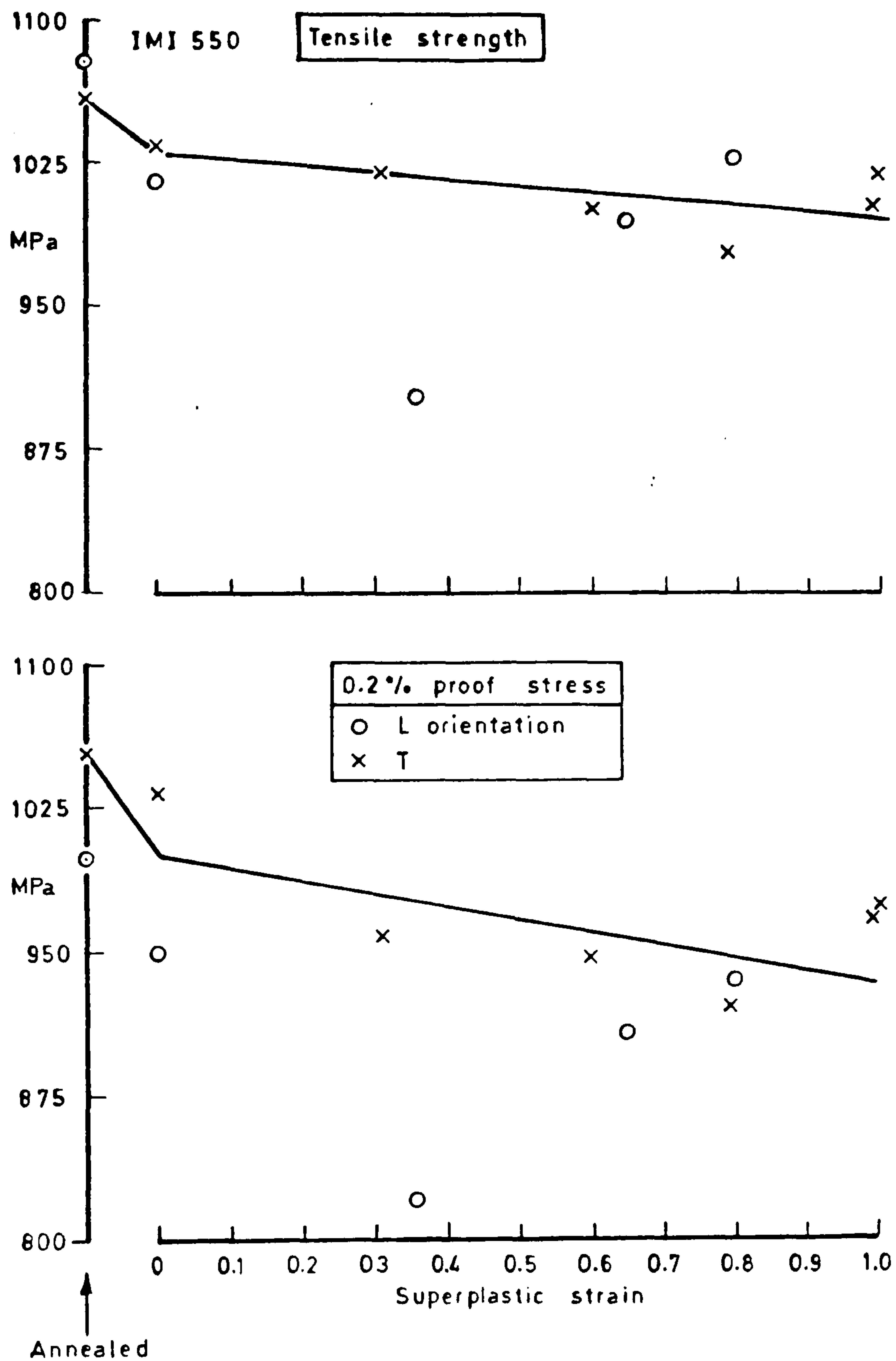
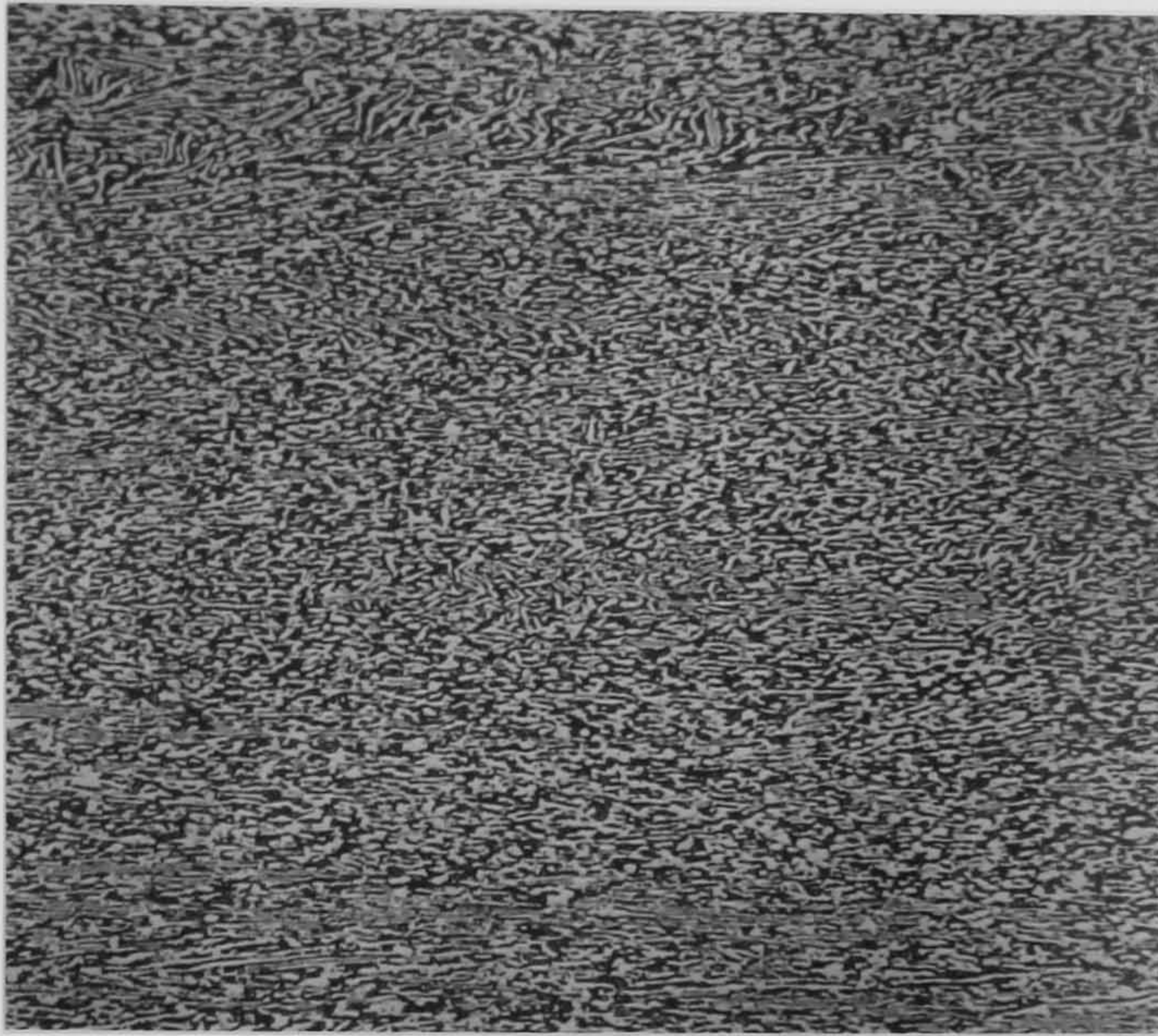
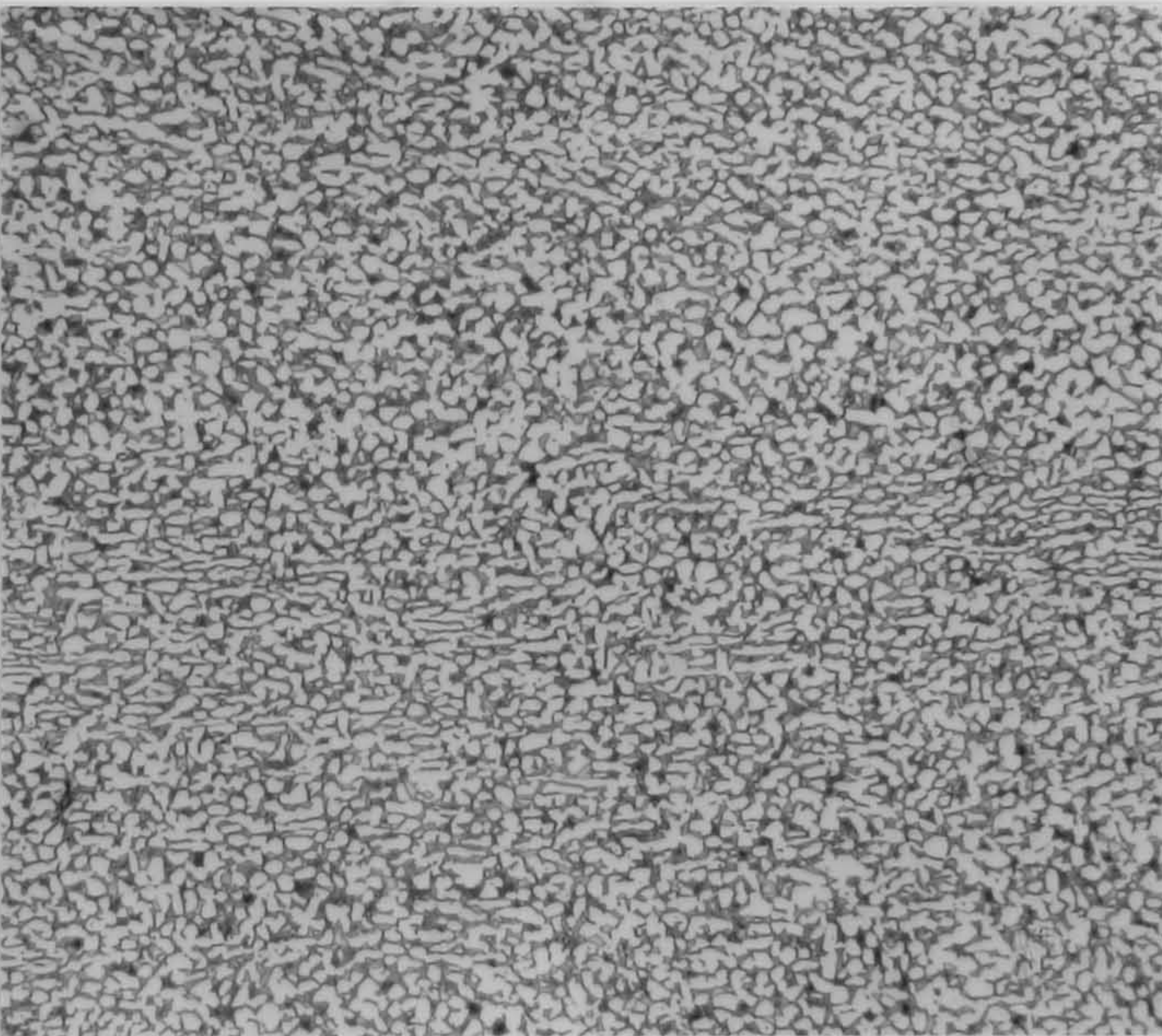


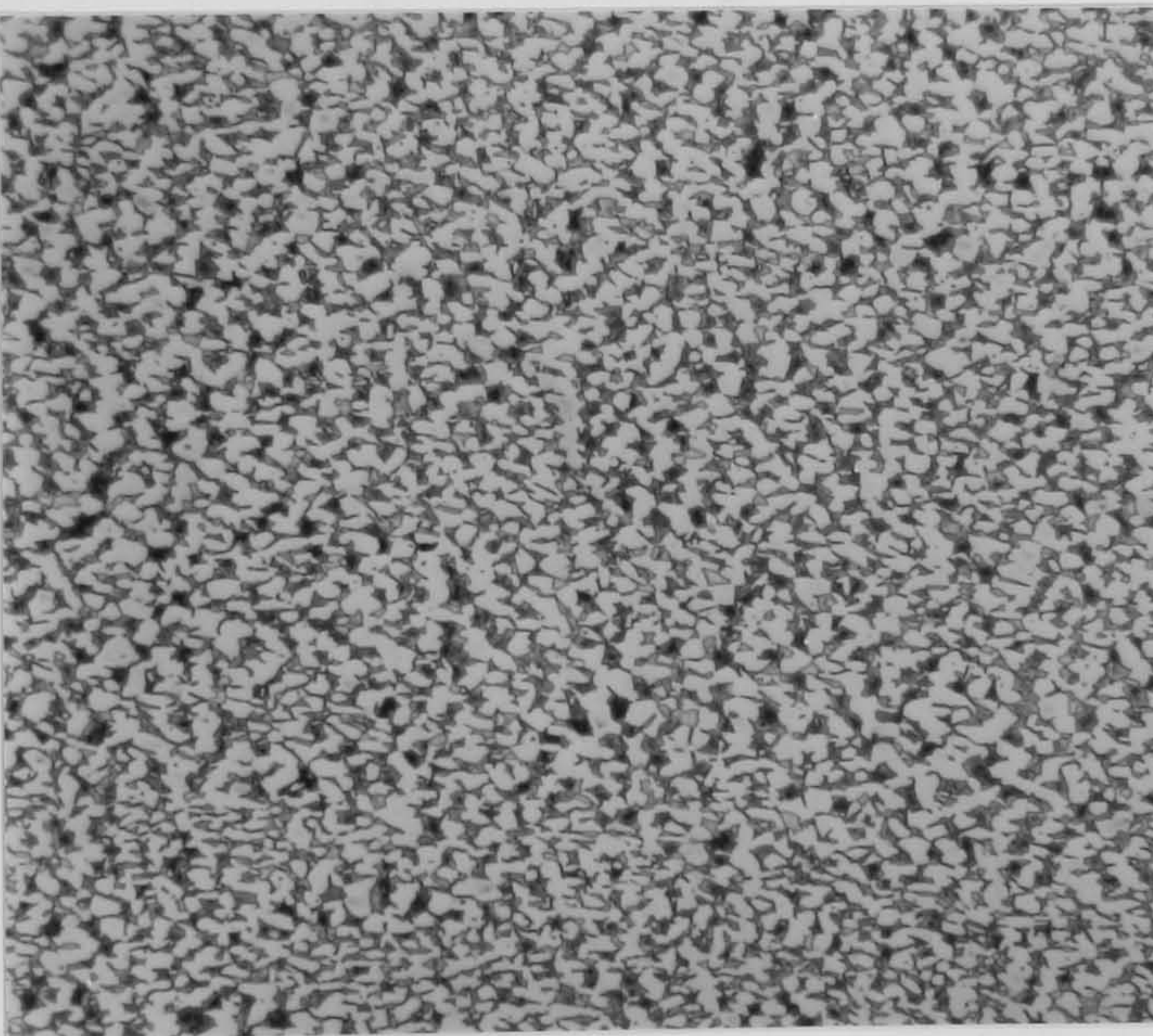
Fig 23 Effect of superplastic strain at 900°C and $3 \times 10^{-4} \text{s}^{-1}$ on room temperature strength of IMI 550 sheet



a) As-received



b) Annealed 900 °C 1hr



c) After 150% superplastic strain at 900 °C and $3 \times 10^{-4} \text{ s}^{-1}$

Fig 24 Microstructure of IMI 550 sheet

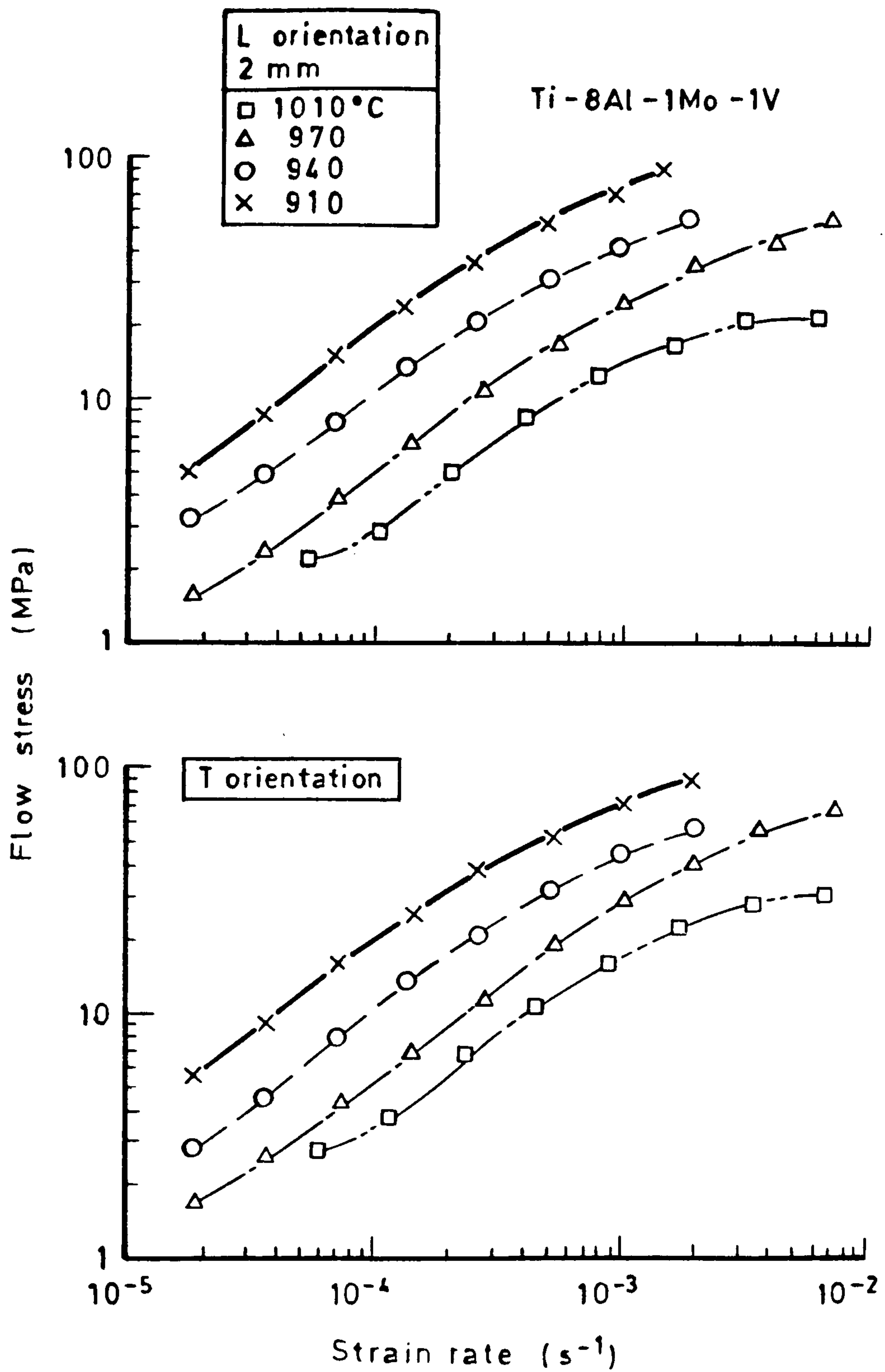


Fig 25 Effect of strain rate on flow stress of Ti-8Al-1Mo-1V sheet pulled at 910, 940, 970 and 1010 °C

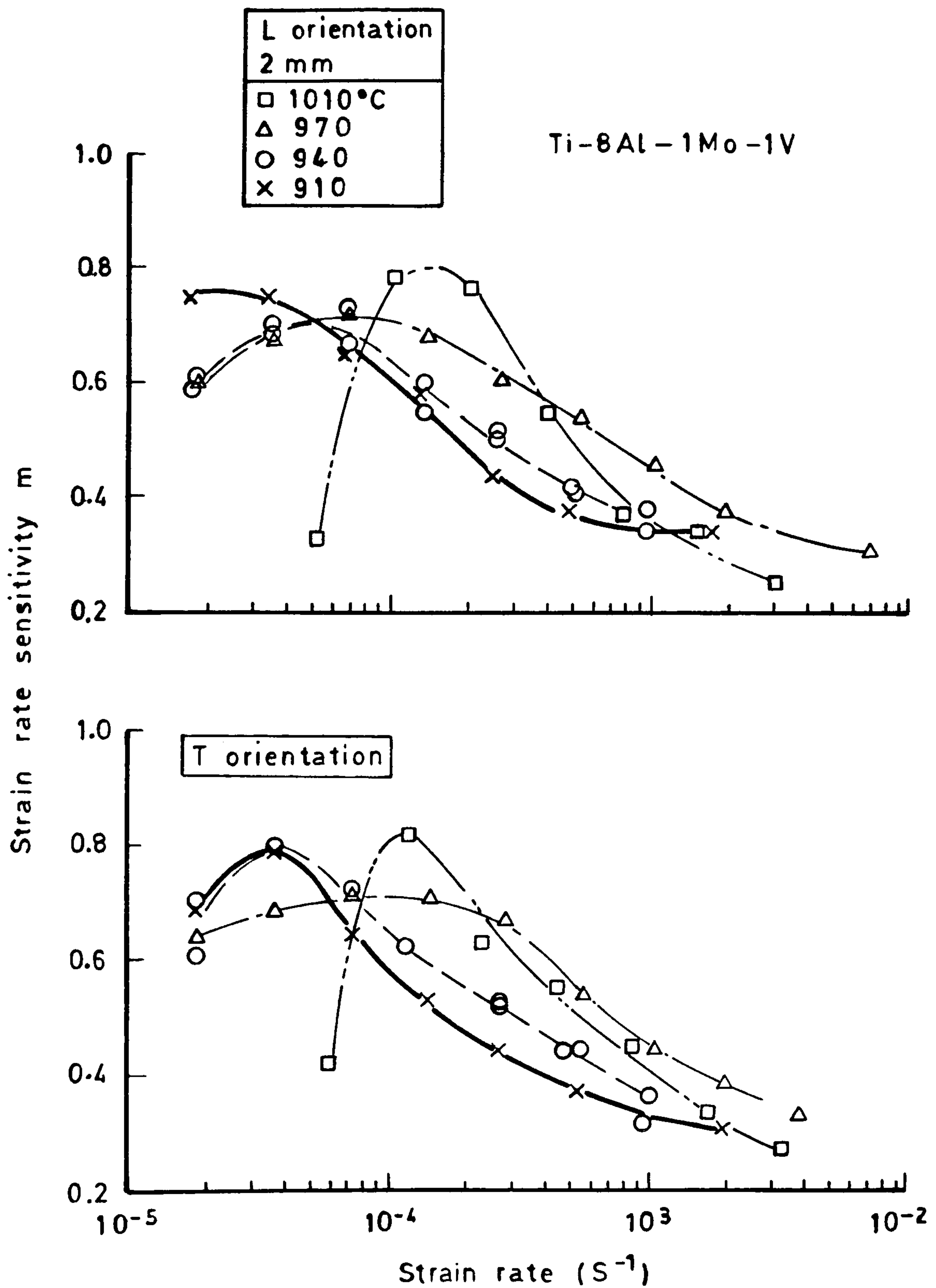


Fig 26 Effect of strain rate on m value of Ti-8Al-1Mo-1V sheet pulled at 910, 940, 970 and 1010 °C

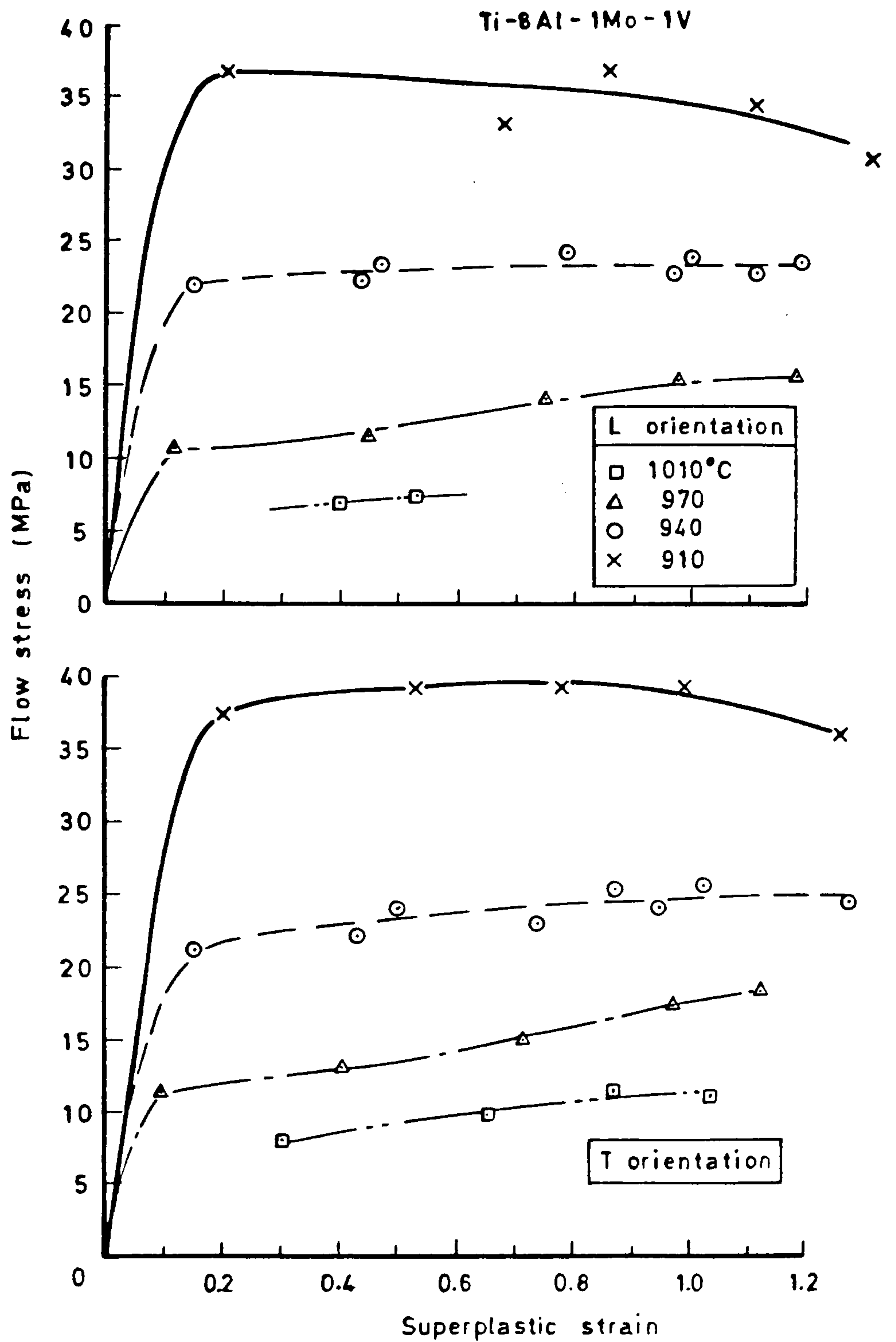
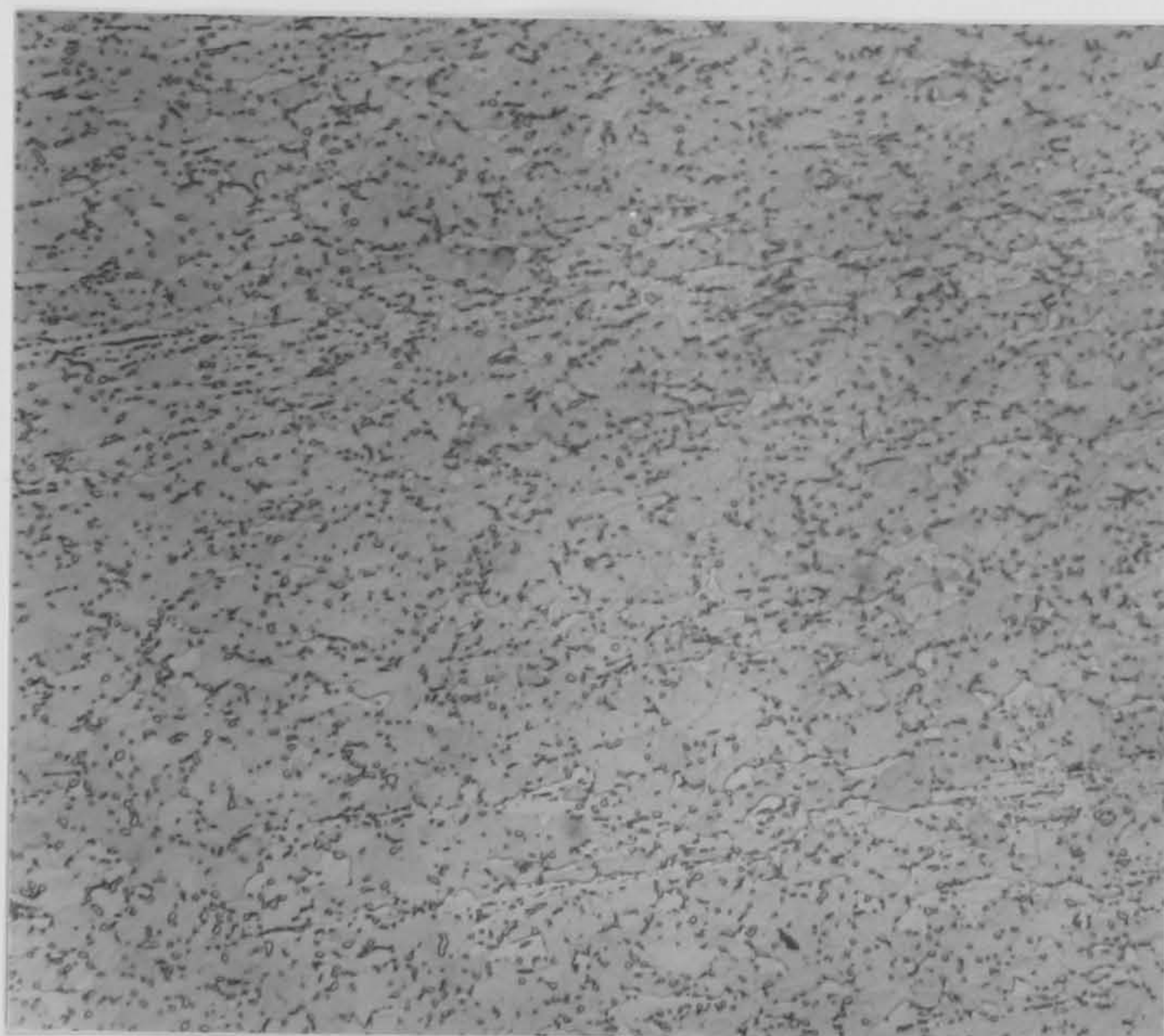
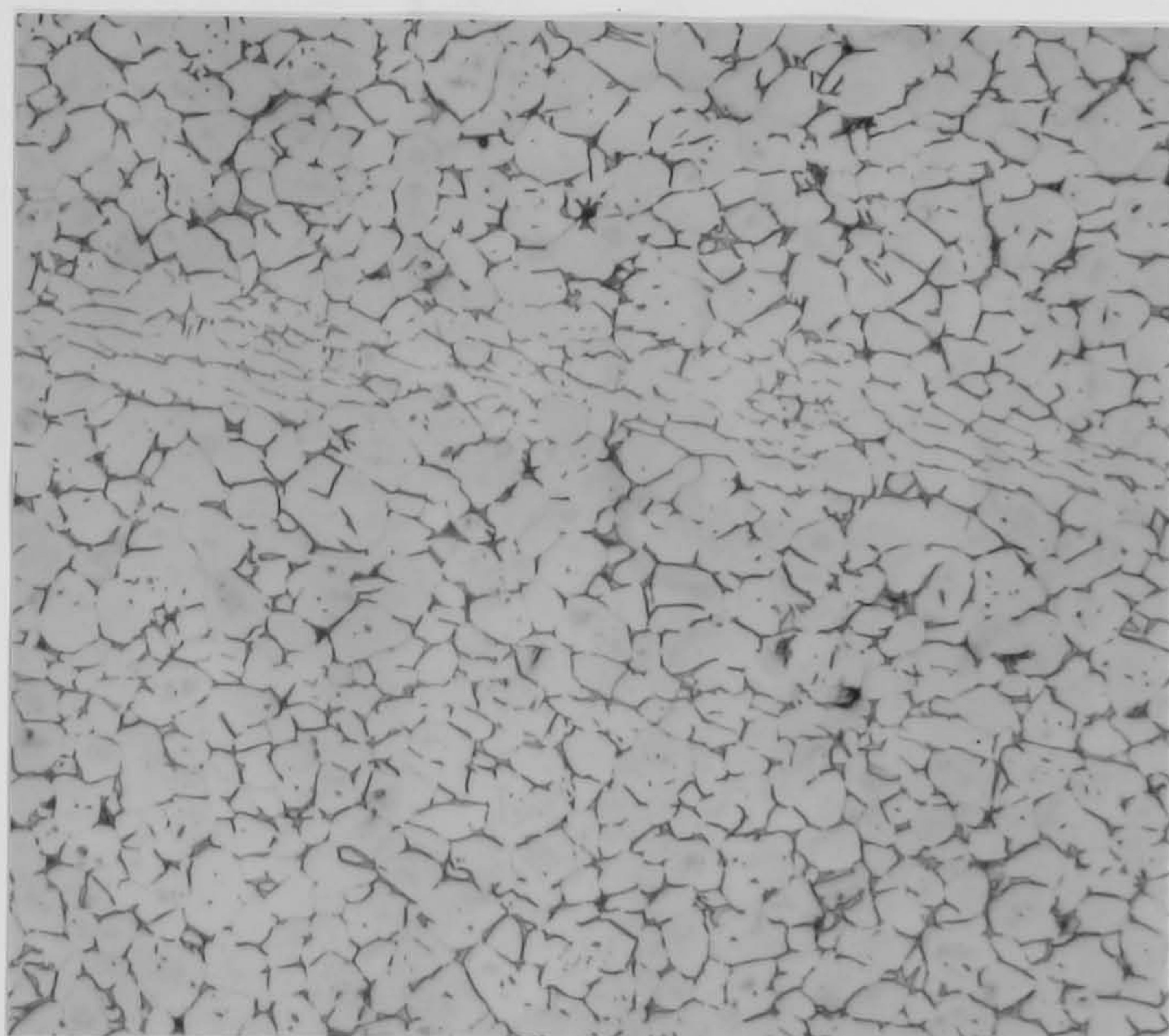


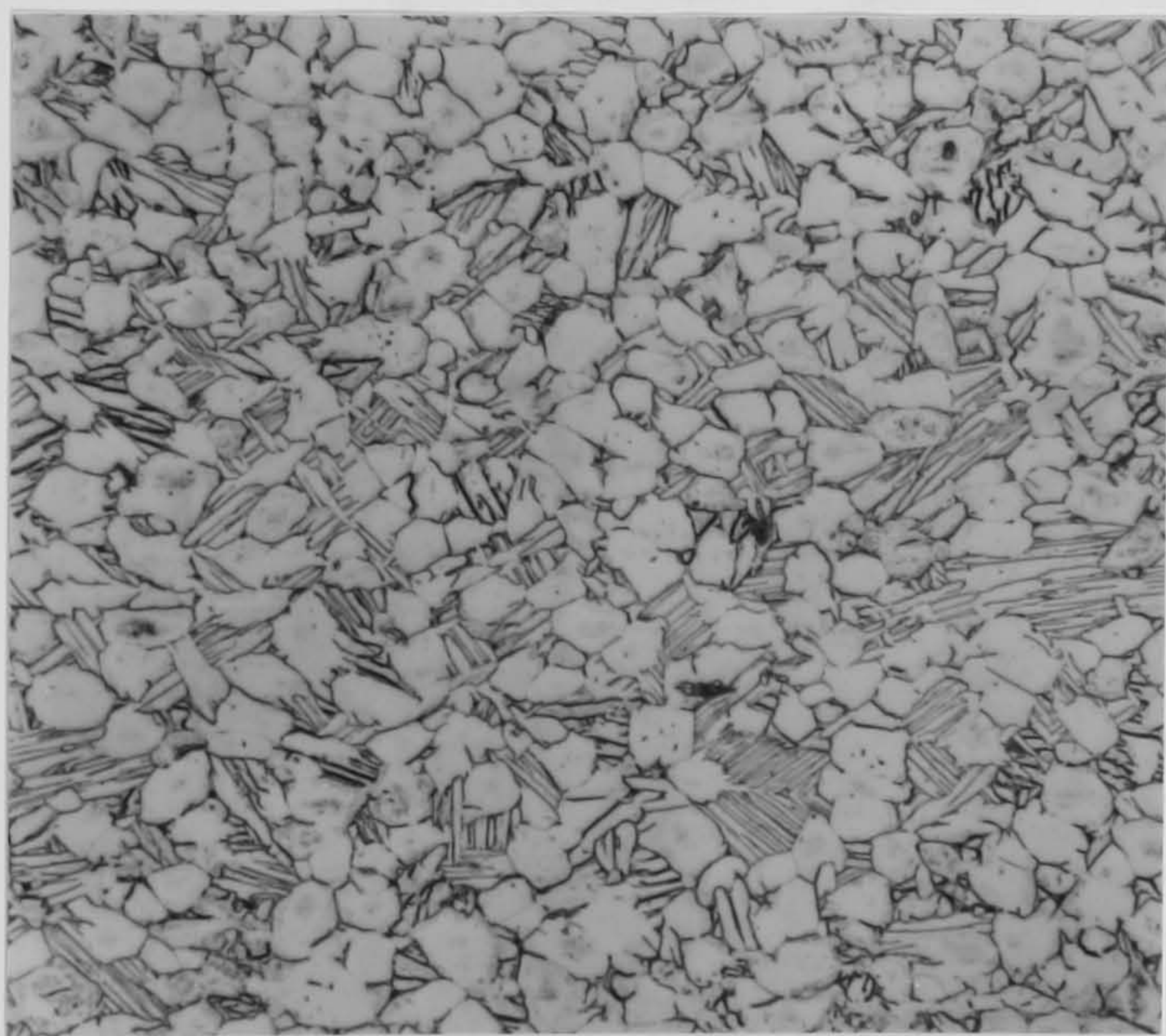
Fig 27 Effect of strain at 910, 940, 970 and 1010 °C and $3 \times 10^{-4} \text{ s}^{-1}$ on flow stress of Ti-8Al-1Mo-1V sheet



a) As-received



b) Annealed 1010 °C 1hr



c) After 150% superplastic strain at 1010 °C and $3 \times 10^{-4} \text{ s}^{-1}$

Fig 28 Microstructure of Ti-8Al-1Mo-1V sheet

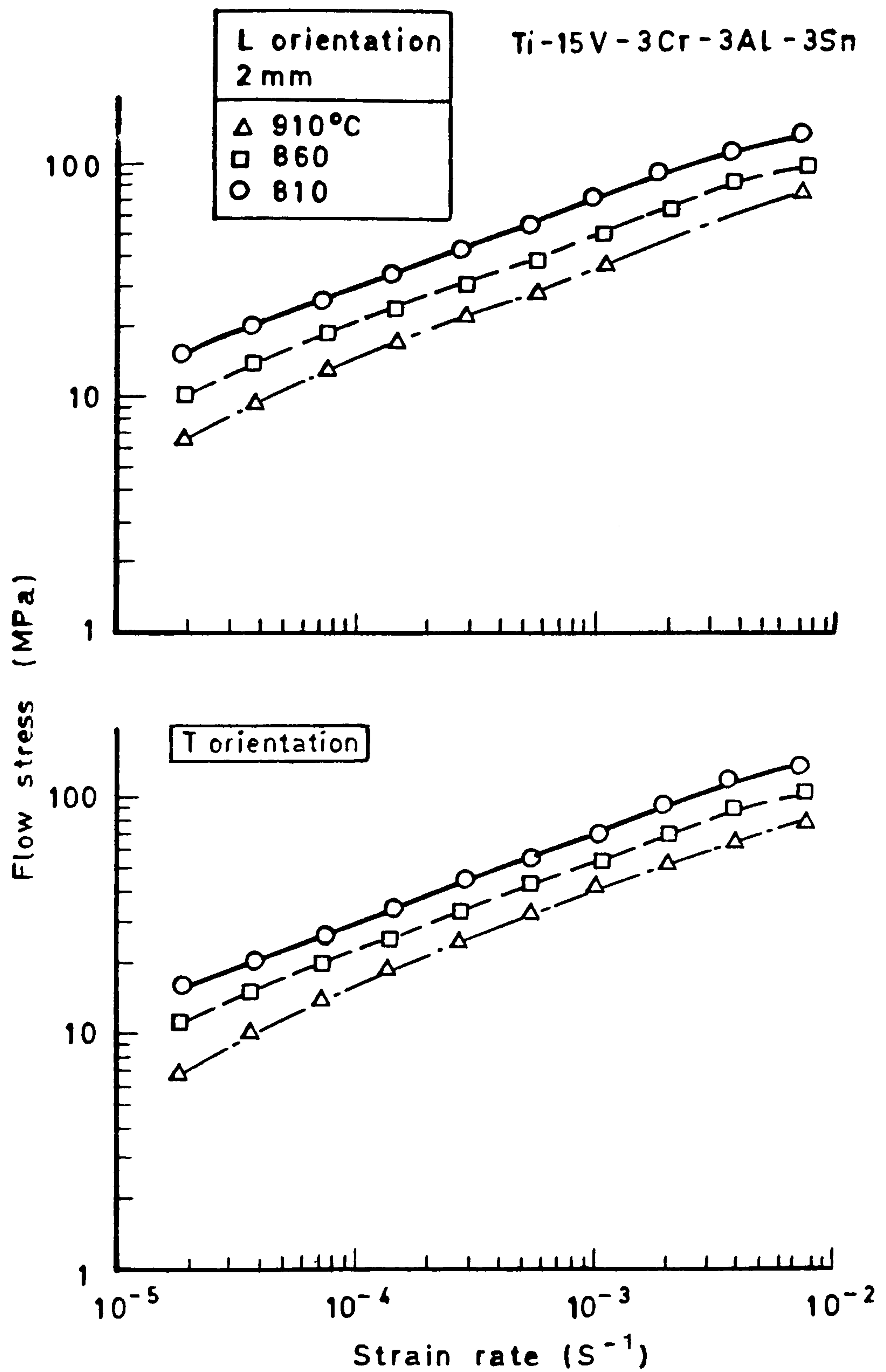


Fig 29 Effect of strain rate on flow stress of Ti-15V-3Cr-3Al-3Sn sheet pulled at 810, 860 and 910 °C

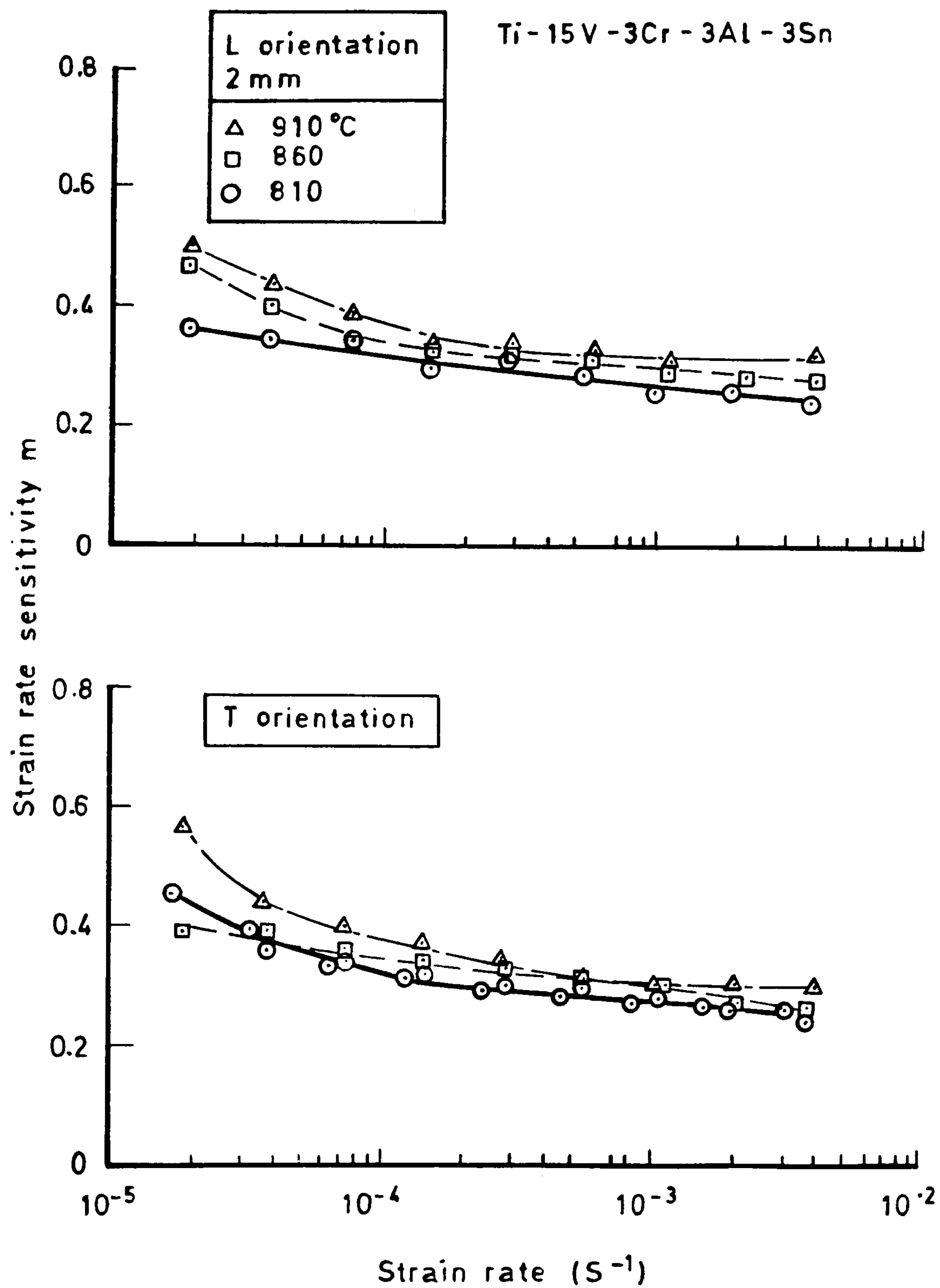


Fig 30 Effect of strain rate on m value of Ti-15V-3Cr-3Al-3Sn sheet pulled at 810, 860, and 910 °C

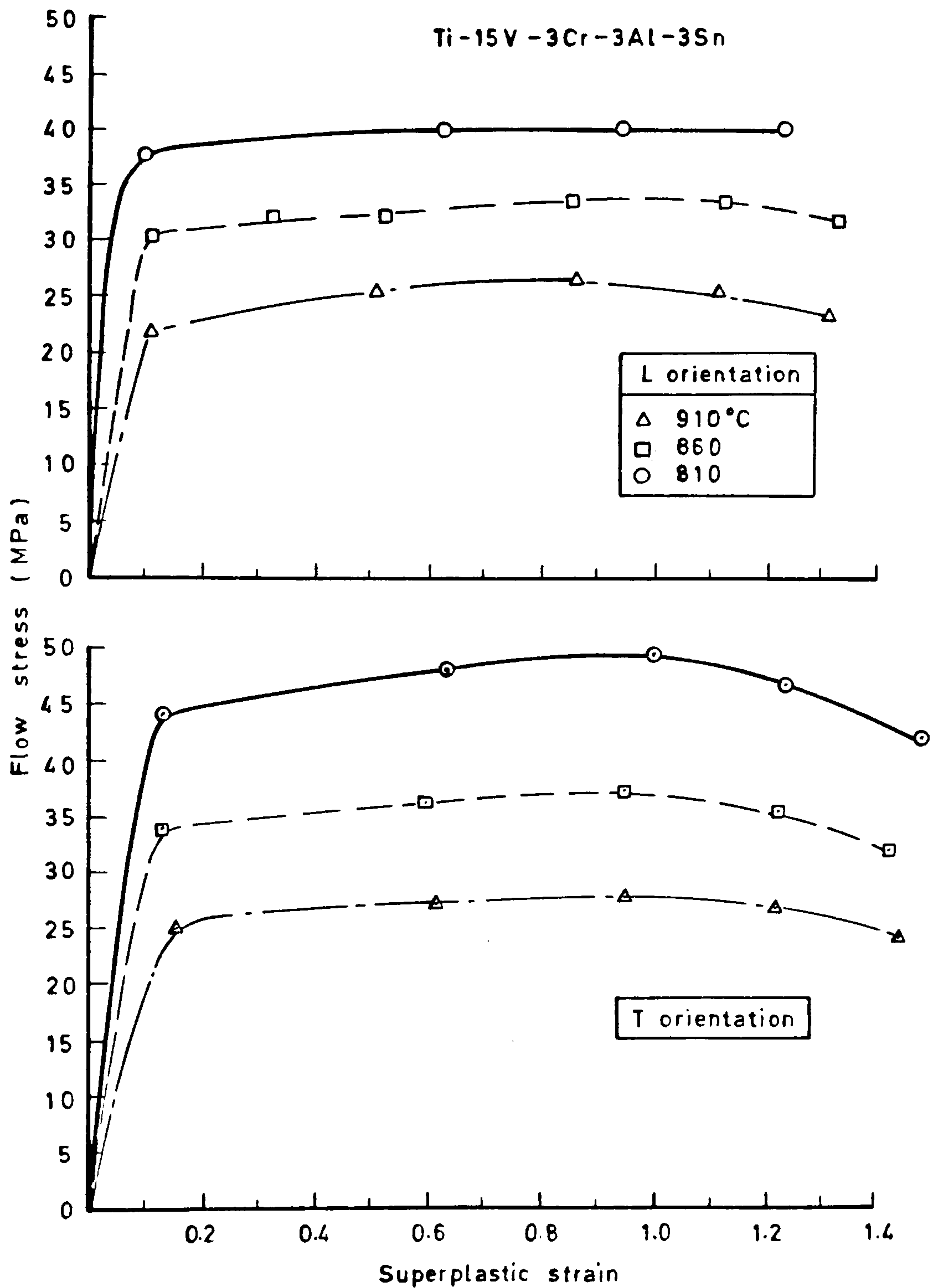
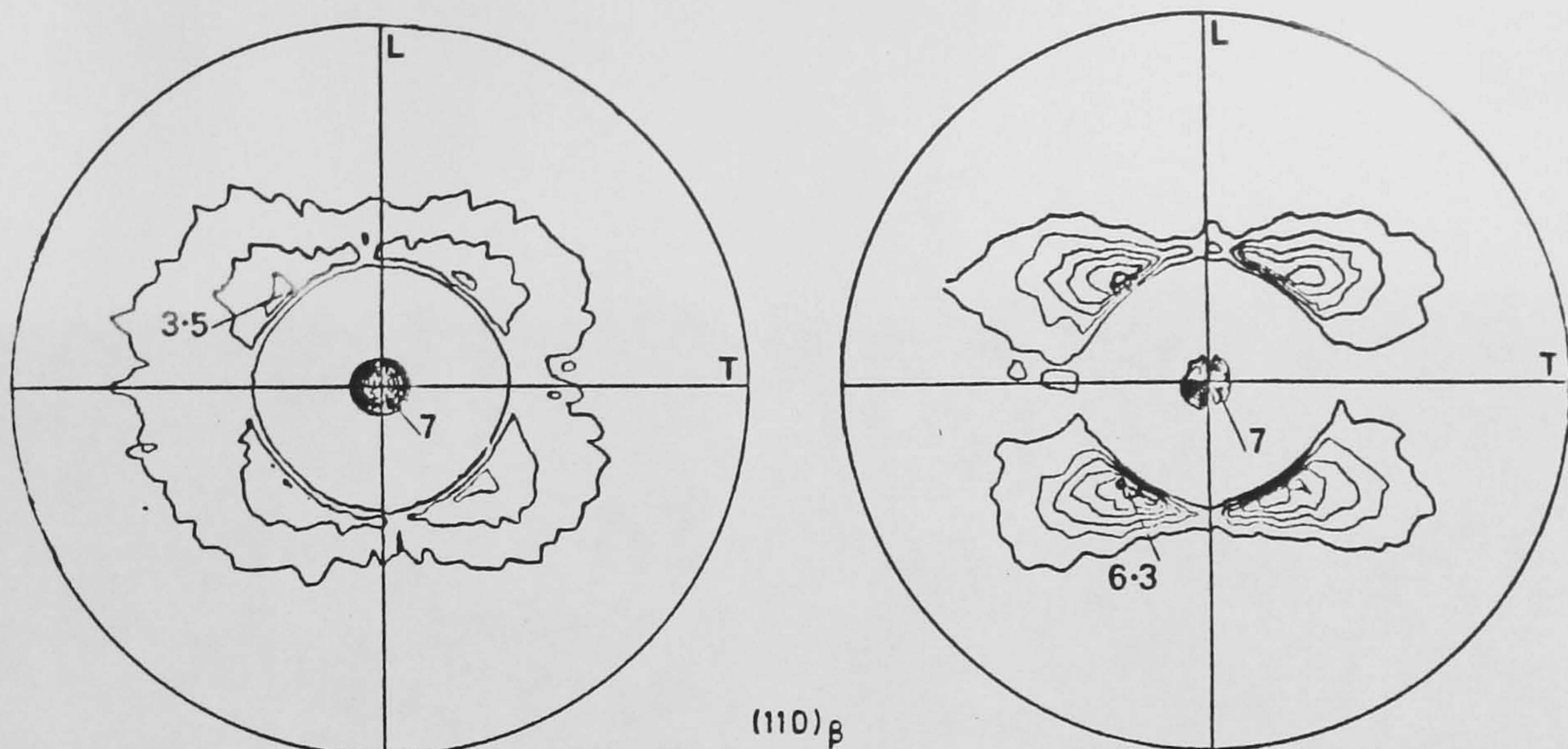


Fig 31 Effect of strain at 810, 860 and 910 °C and $3 \times 10^{-4} \text{ s}^{-1}$ on flow stress of Ti-15V-3Cr-3Al-3Sn sheet



a) As-received

b) After 238% strain at 810 °C
and $3 \times 10^{-4} \text{ s}^{-1}$

Fig 32 Texture of Ti-15V-3Cr-3Al-3Sn sheet.

Contour intervals 1 x random

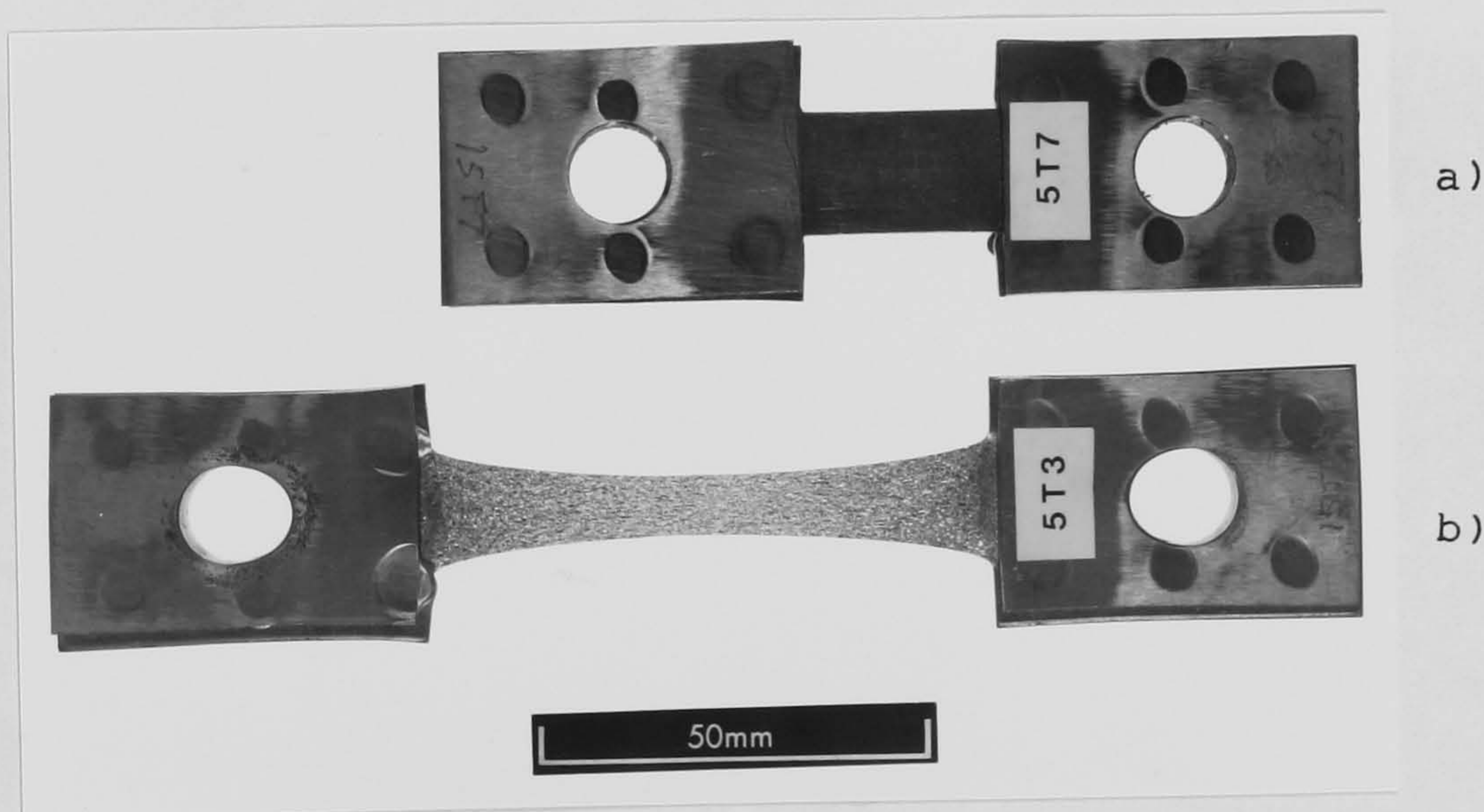
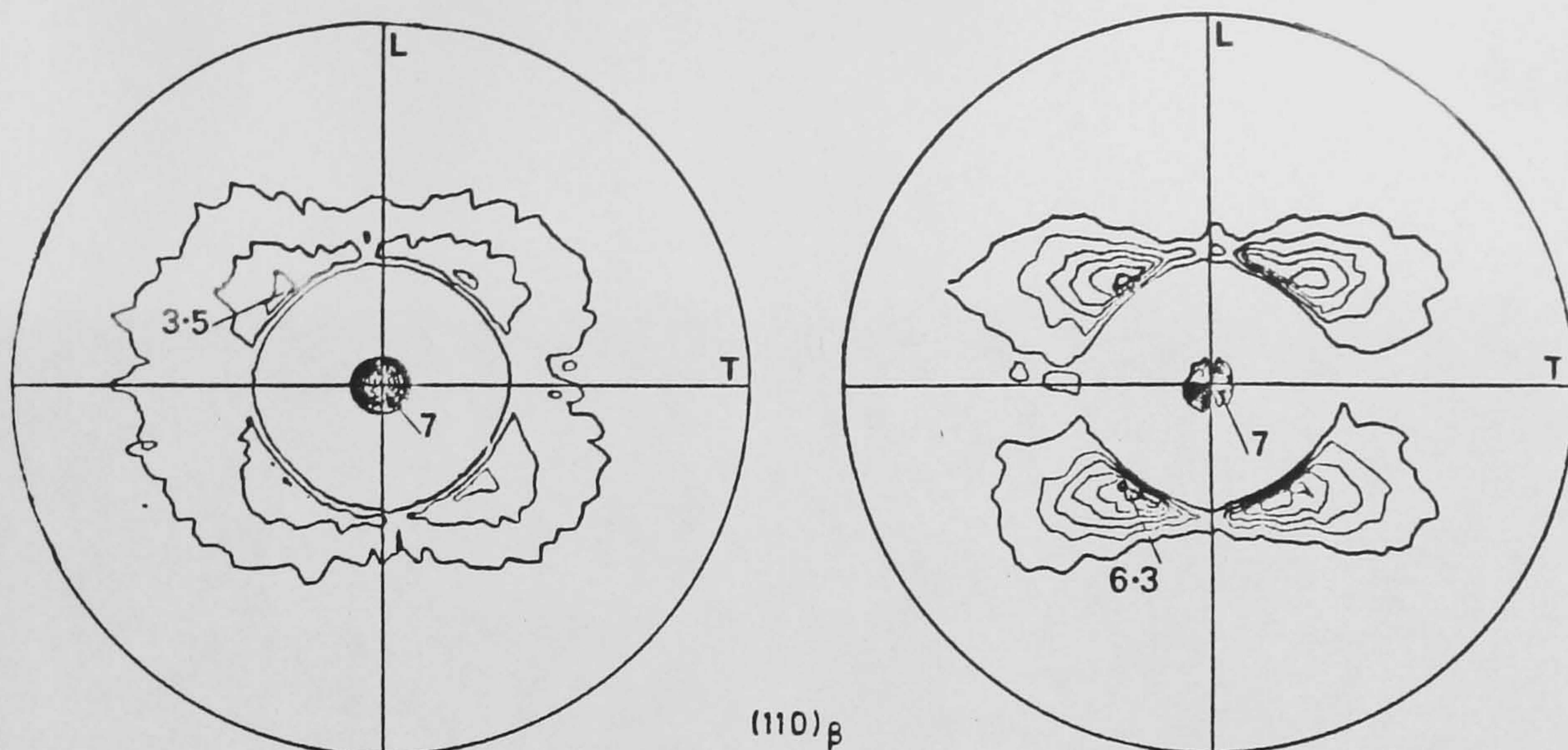


Fig 33 Ti-15V-3Cr-3Al-3Sn sheet test pieces

a) Before testing

b) After 200% strain at 910 °C and $3 \times 10^{-4} \text{ s}^{-1}$



a) As-received

b) After 238% strain at 810 °C
and $3 \times 10^{-4} \text{ s}^{-1}$

Fig 32 Texture of Ti-15V-3Cr-3Al-3Sn sheet.

Contour intervals 1 x random

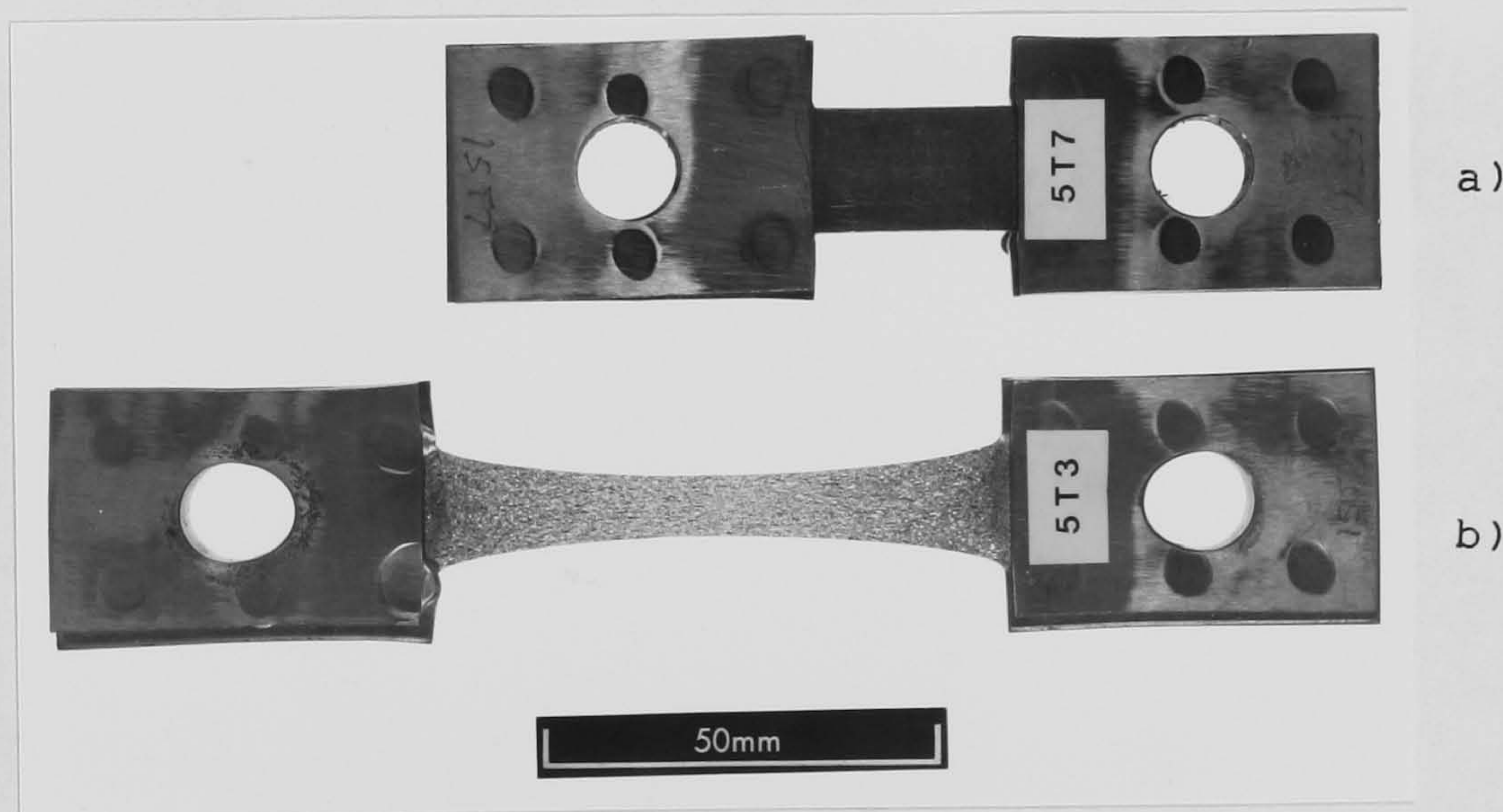
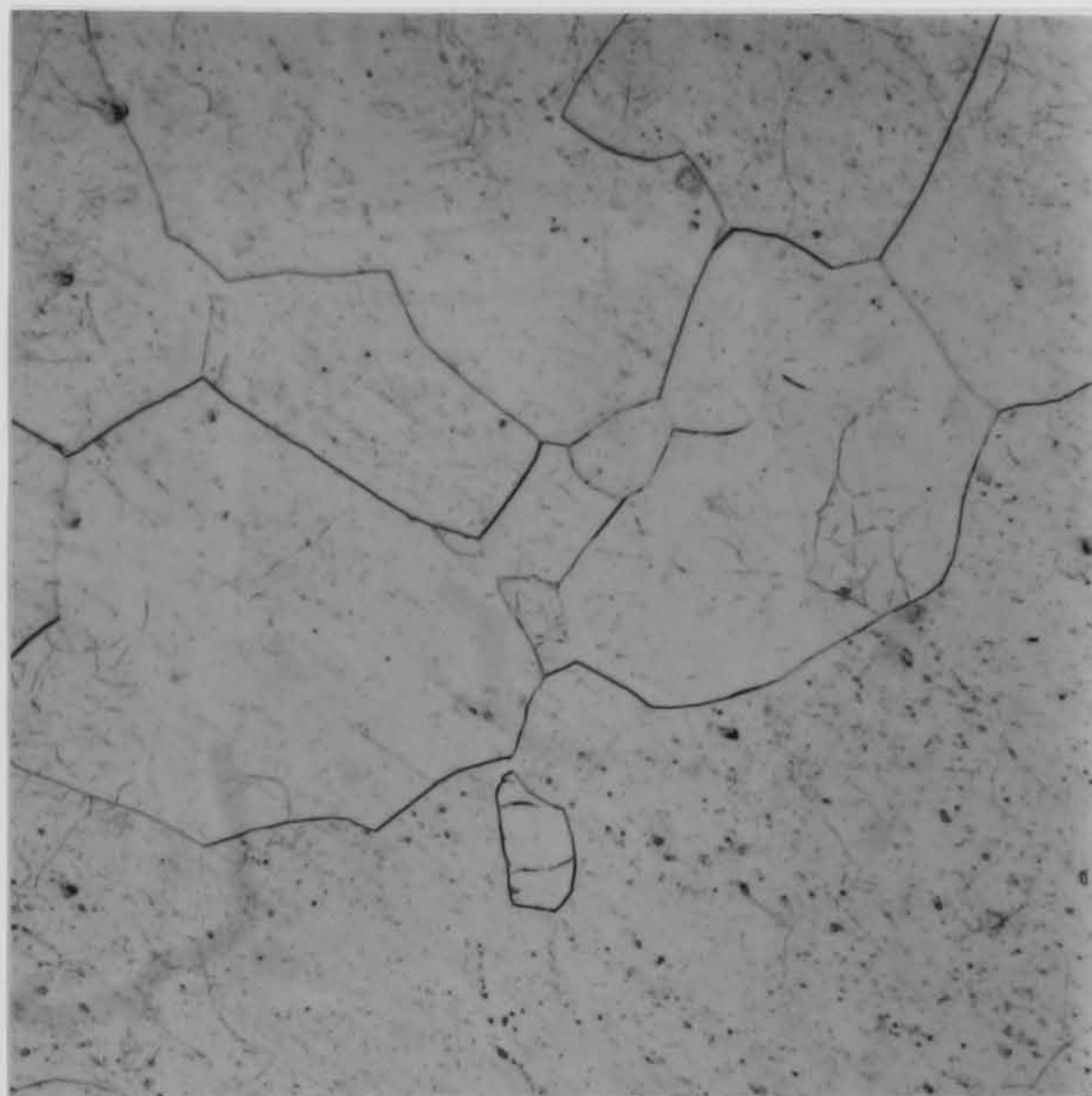


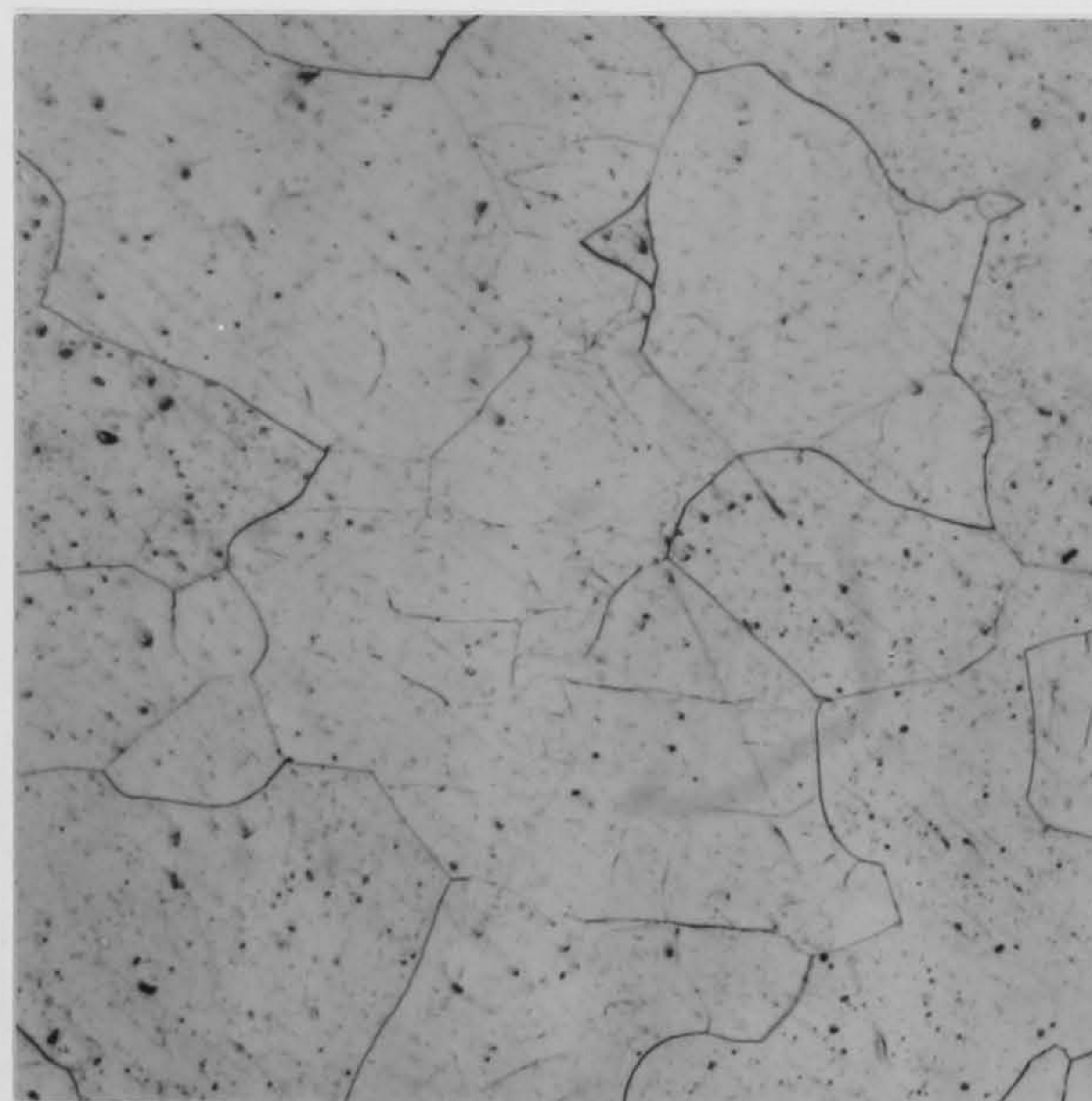
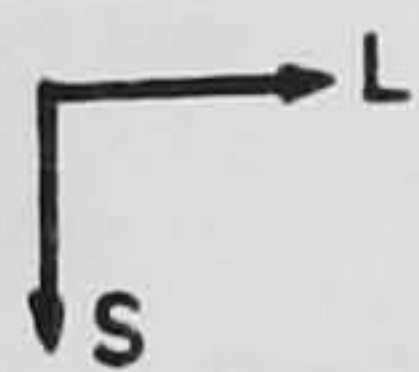
Fig 33 Ti-15V-3Cr-3Al-3Sn sheet test pieces

a) Before testing

b) After 200% strain at 910 °C and $3 \times 10^{-4} \text{ s}^{-1}$

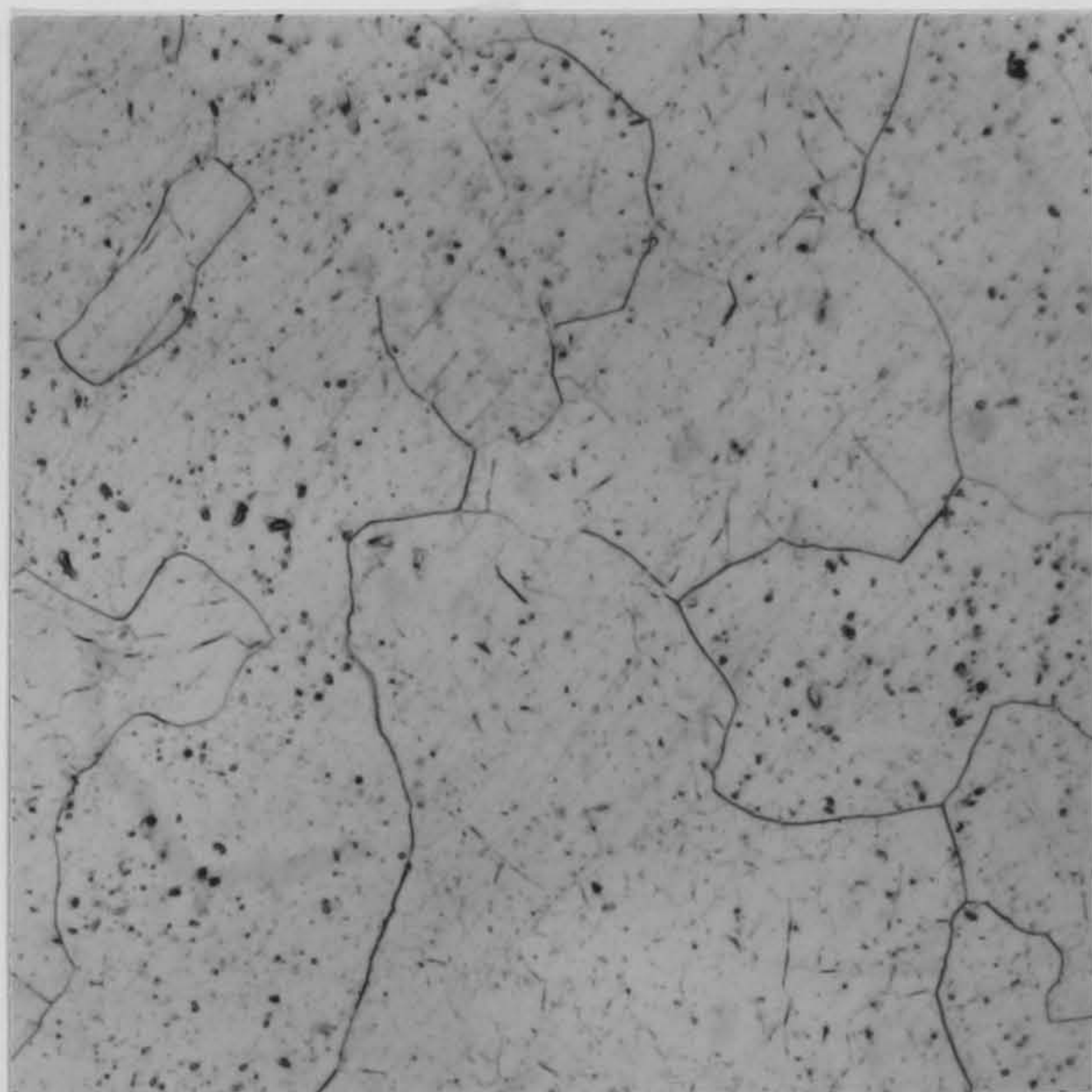


a) Annealed at 810 °C
(unstrained)

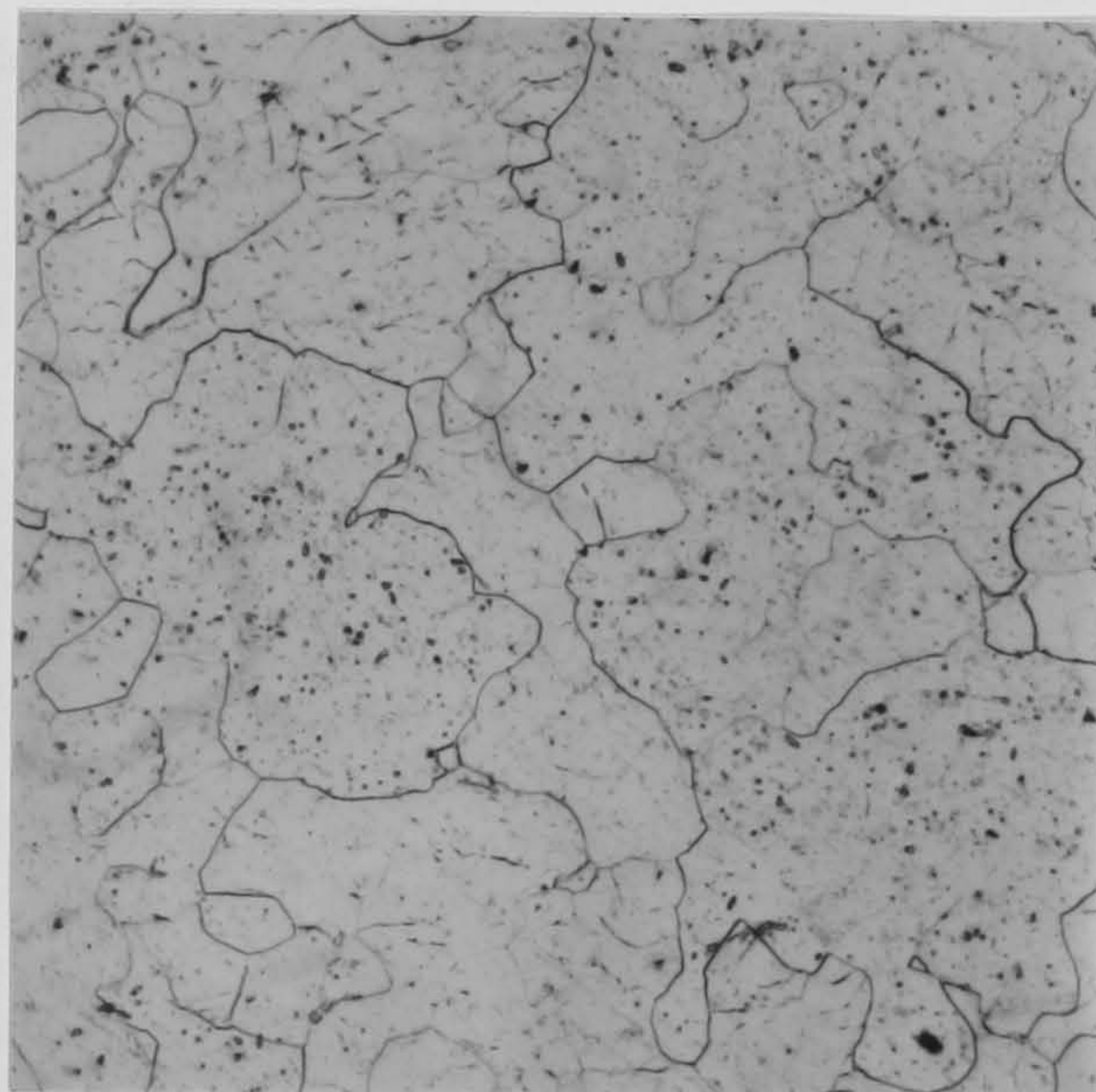


b) 15% strain

40 μm

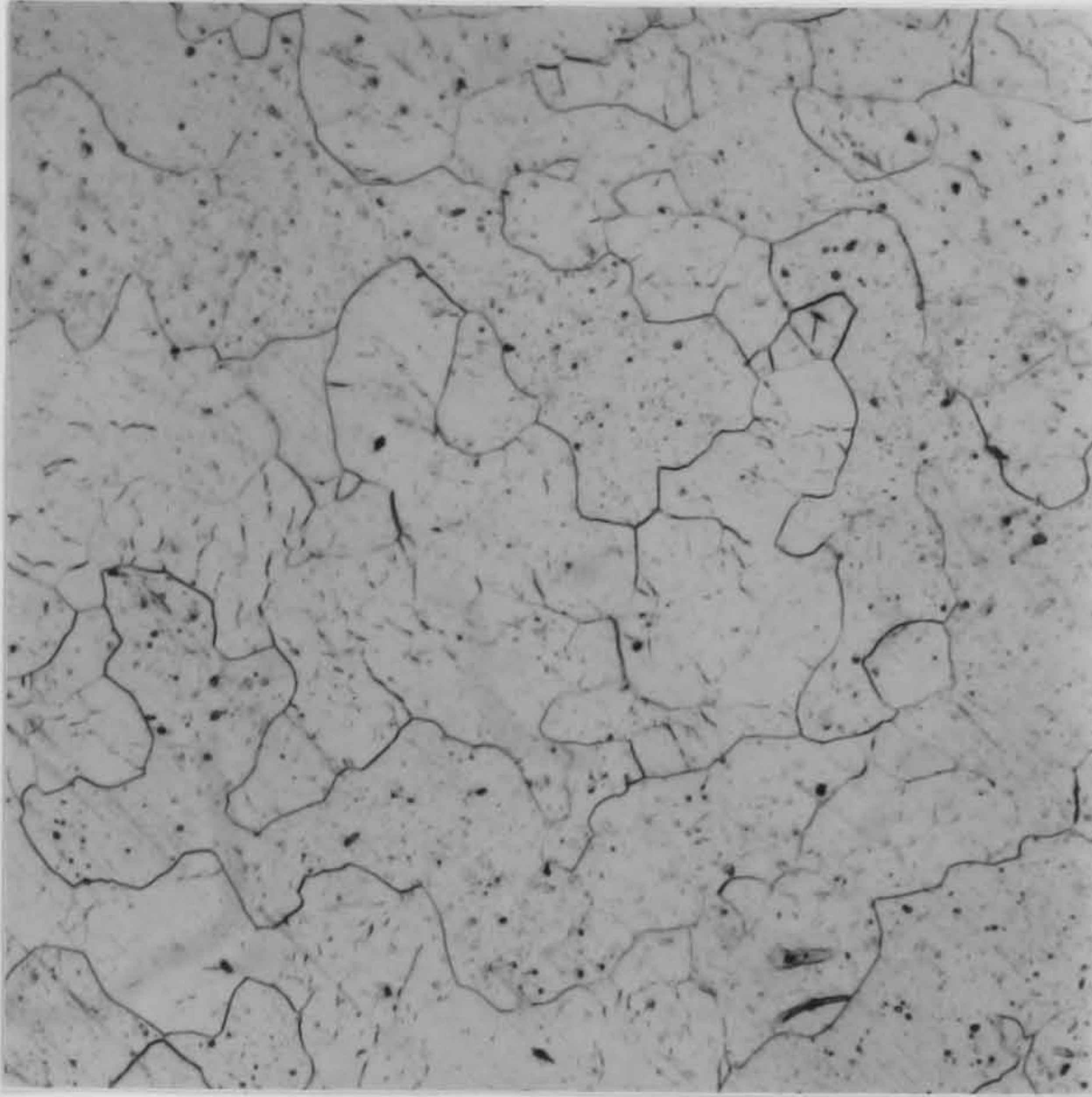


c) 46% strain

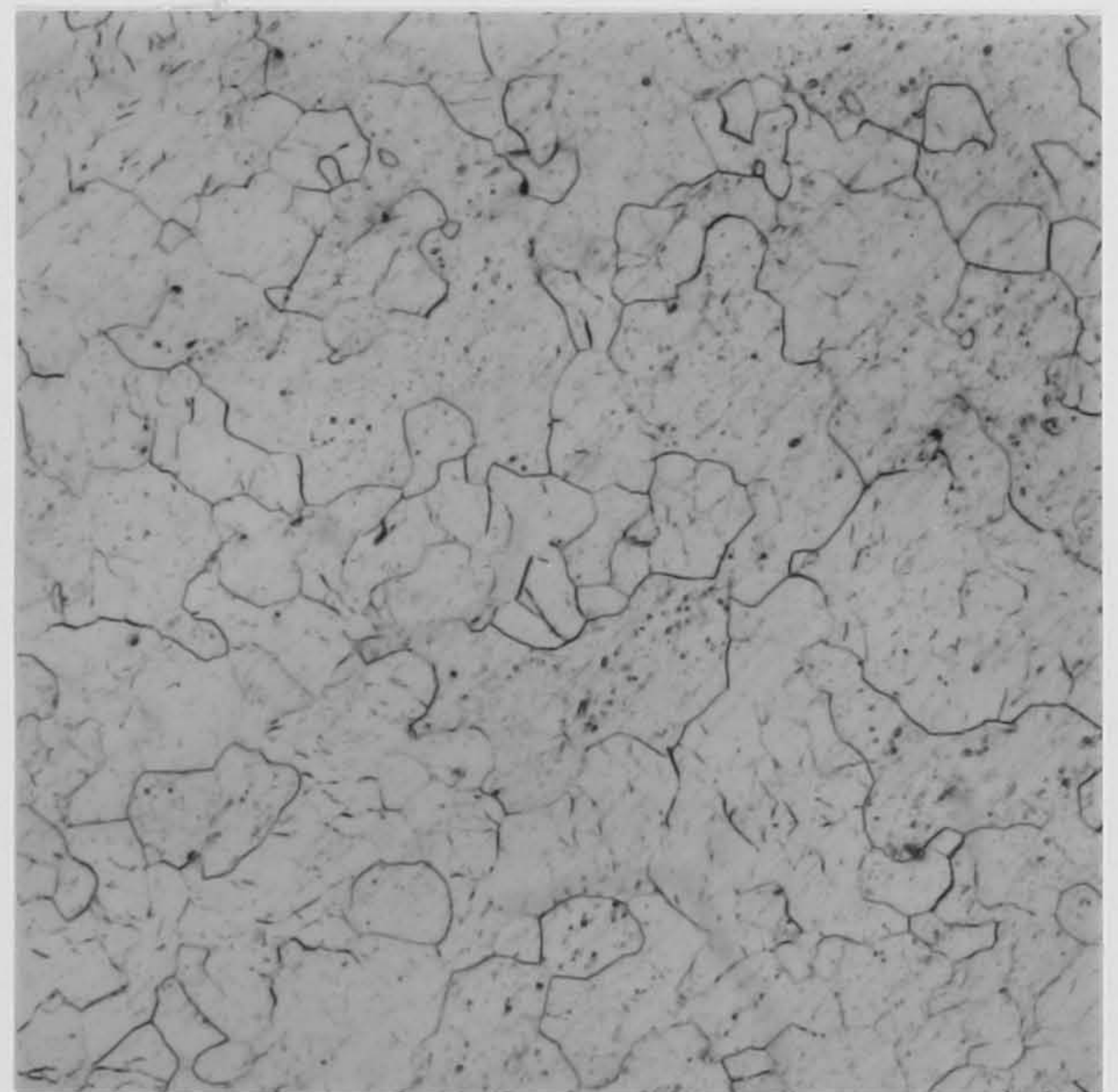


d) 197% strain

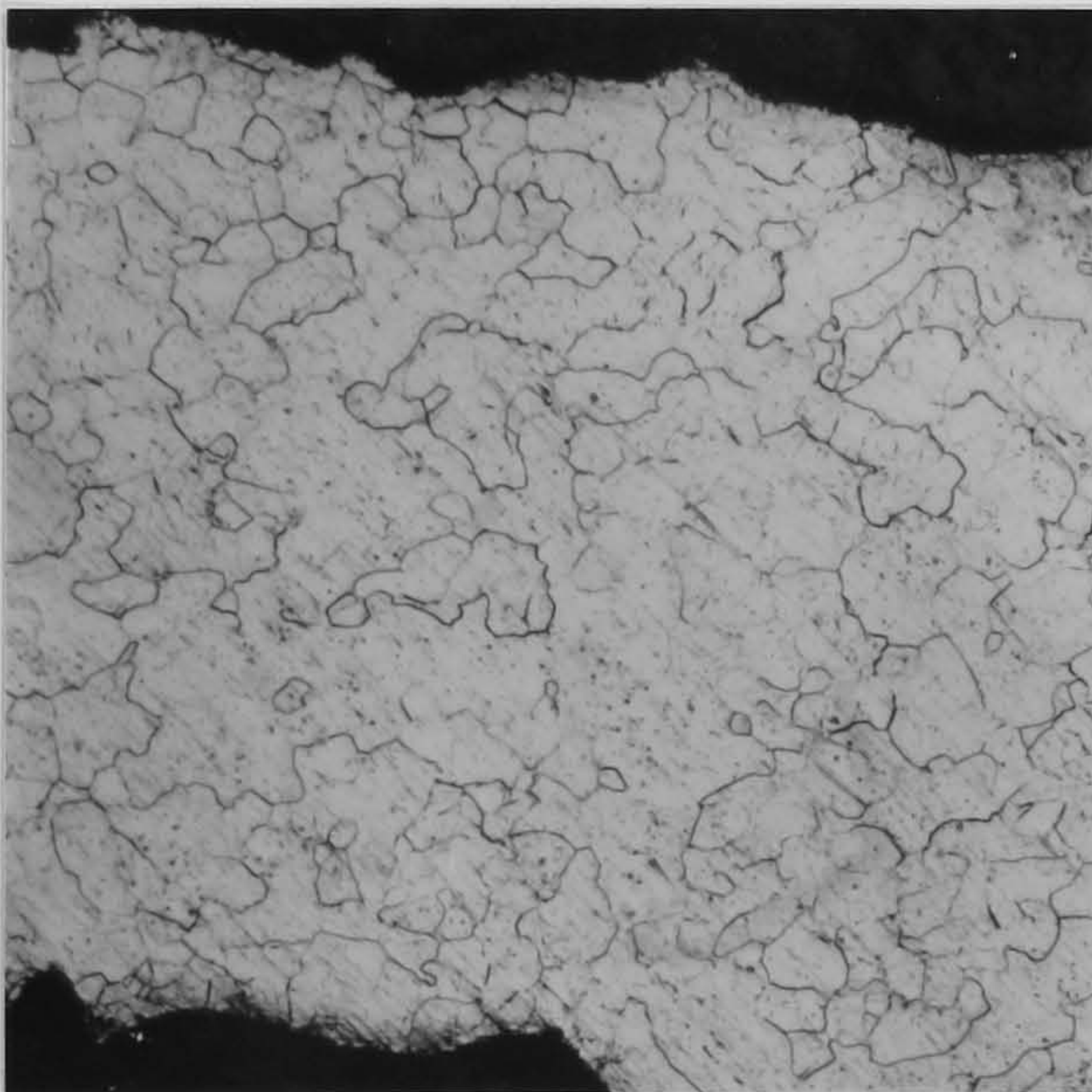
Fig 34 Effect of strain at 810 °C and $3 \times 10^{-4} \text{s}^{-1}$
on the microstructure of Ti-15V-3Cr-3Al-3Sn sheet



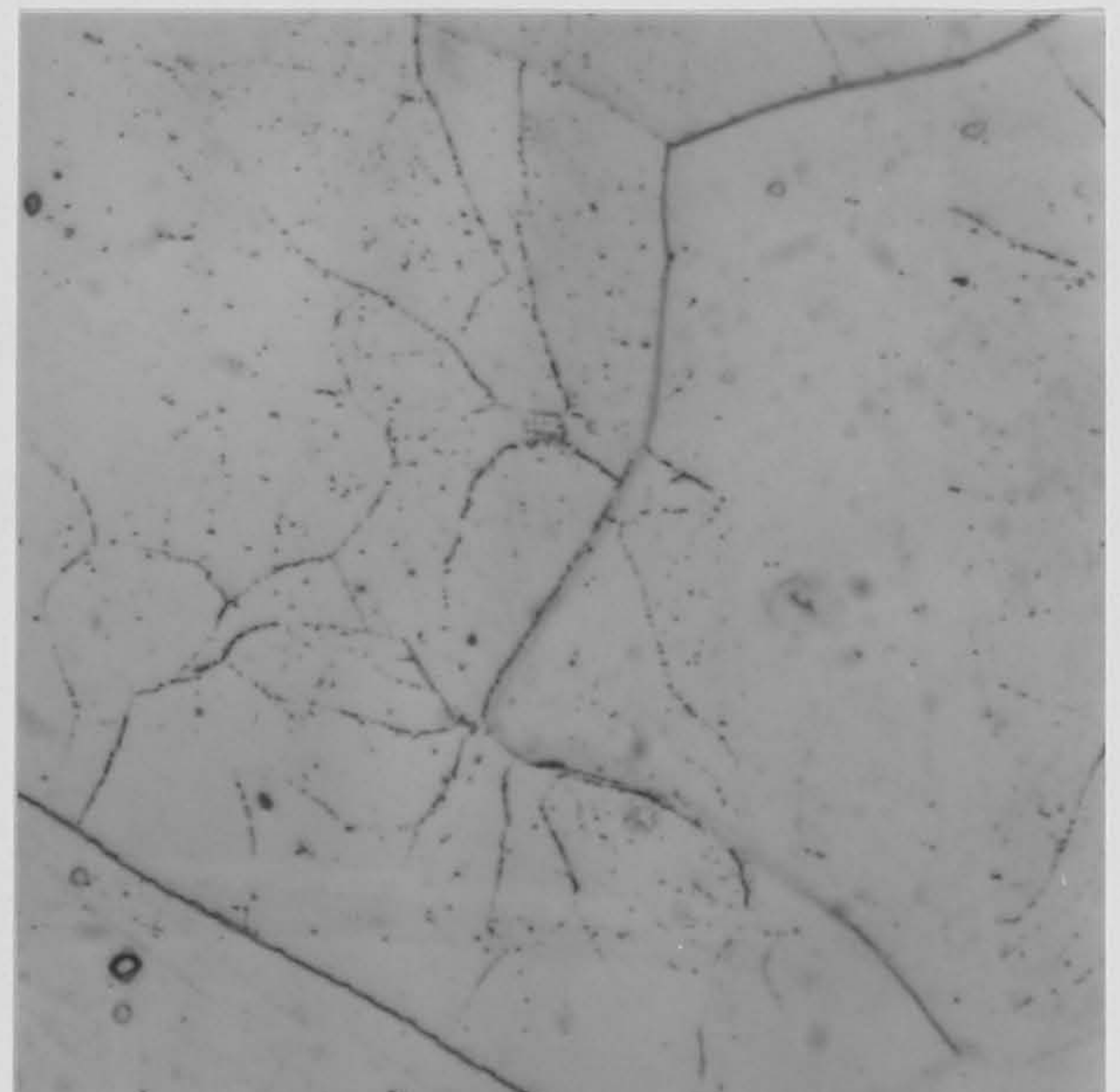
e) 373% strain



f) 634% strain



g) 1652% strain
(necked region)



h) Precipitation on subgrain boundaries

20 μm

Fig 34 Effect of strain at 810 °C and $3 \times 10^{-4} \text{ s}^{-1}$
on the microstructure of Ti-15V-3Cr-3Al-3Sn sheet.
Magnification 160x except h): (630x)

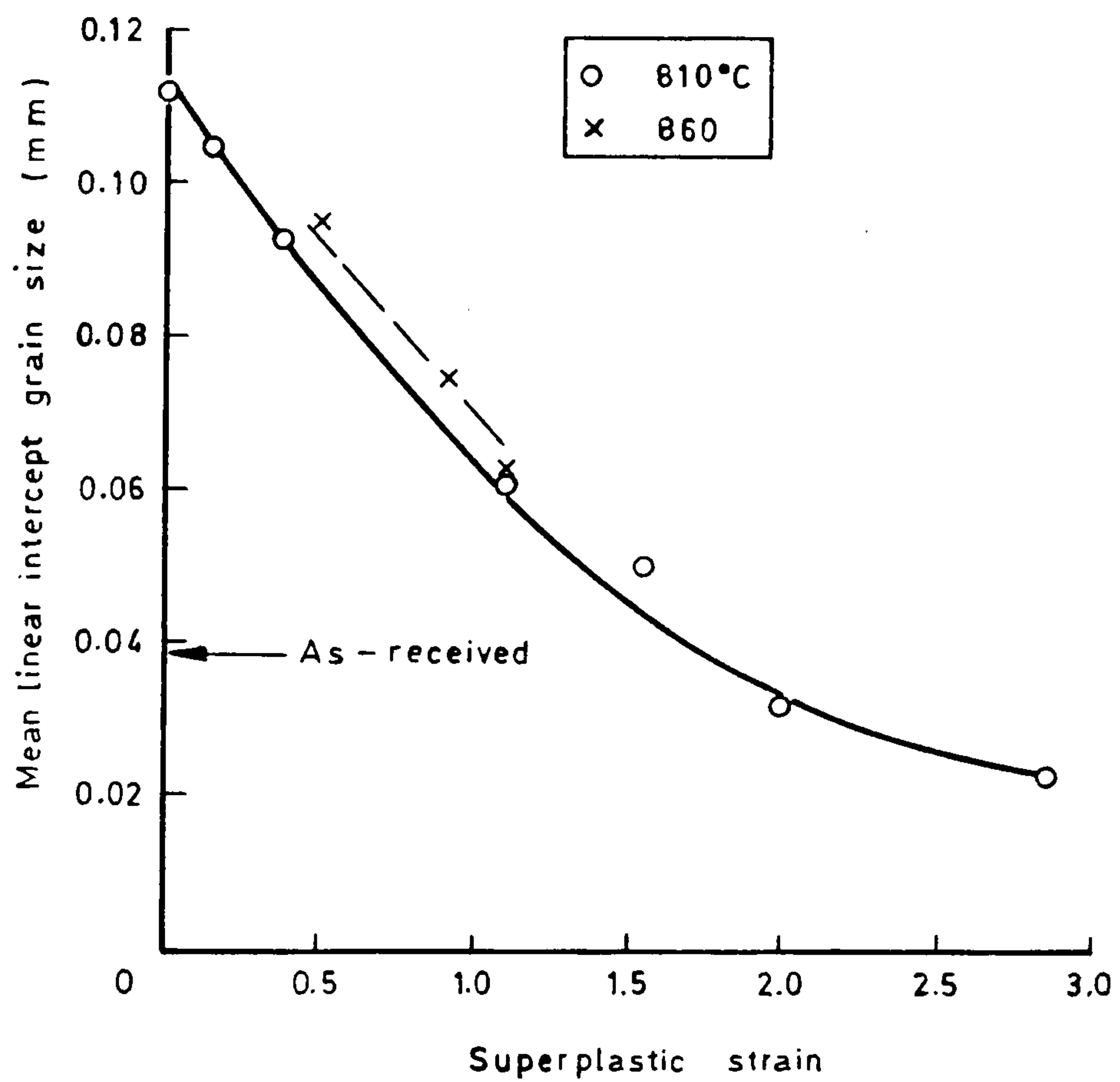


Fig 35 Effect of strain at 810 and 860 °C and $3 \times 10^{-4} \text{ s}^{-1}$ on grain size (measured in the L direction) of Ti-15V-3Cr-3Al-3Sn sheet

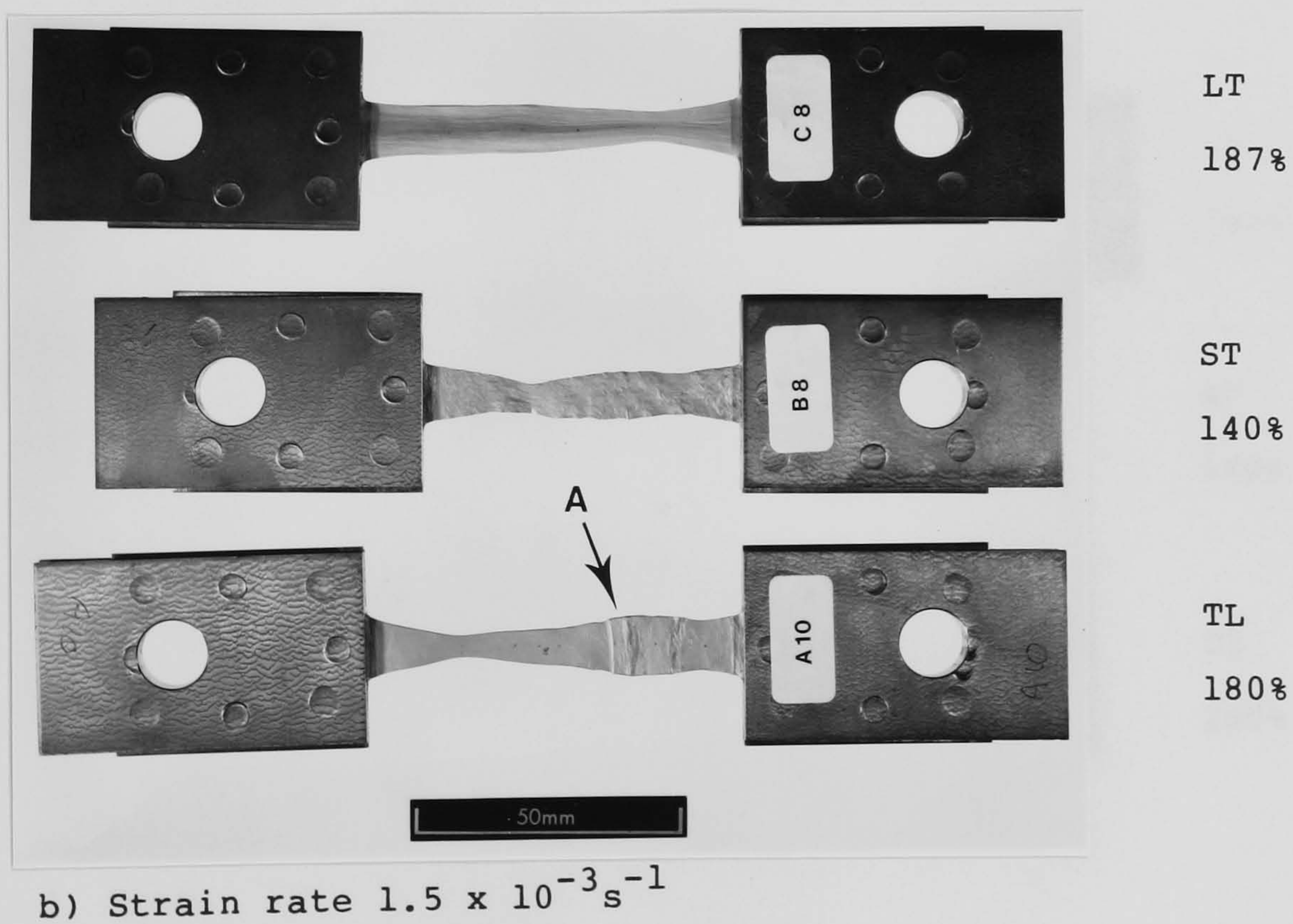
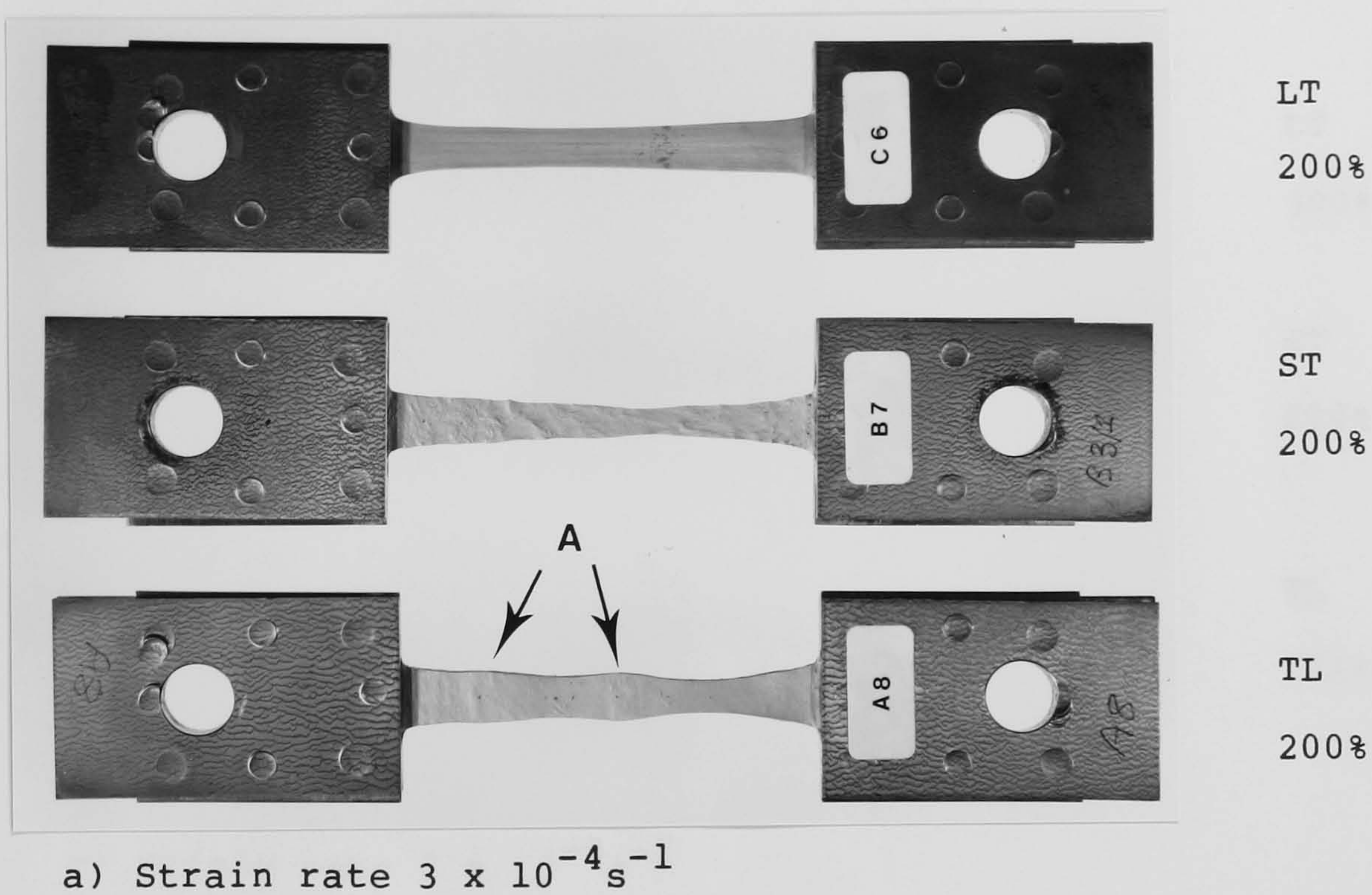
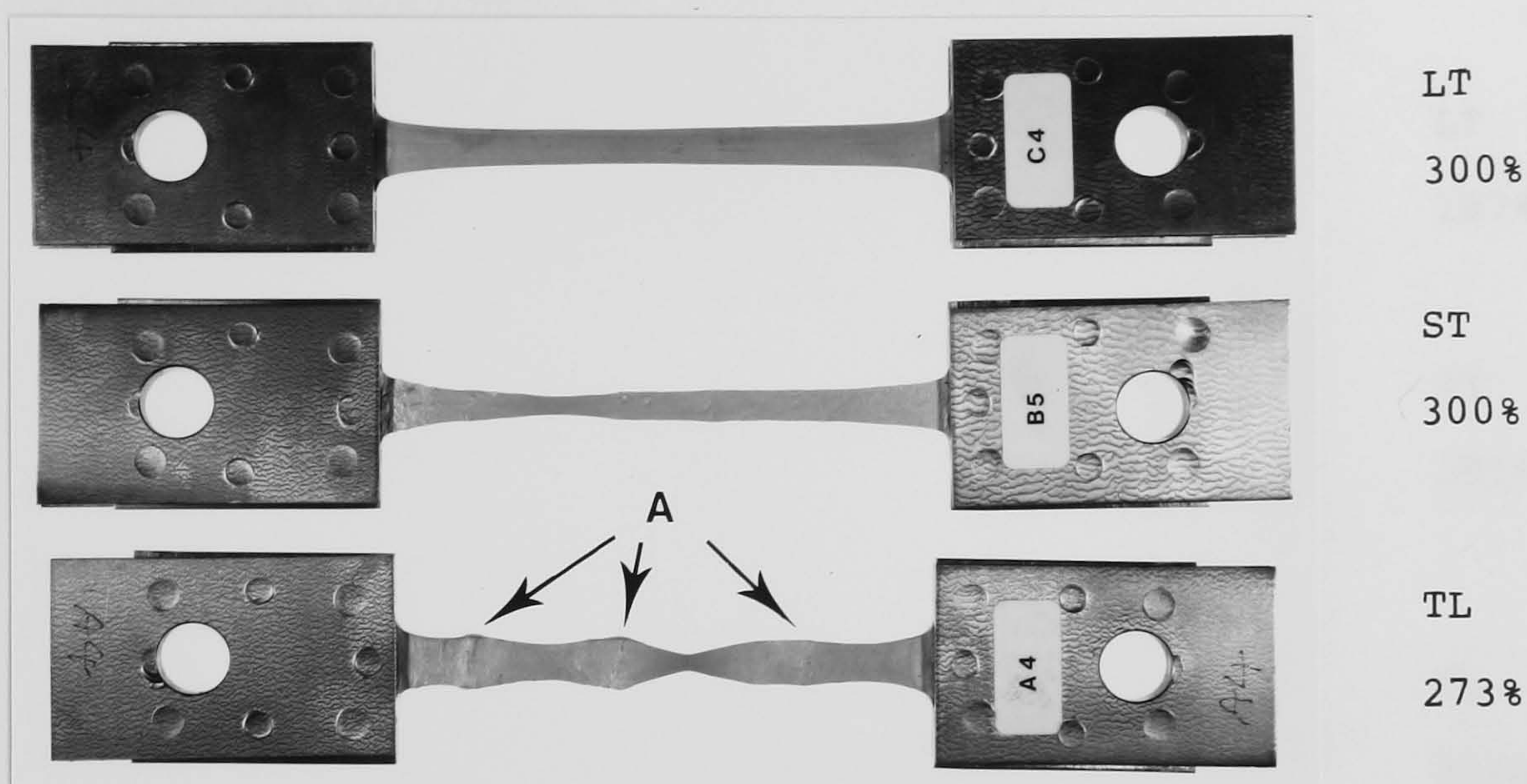
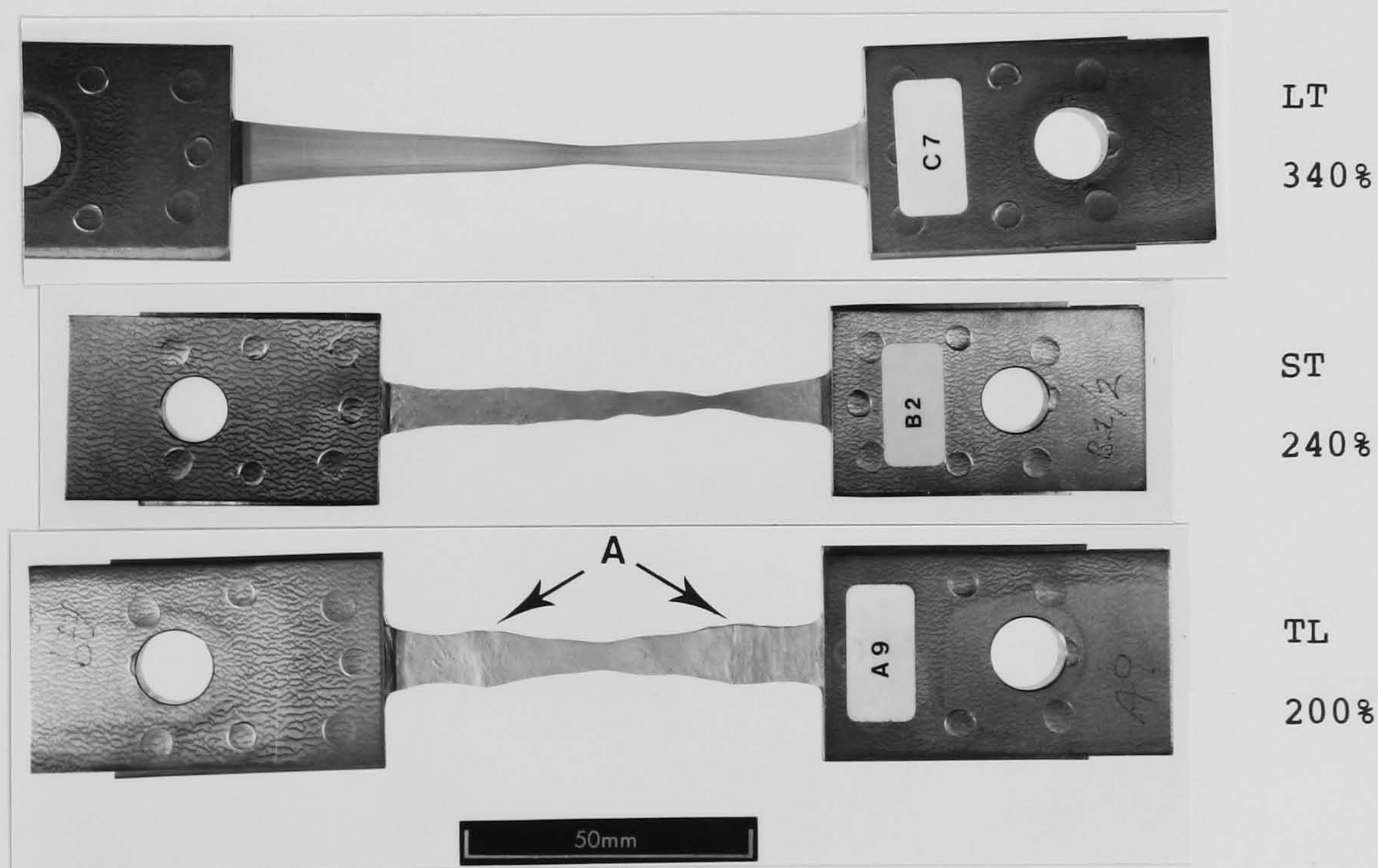


Fig 36 Ti-6Al-4V sheet test pieces machined from bar. Test piece shape after strain at 800 °C. Magnification 0.6x

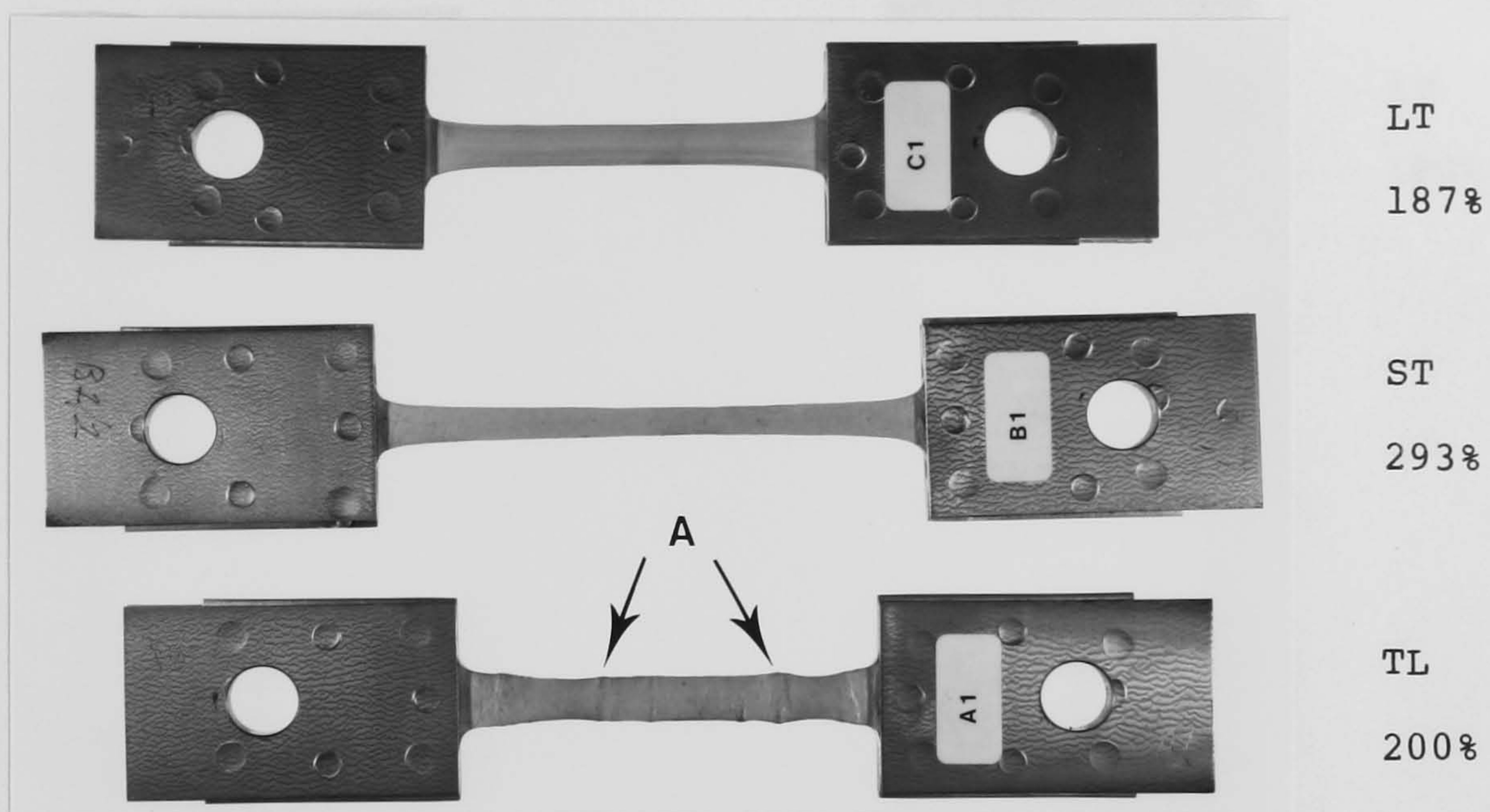


c) Strain rate $3 \times 10^{-4} \text{ s}^{-1}$

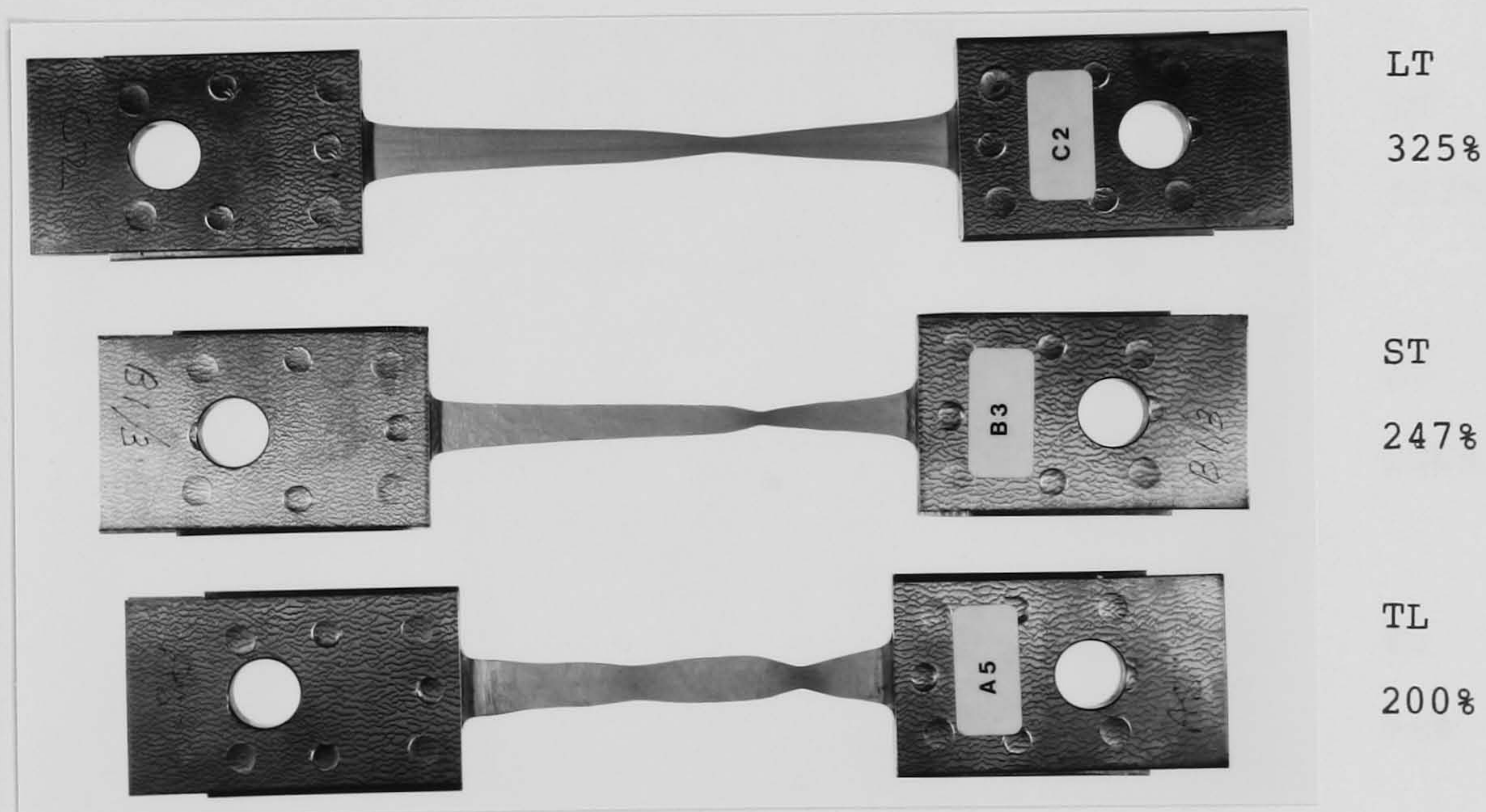


d) Strain rate $1.5 \times 10^{-3} \text{ s}^{-1}$

Fig 36 Ti-6Al-4V sheet test pieces machined from bar. Test piece shape after strain at 875 °C. Magnification 0.6x

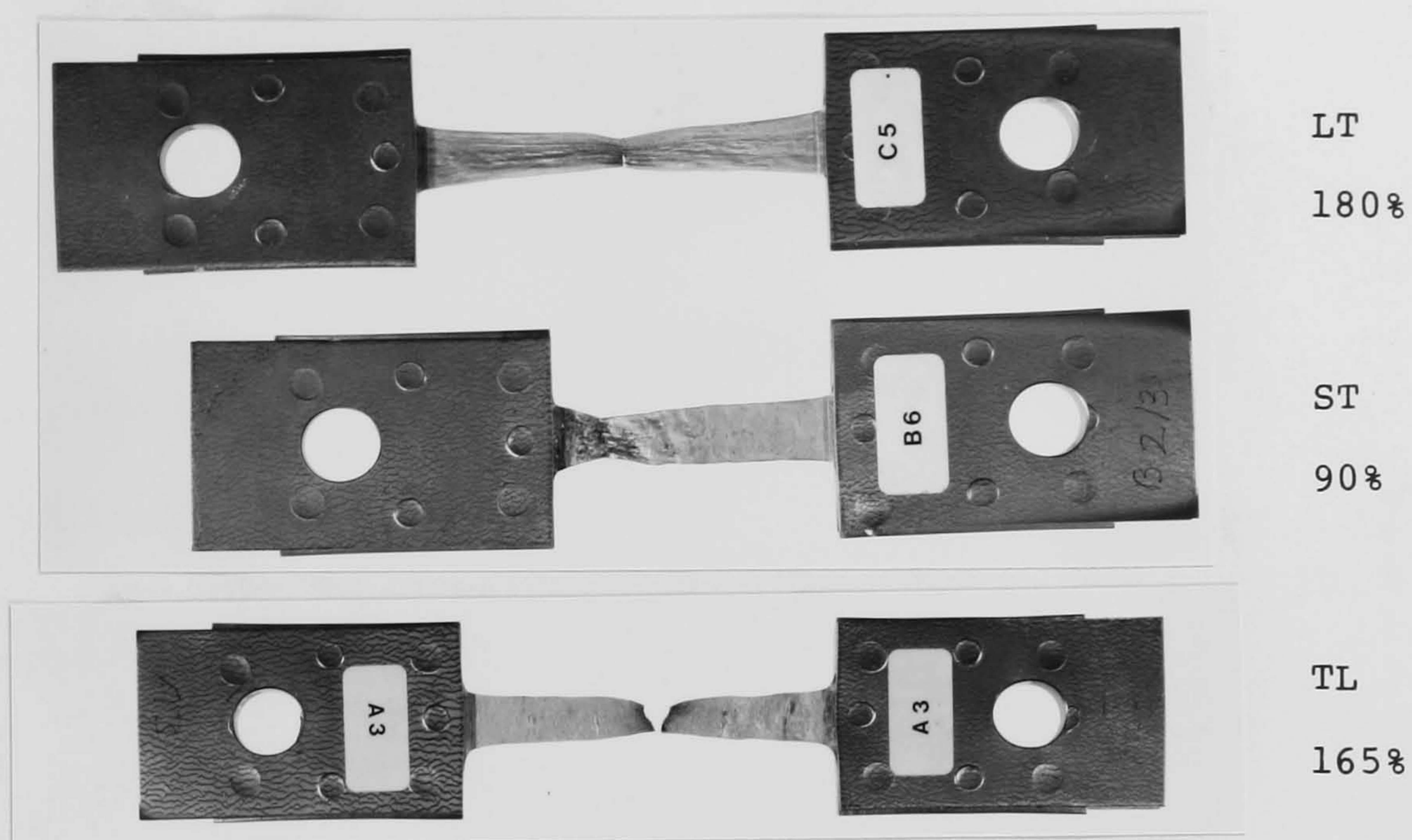


e) Strain rate $3 \times 10^{-4} \text{ s}^{-1}$

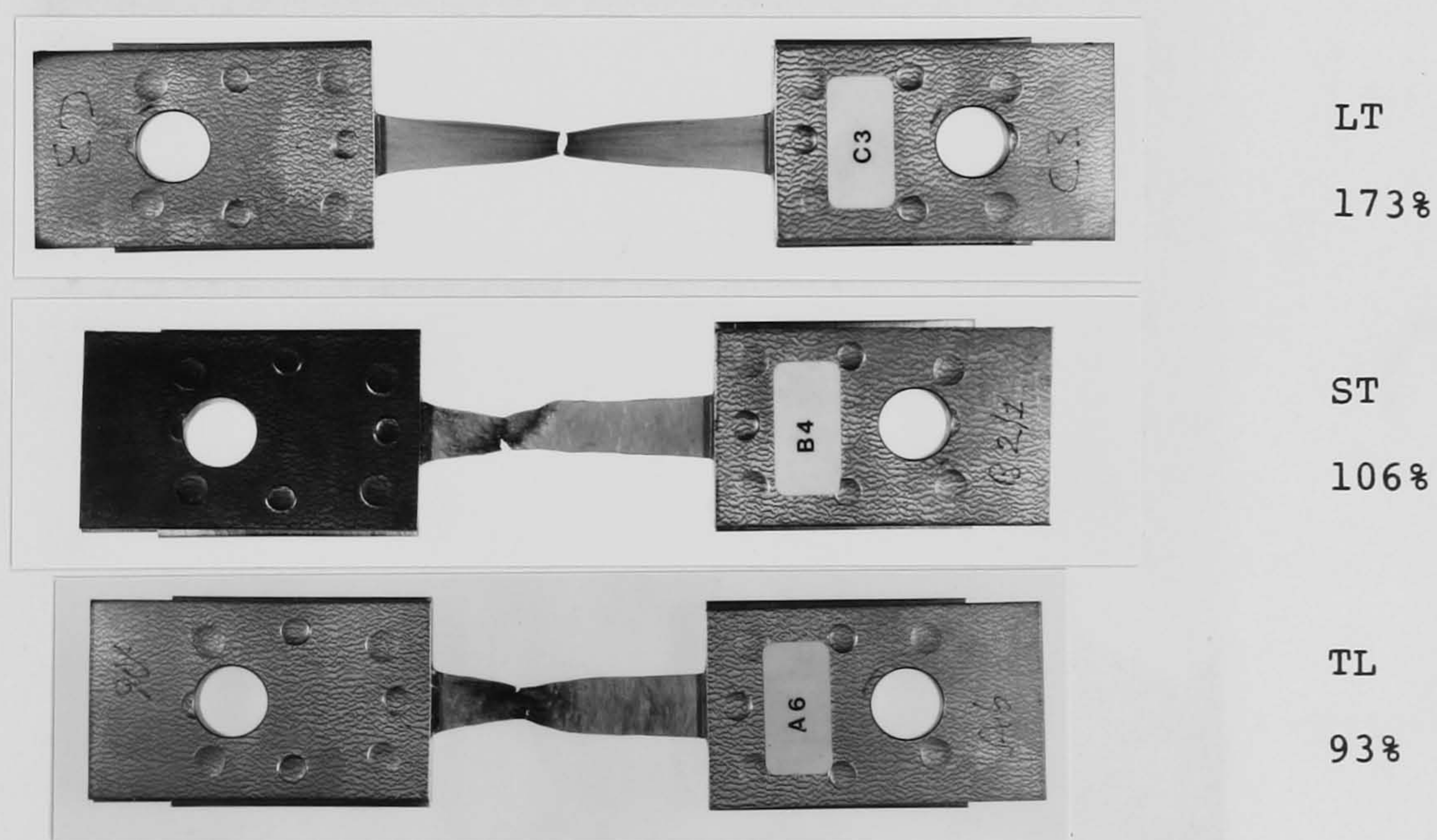


f) Strain rate $1.5 \times 10^{-3} \text{ s}^{-1}$

Fig 36 Ti-6Al-4V sheet test pieces machined from bar. Test piece shape after strain at 925 °C. Magnification 0.6x

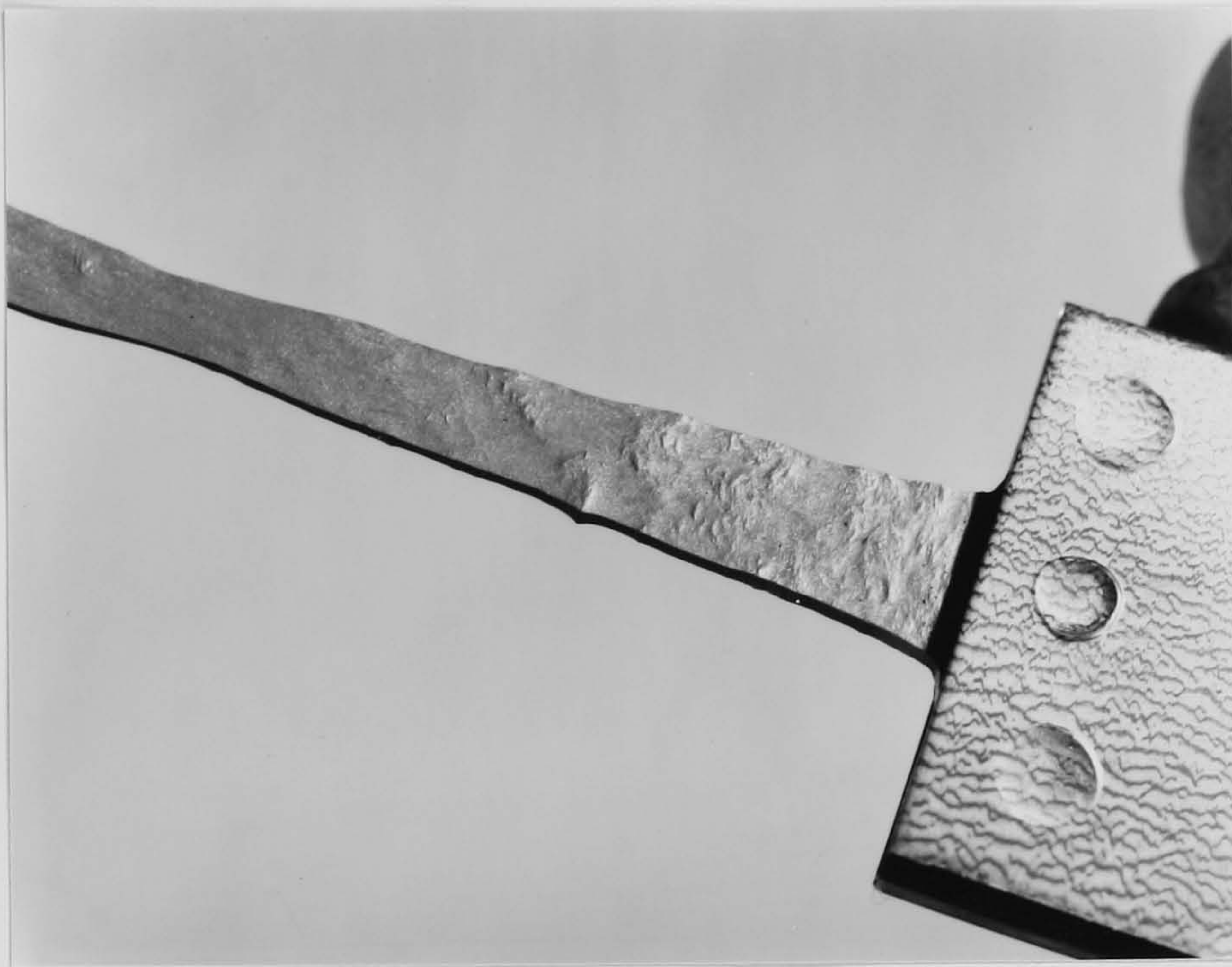


g) Strain rate $3 \times 10^{-4} \text{ s}^{-1}$



h) Strain rate $1.5 \times 10^{-3} \text{ s}^{-1}$

Fig 36 Ti-6Al-4V sheet test pieces machined from bar. Test piece shape after strain at 975 °C. Magnification 0.6x



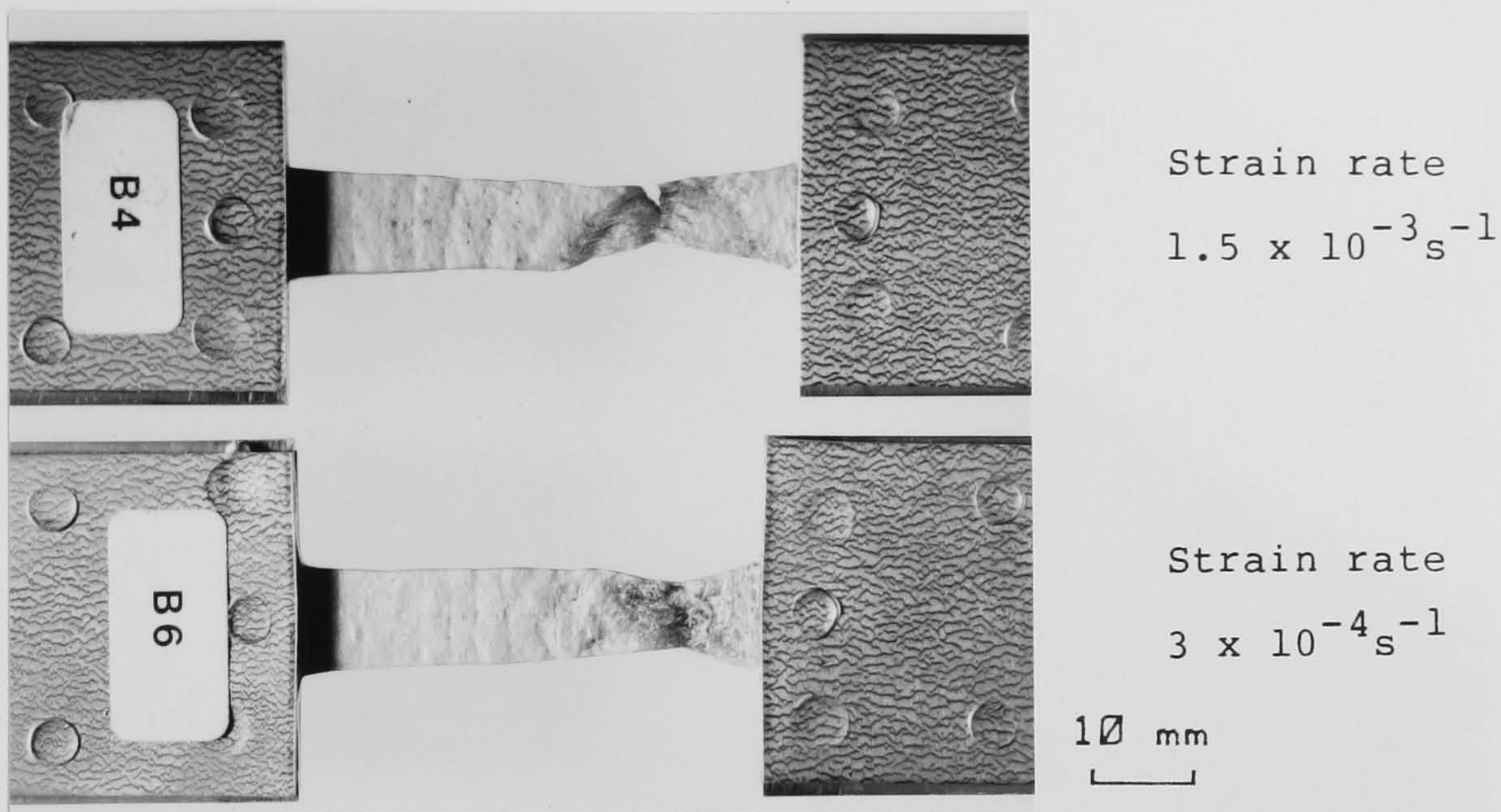
a) ST orientation test piece B7

5 mm
└───┘

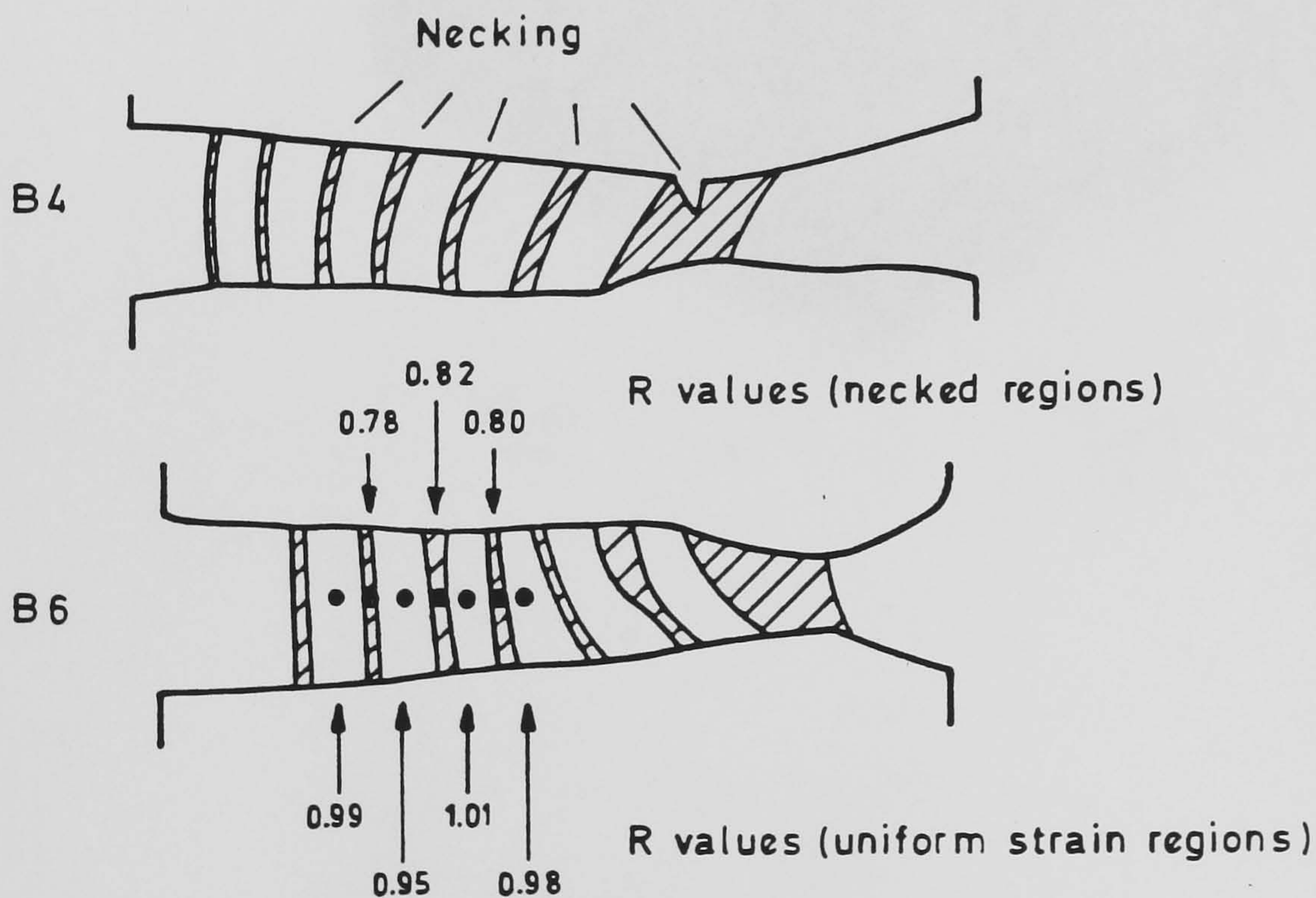


b) LT orientation test piece C6

Fig 37 Effect of superplastic strain at 800°C and $3 \times 10^{-4} \text{s}^{-1}$
on Ti-6Al-4V sheet test pieces machined from bar



a) ST orientation test pieces B4 and B6



b) Position of necks and R value measurements

Fig 38 Necking of ST orientation test pieces pulled at 975 °C

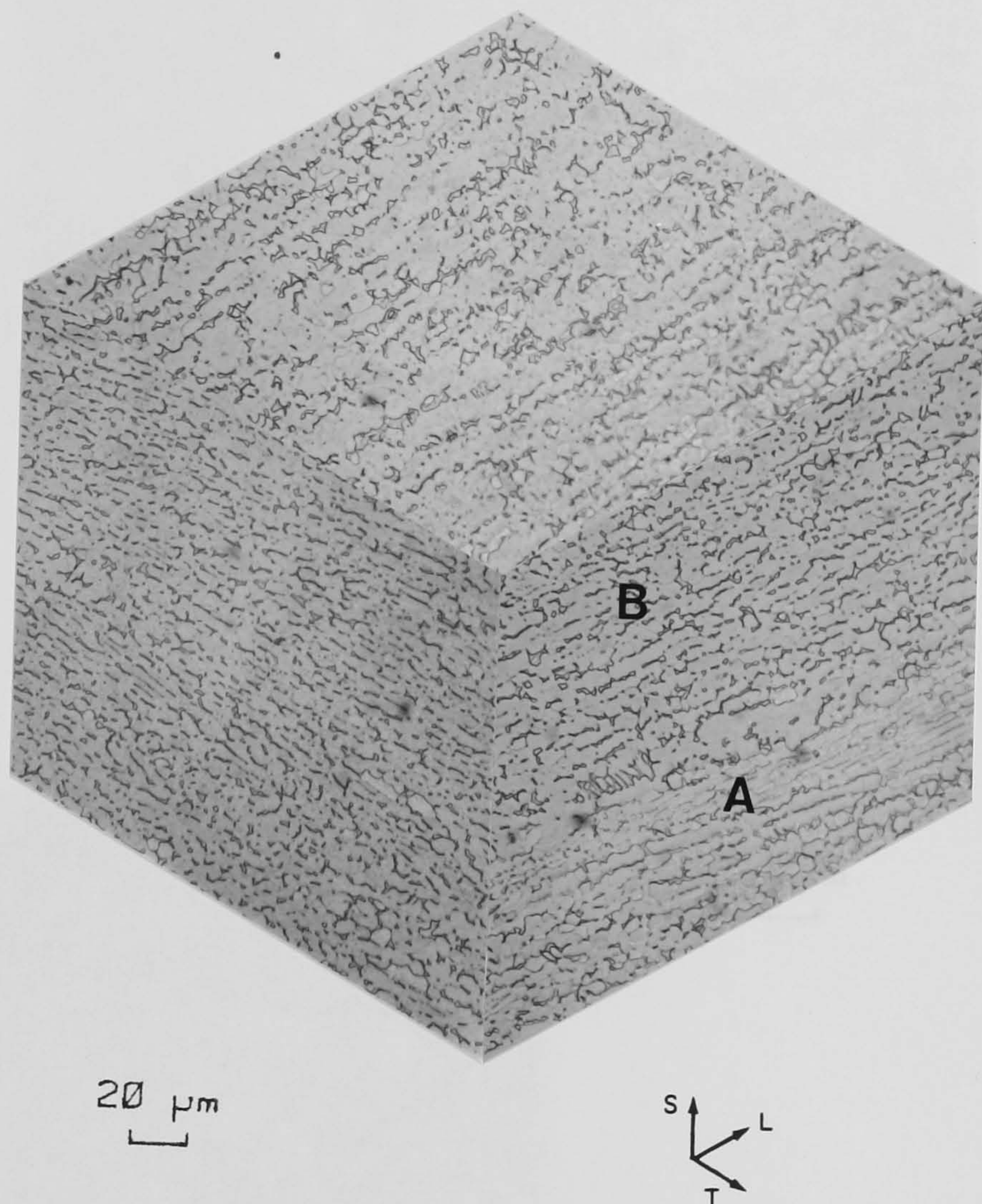
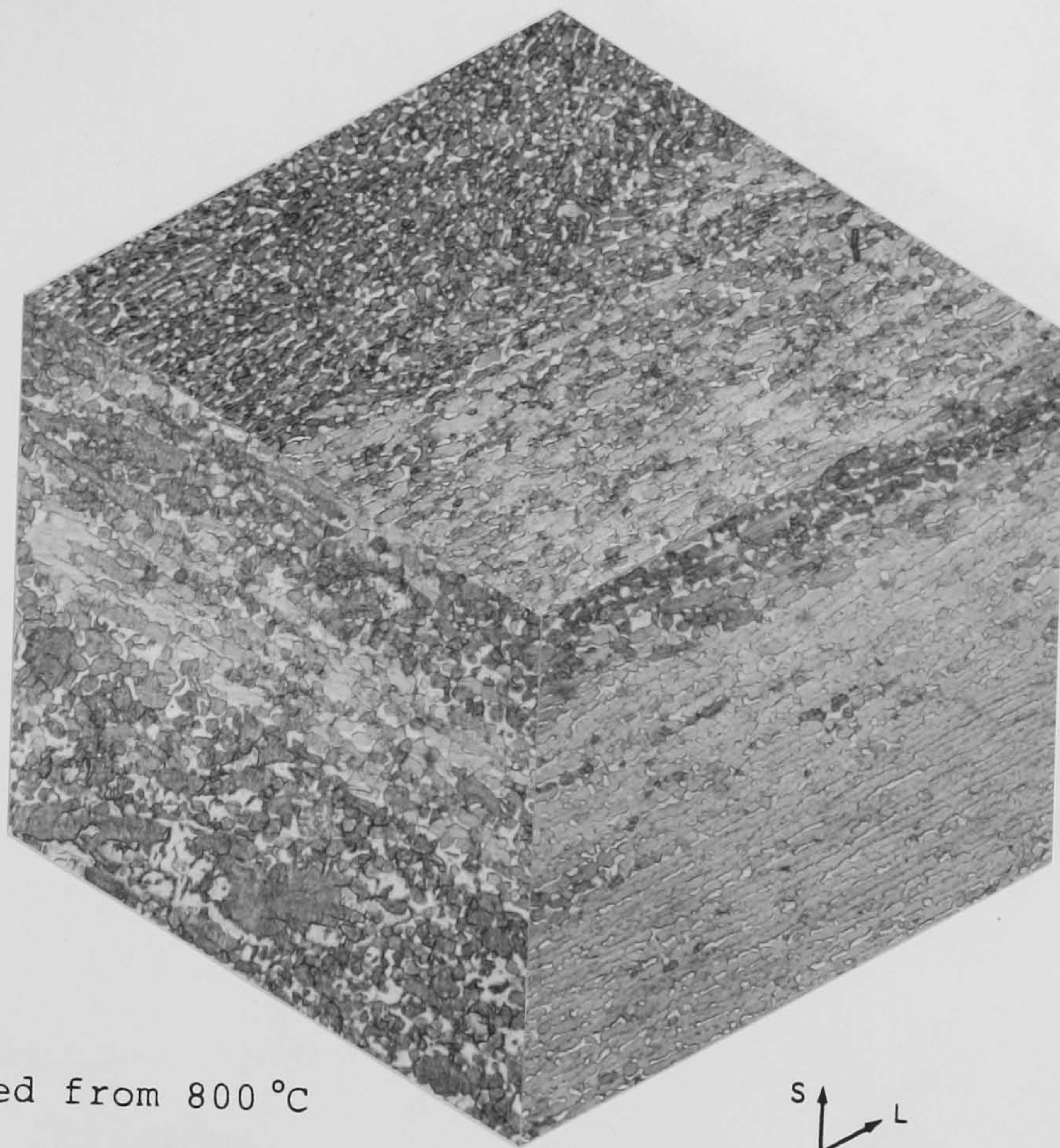
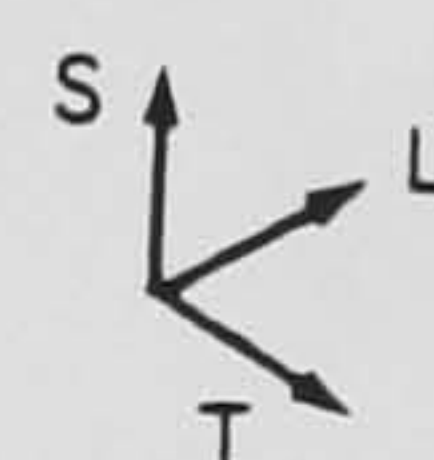


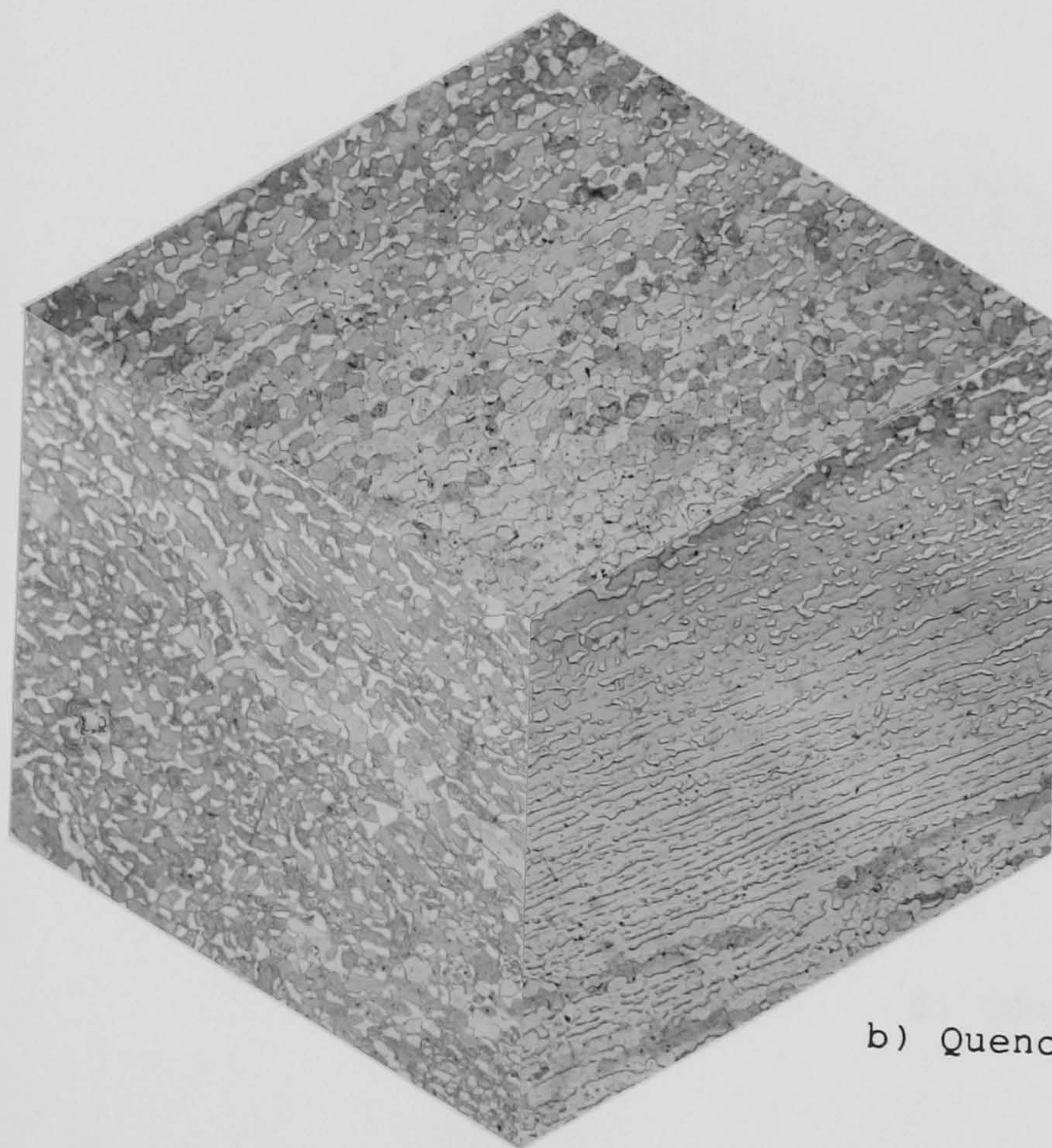
Fig 39 Mill annealed (as-received) Ti-6Al-4V bar showing aligned (A) and non-aligned (B) microstructure



a) Quenched from 800 °C



20 μm



b) Quenched from 875 °C

Fig 40 Microstructures of quenched Ti-6Al-4V bar

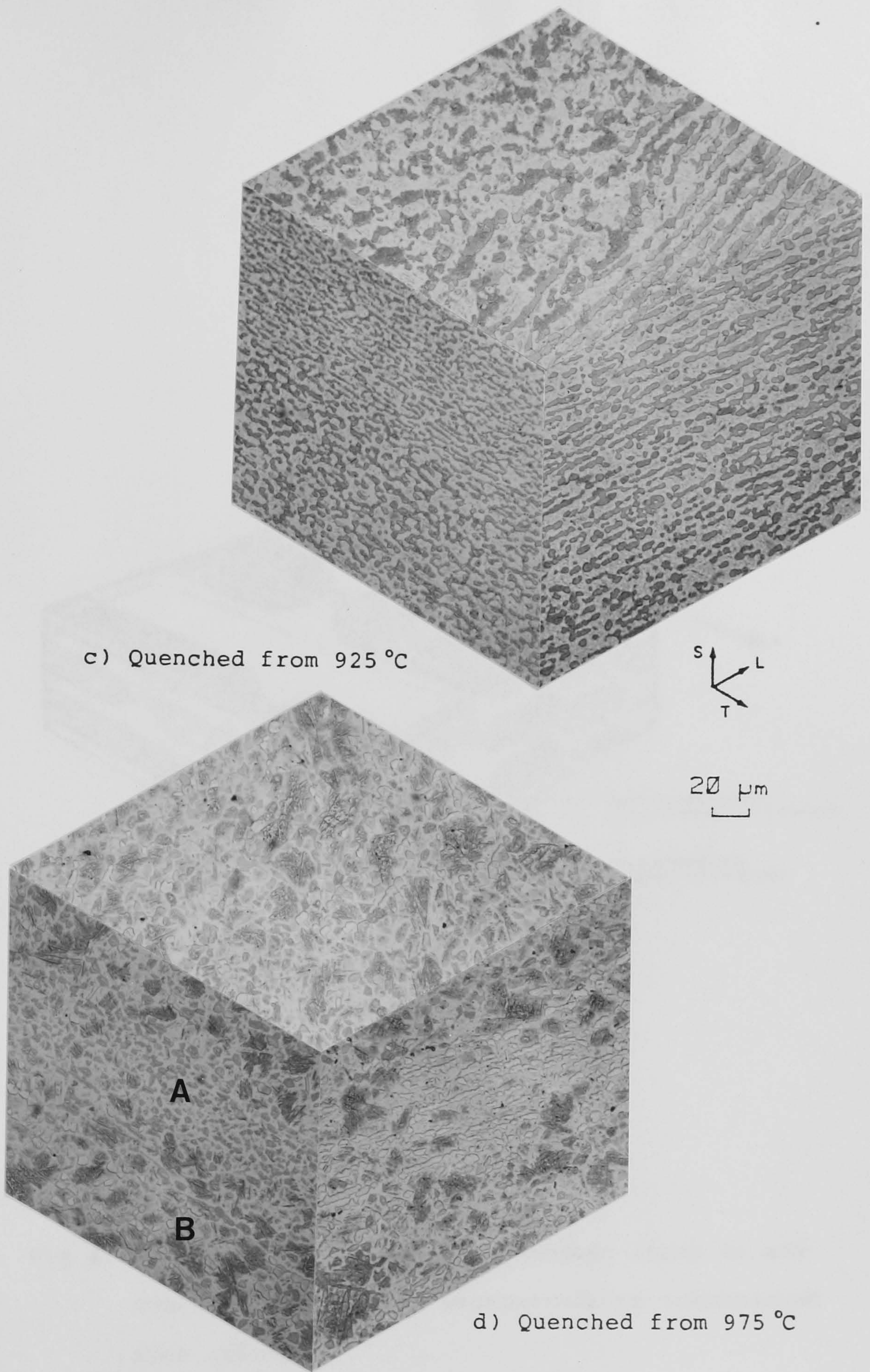


Fig 40 Microstructures of quenched Ti-6Al-4V bar

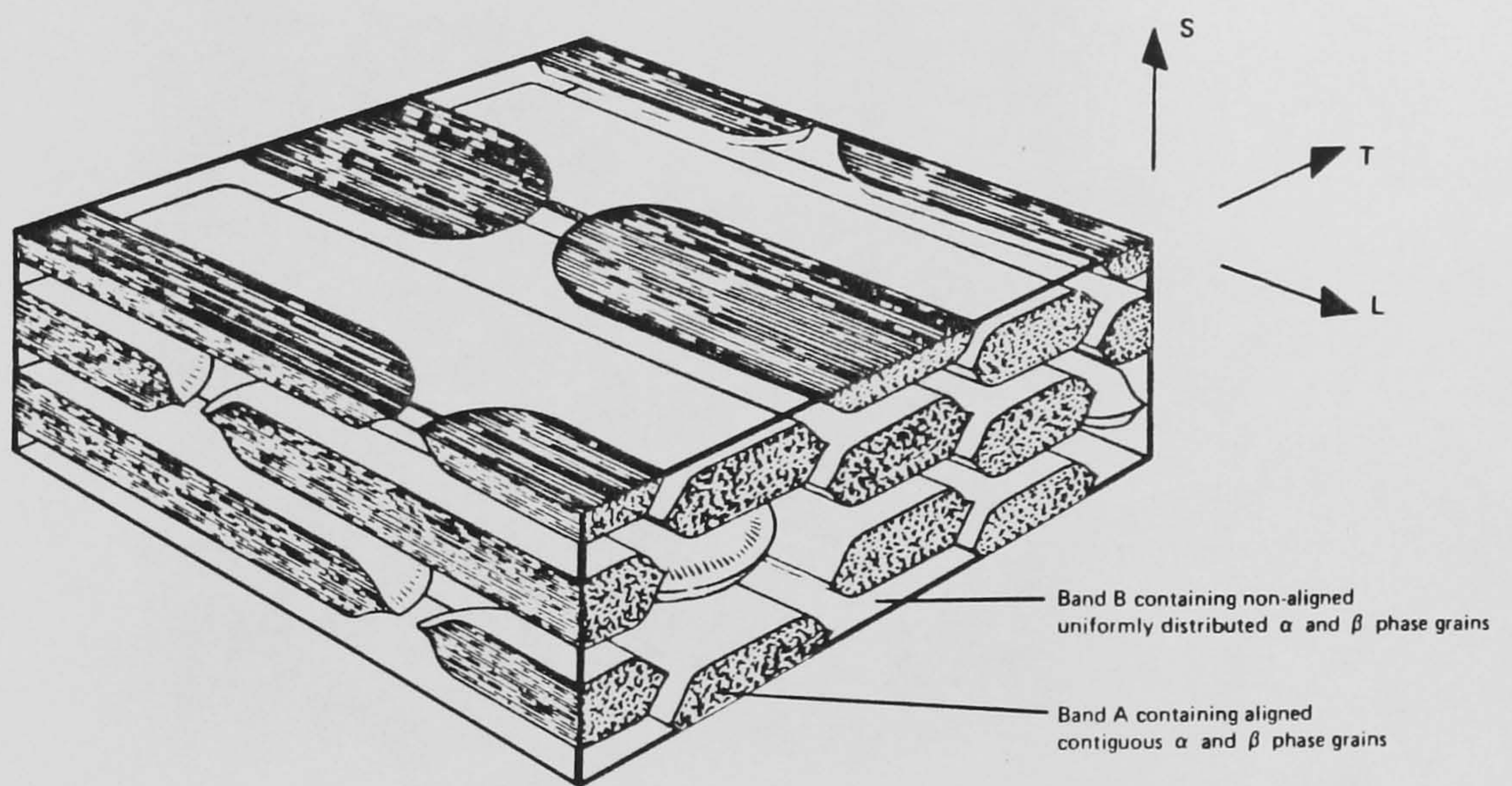


Fig 41 Schematic distribution of aligned (band A) and non-aligned (band B) microstructure (reproduced from ref. 51)

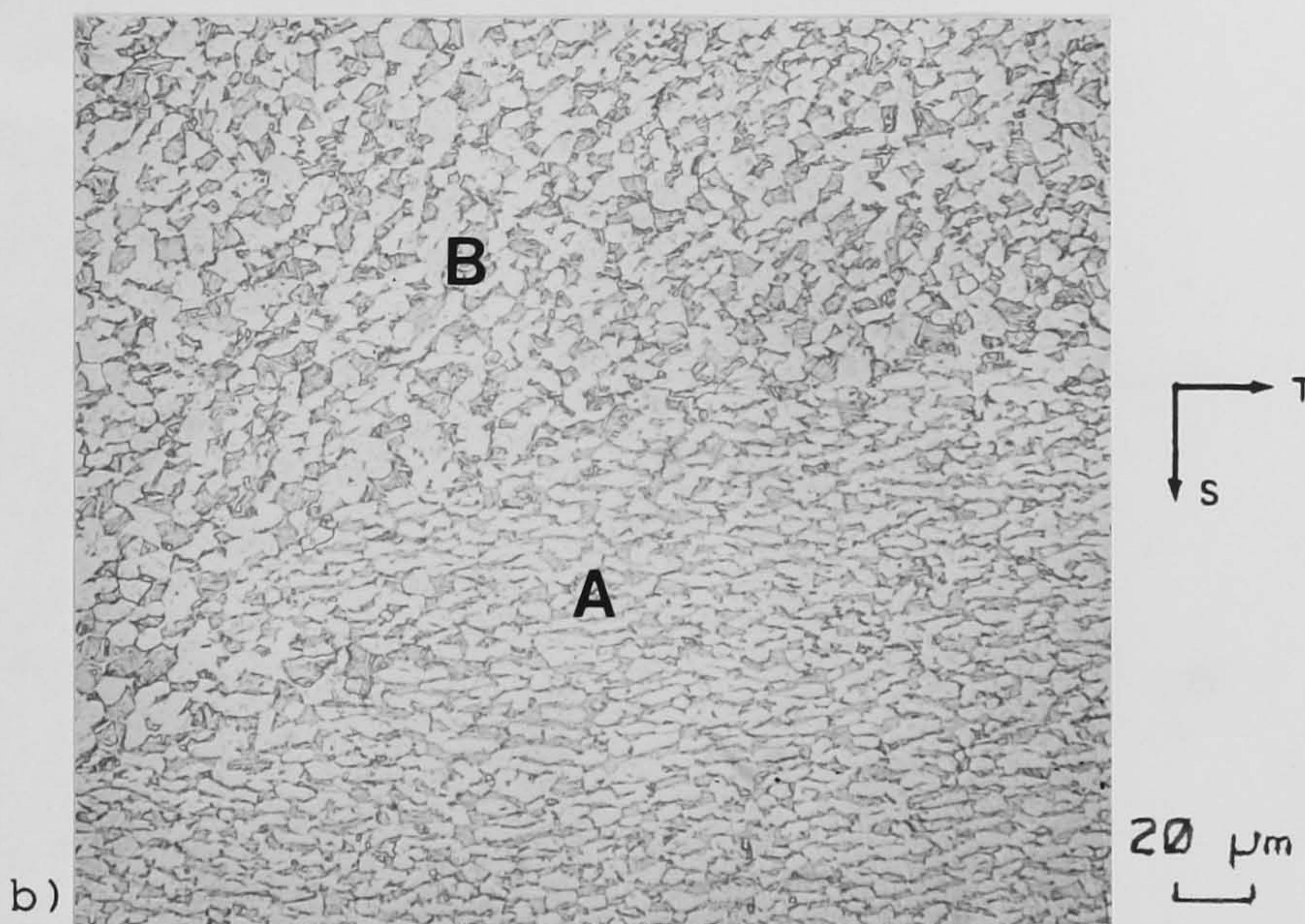
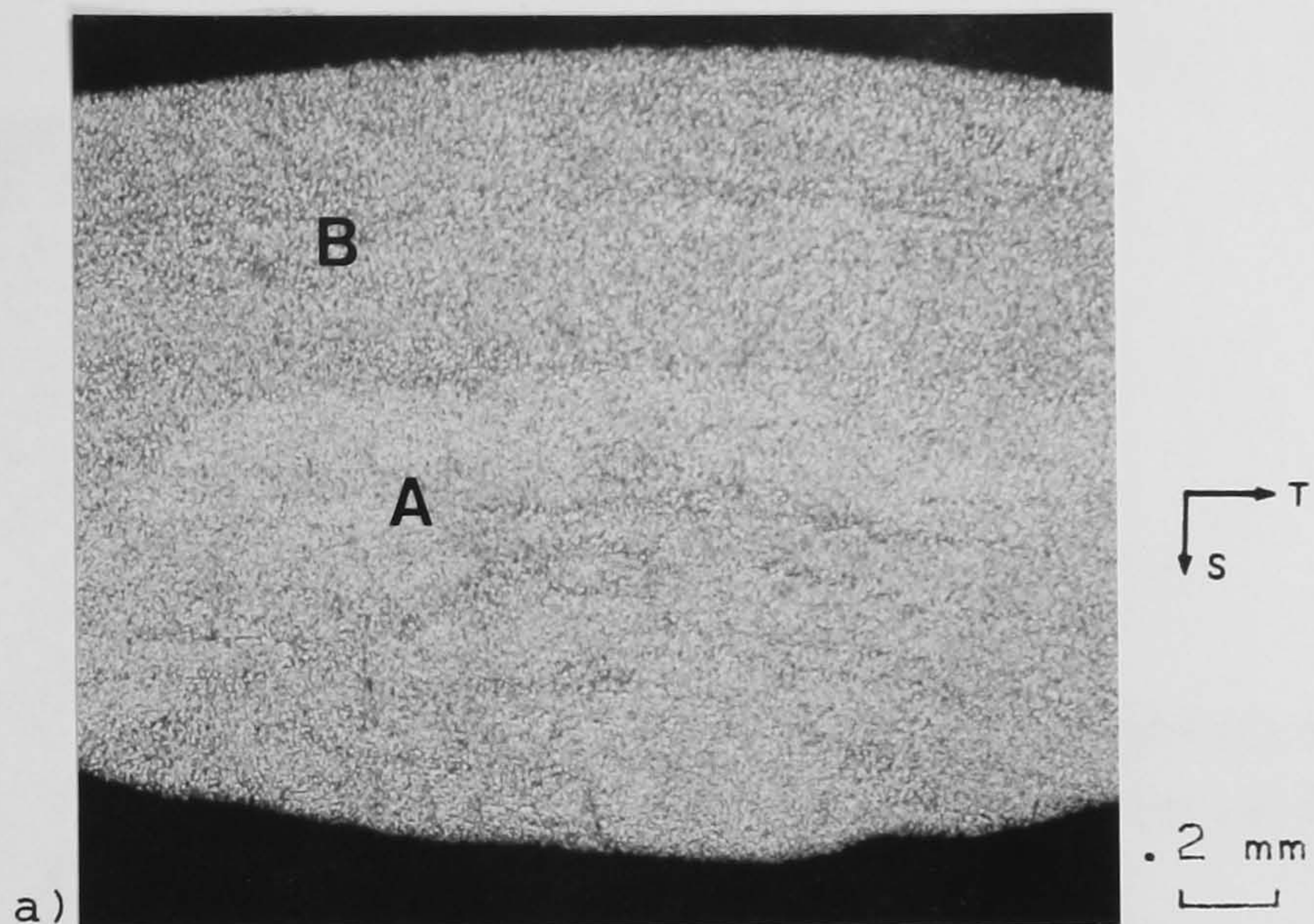


Fig 42 TL orientation test piece after 200% strain at 925 °C and $3 \times 10^{-4} \text{ s}^{-1}$. Tensile axis T

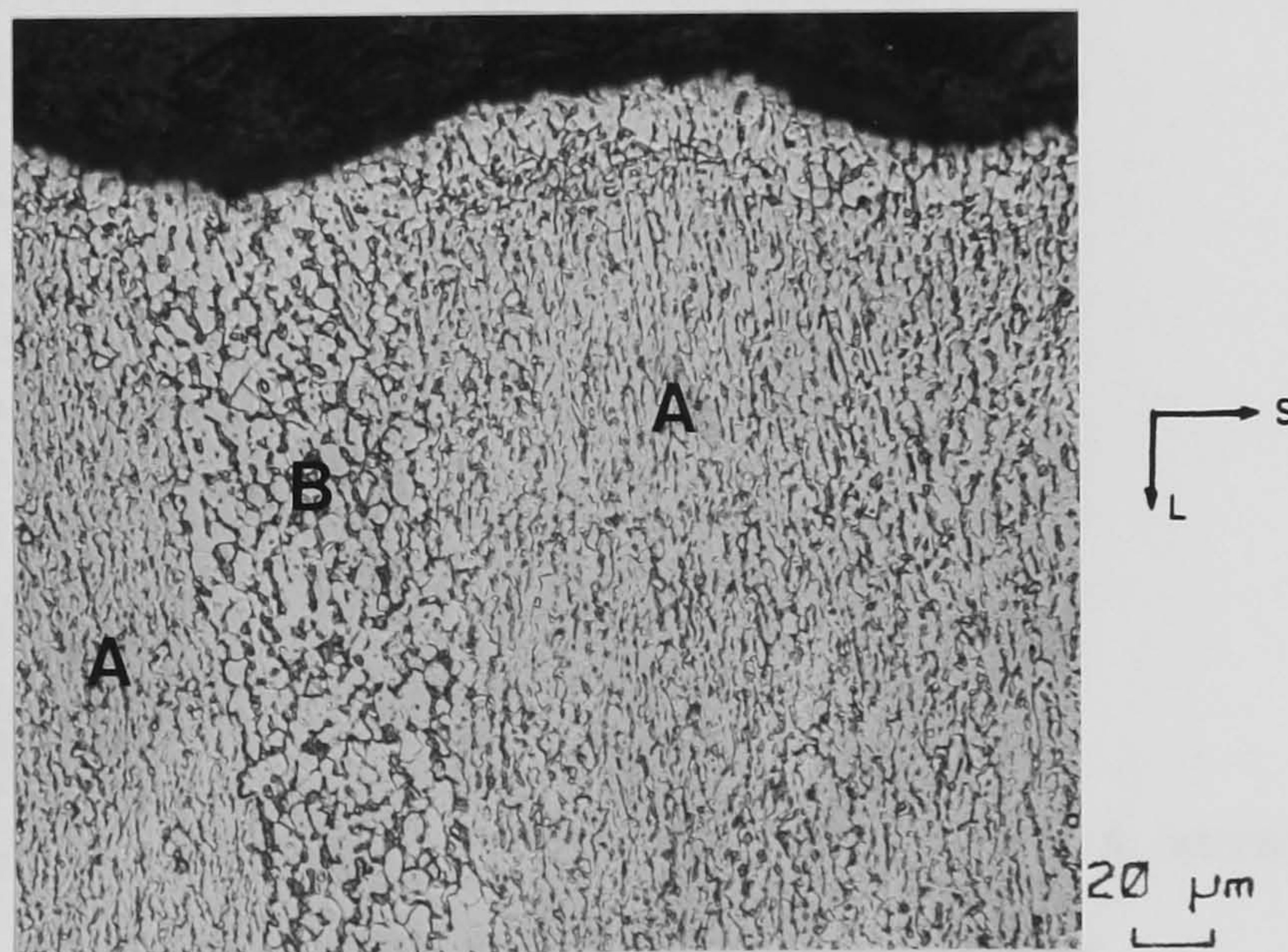
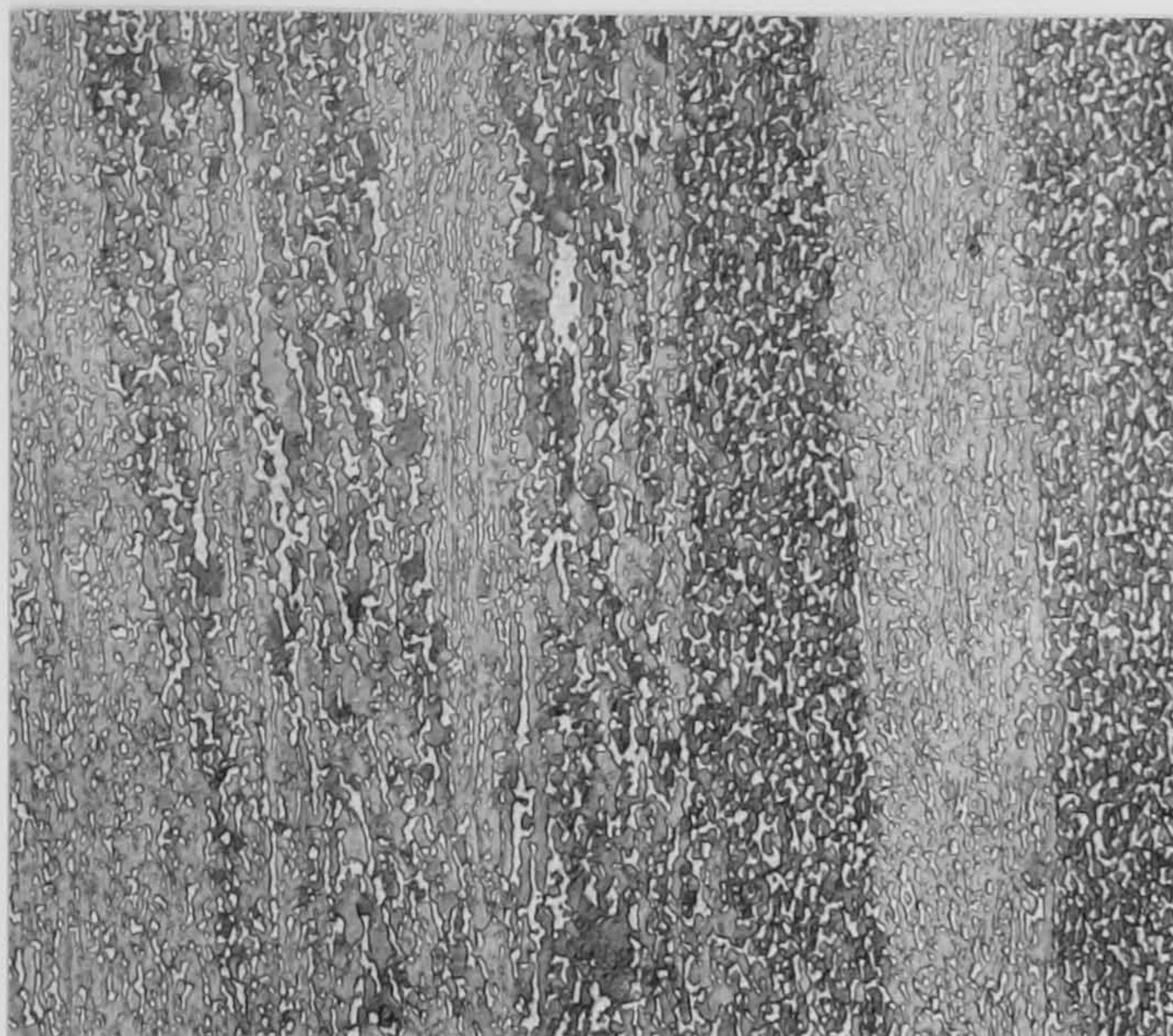
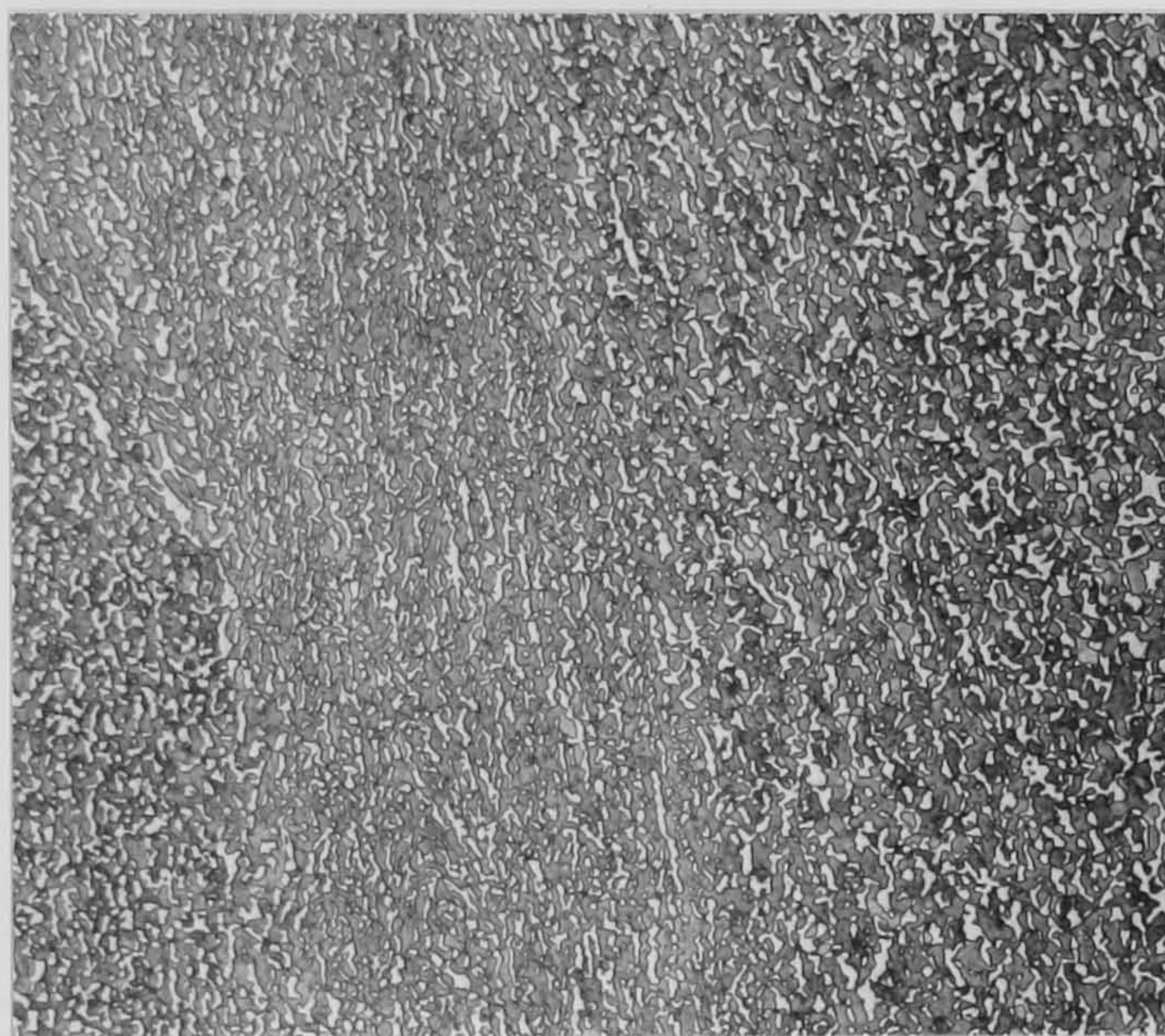


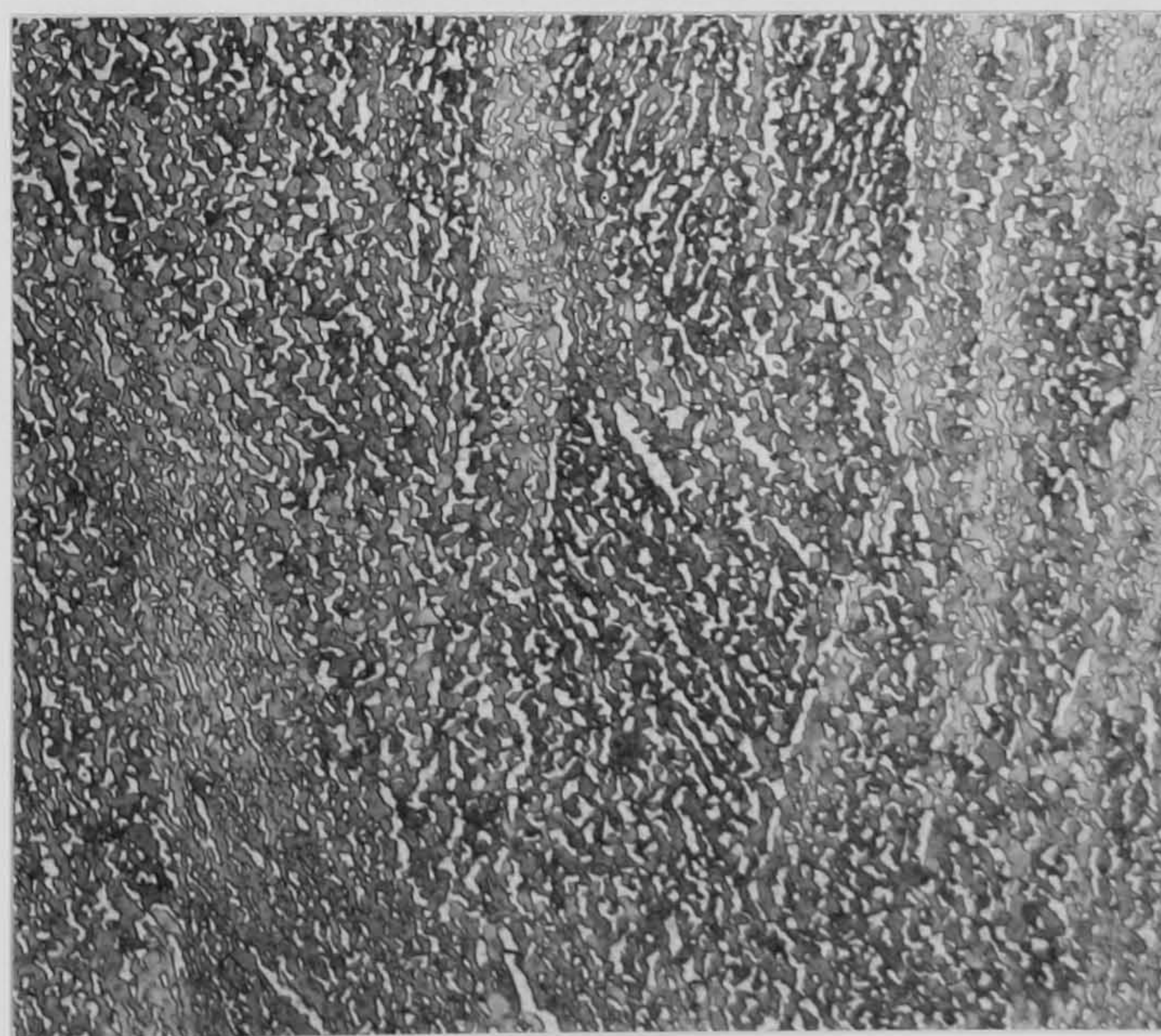
Fig 43 ST orientation test piece after 200% strain at 875 °C and $3 \times 10^{-4} \text{ s}^{-1}$. Tensile axis S



a) Annealed at 875 °C
(unstrained)

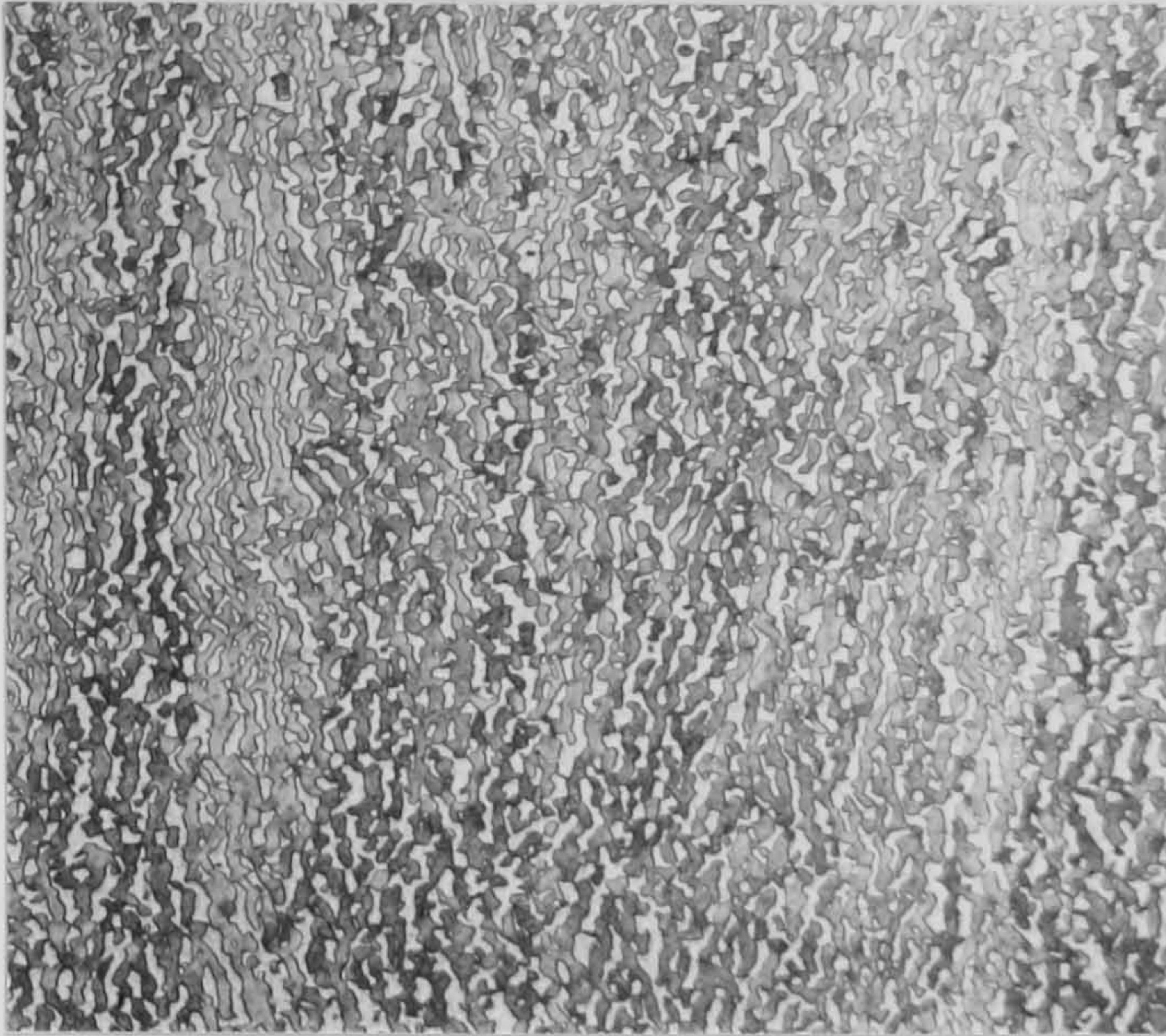


b) 12% strain

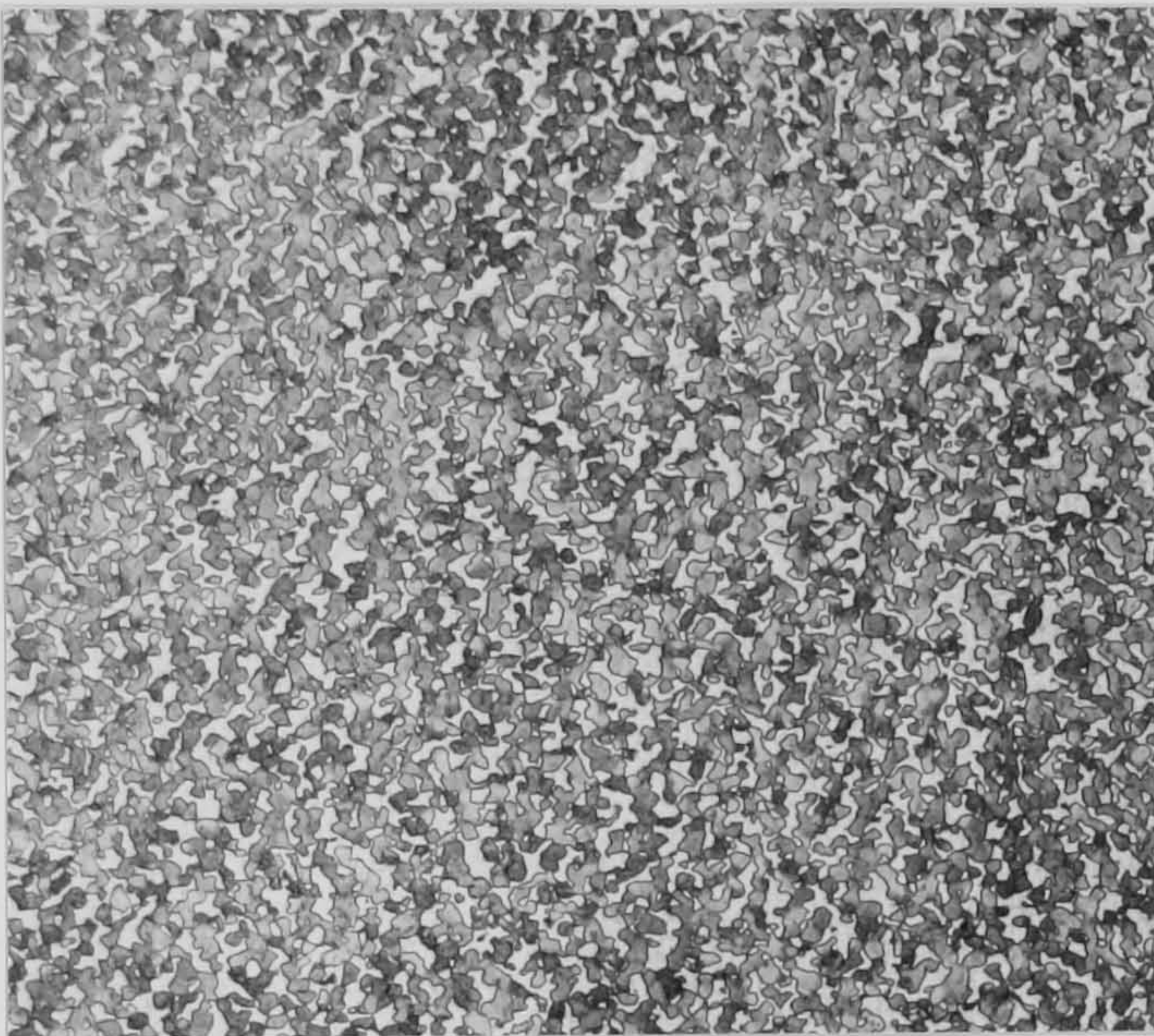


c) 24% strain

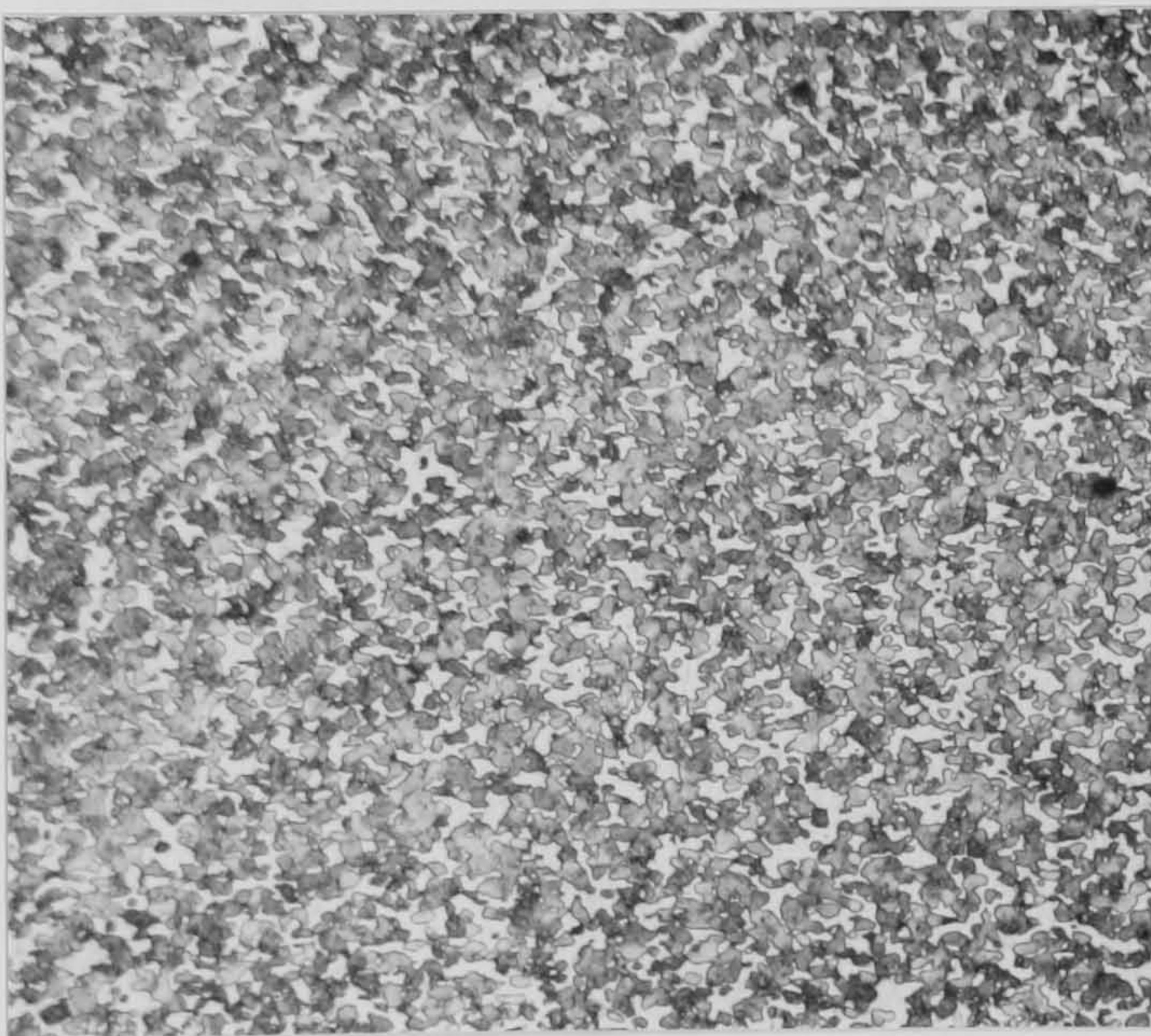
Fig 44 Effect of strain at 875 °C and $3 \times 10^{-4} \text{ s}^{-1}$ on
microstructure of Ti-6Al-4V (reheated and quenched).
ST orientation test piece B5. Tensile axis S



d) 169% strain

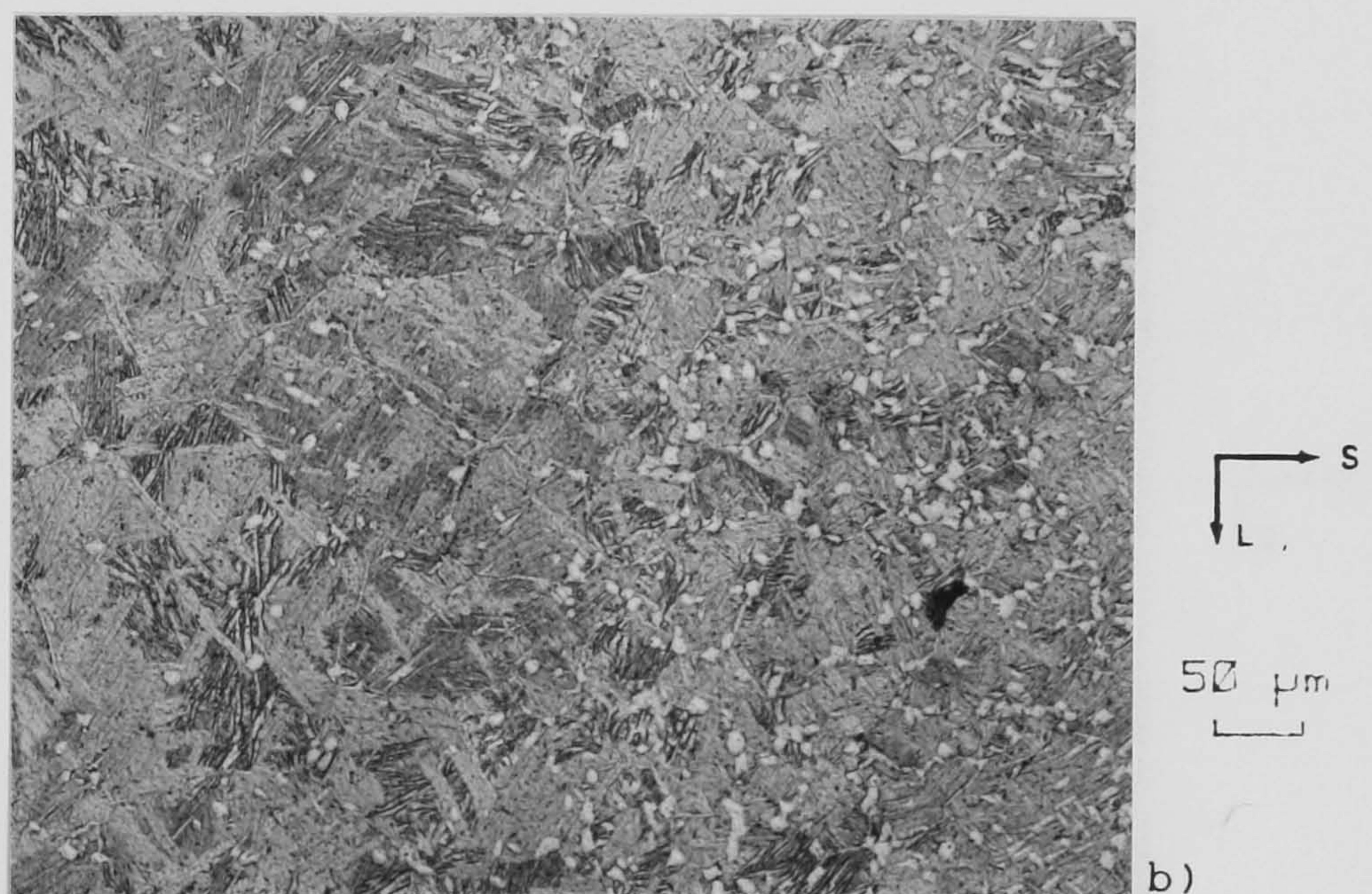
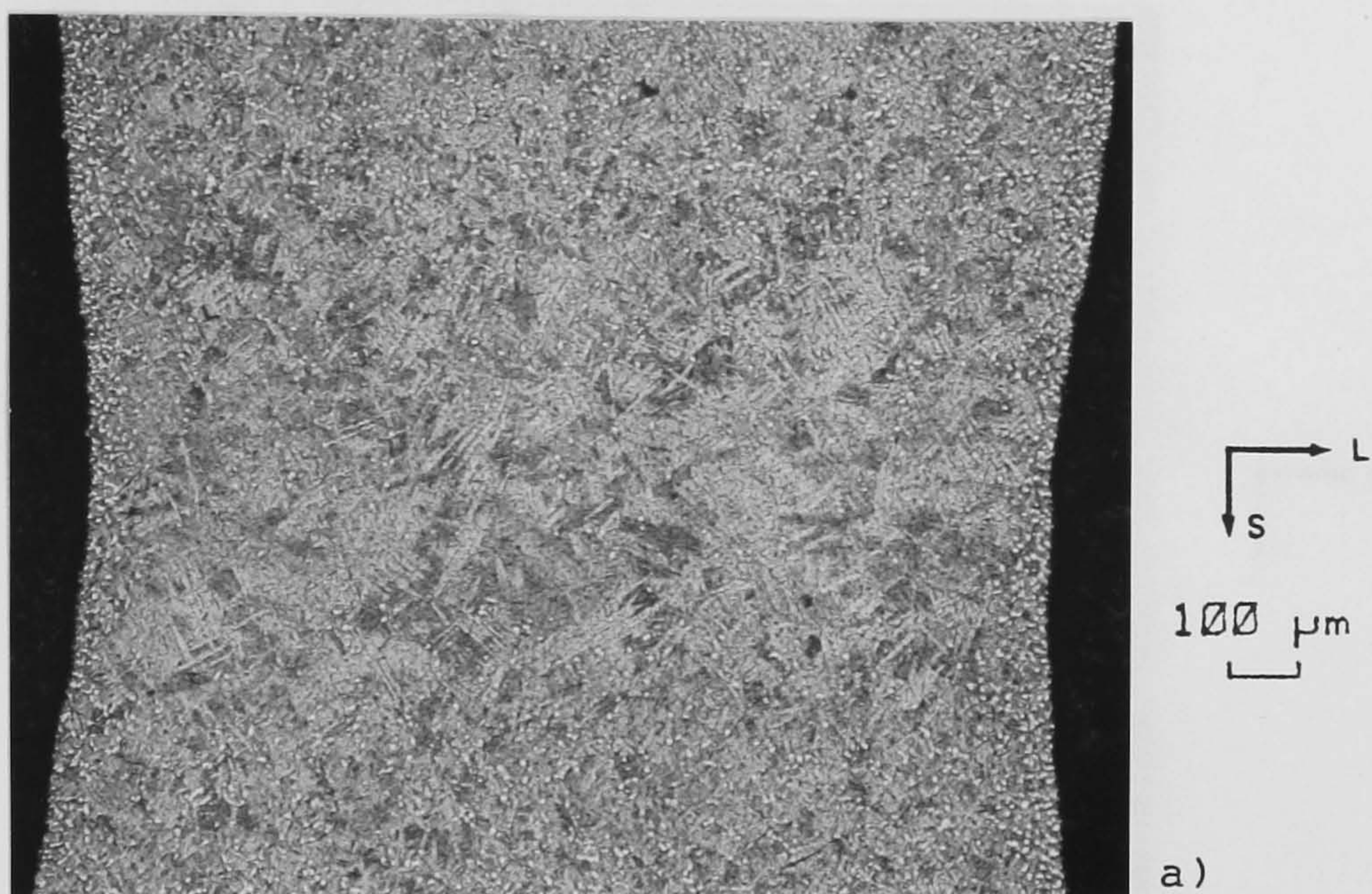


e) 468% strain



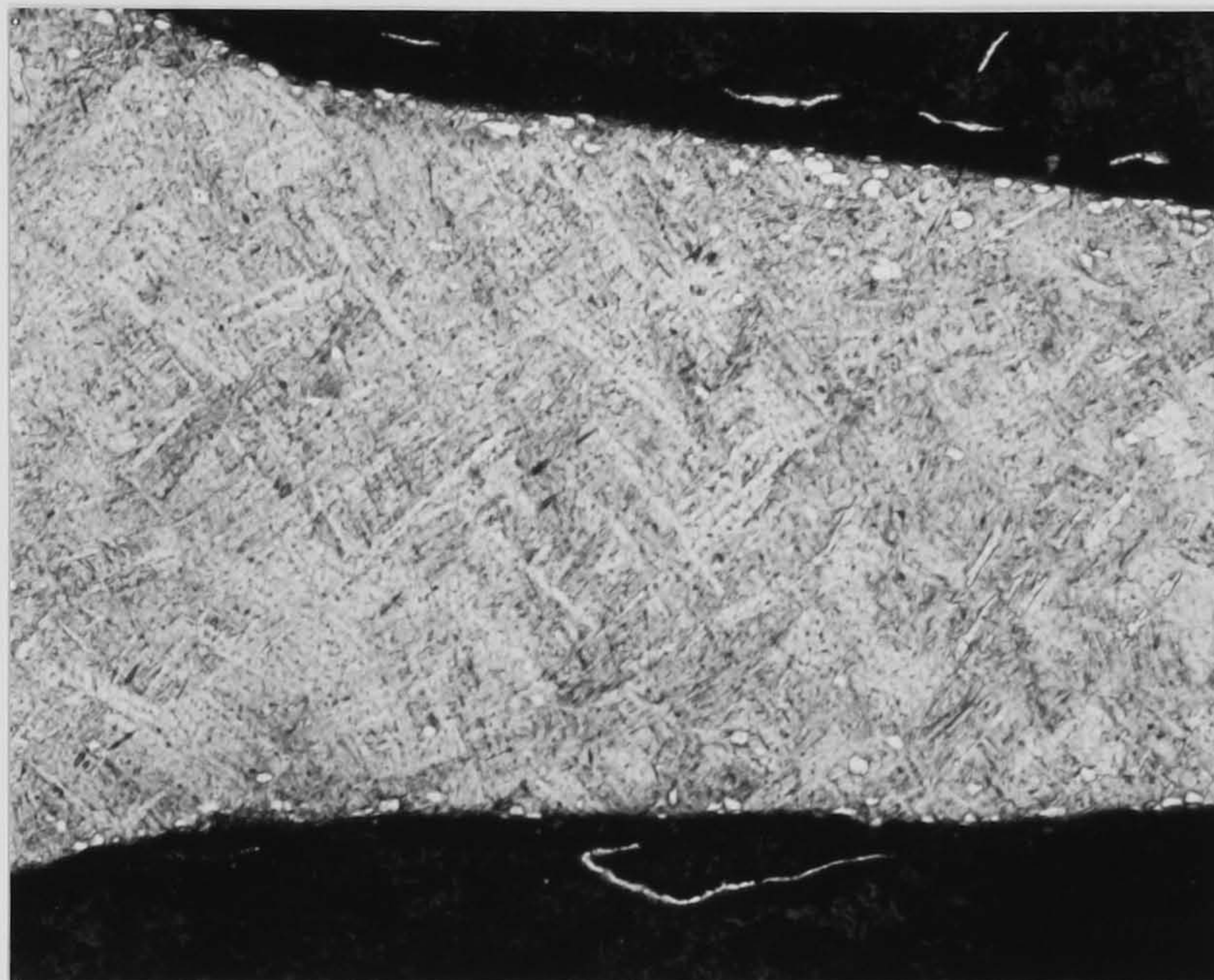
f) 2000% strain
(necked region)

Fig 44 Effect of strain at 875°C and $3 \times 10^{-4} \text{s}^{-1}$ on microstructure of Ti-6Al-4V (reheated and quenched). ST orientation test piece B5. Tensile axis S



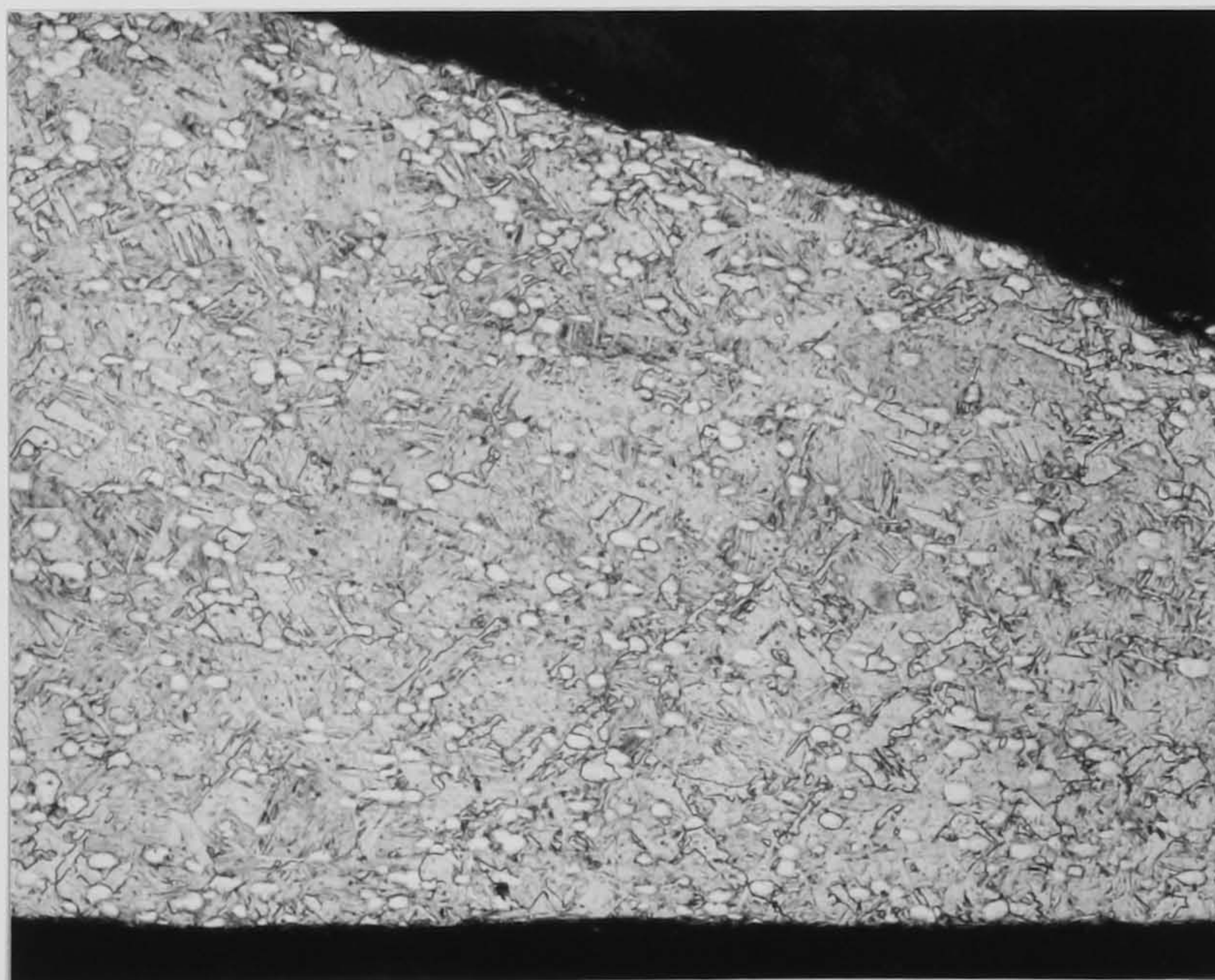
← neck region | uniform strain →

Fig 45 Microstructure inside neck and in adjacent region of ST orientation test piece B4 pulled at 975°C and $1.5 \times 10^{-3} \text{ s}^{-1}$. Tensile axis S



a) ST orientation test piece B4 pulled
at 975 °C and $1.5 \times 10^{-3} \text{ s}^{-1}$

50 μm



b) TL orientation test piece A3 pulled
at 975 C and $3 \times 10^{-4} \text{ s}^{-1}$

Fig 46 Necked regions of ST and TL test pieces
pulled at 975 °C

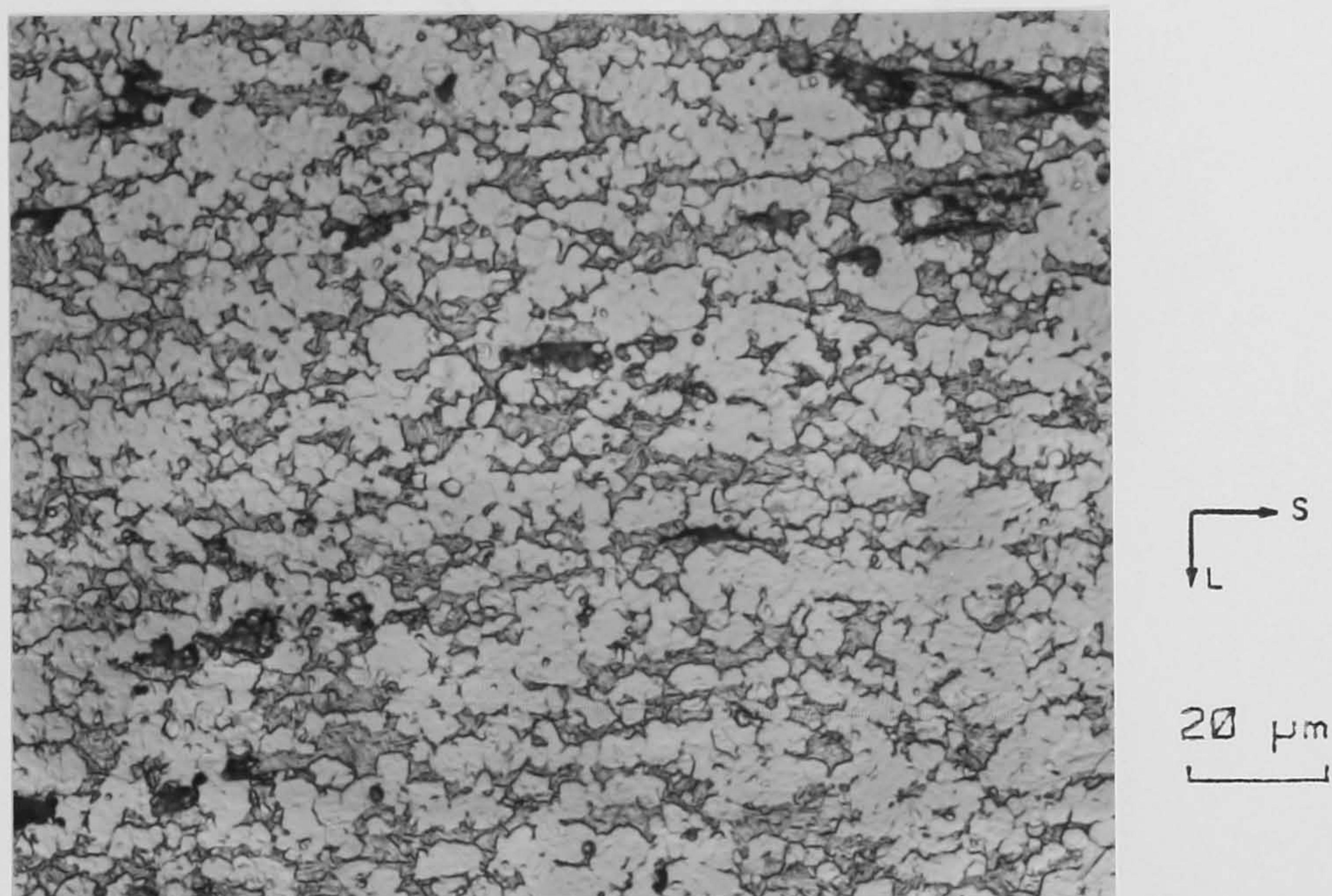


Fig 47 ST orientation test piece B2 pulled at 875 °C and $1.5 \times 10^{-3} \text{ s}^{-1}$. Cavitation inside necked region. Tensile axis S

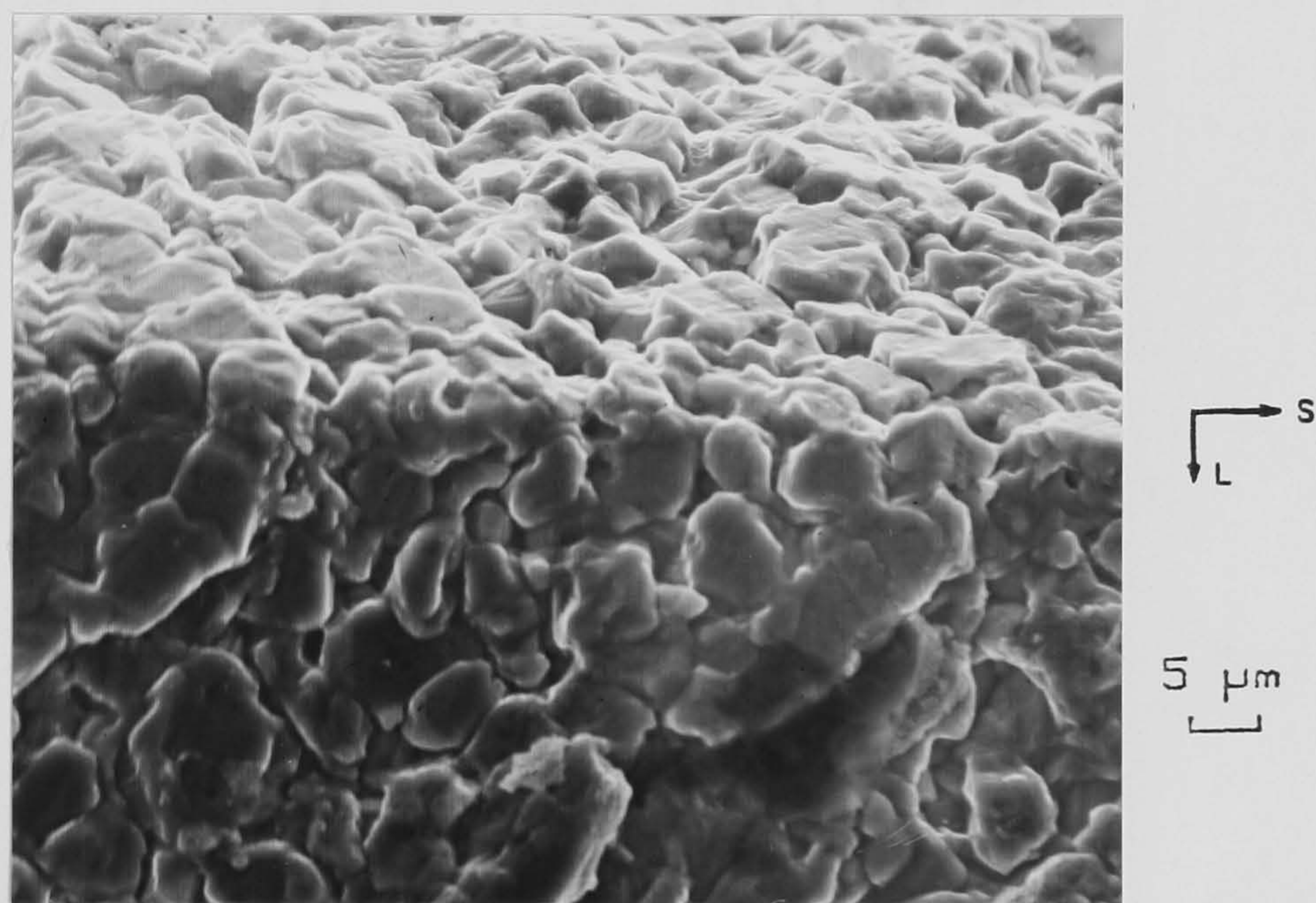


Fig 48 ST orientation test piece B5 pulled at 875 °C and $3 \times 10^{-4} \text{ s}^{-1}$ showing test piece edge. Tensile axis S. Scanning electron micrograph

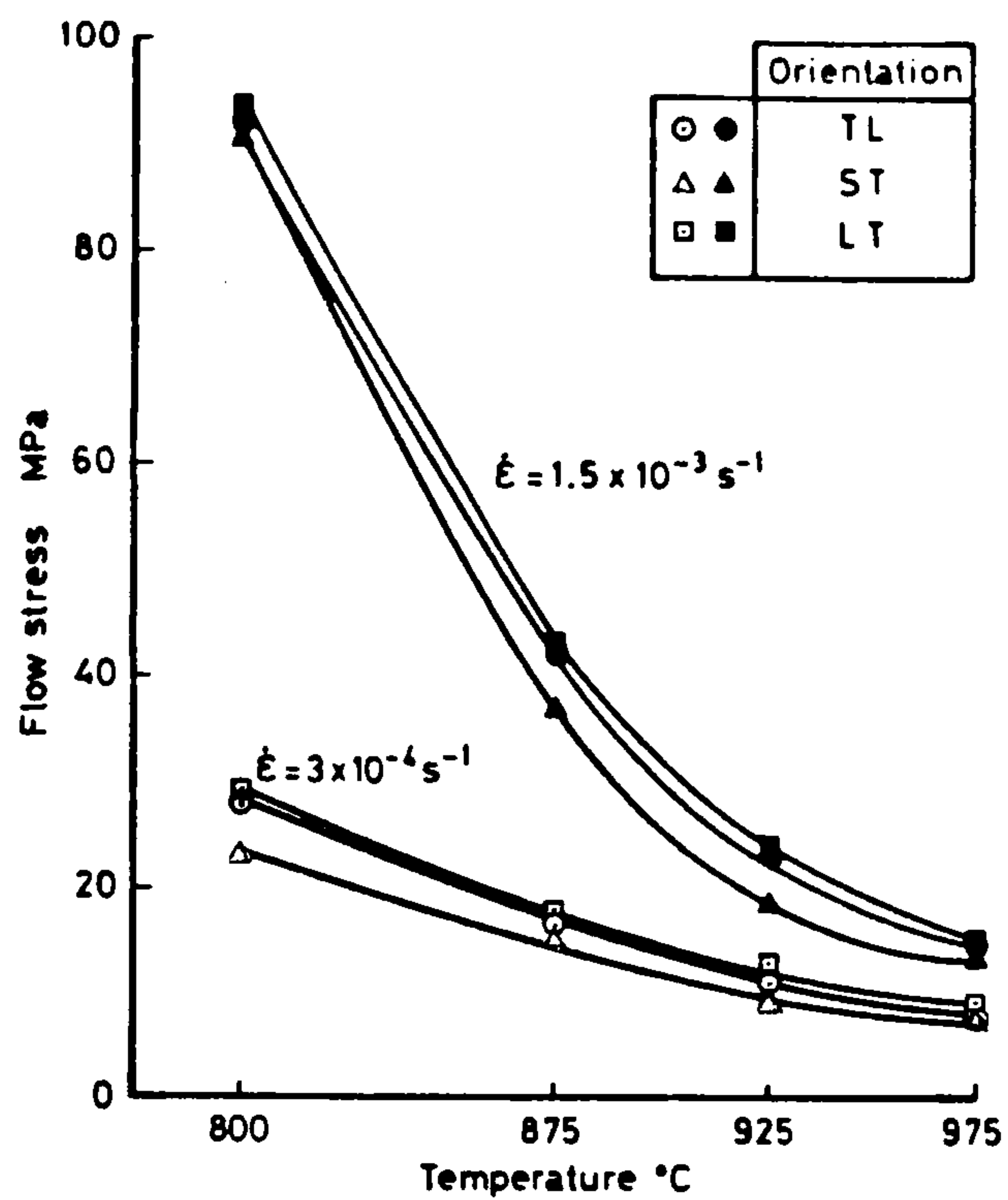


Fig 49 Flow stress at maximum load for each orientation and strain rate for Ti-6Al-4V sheet test pieces machined from bar

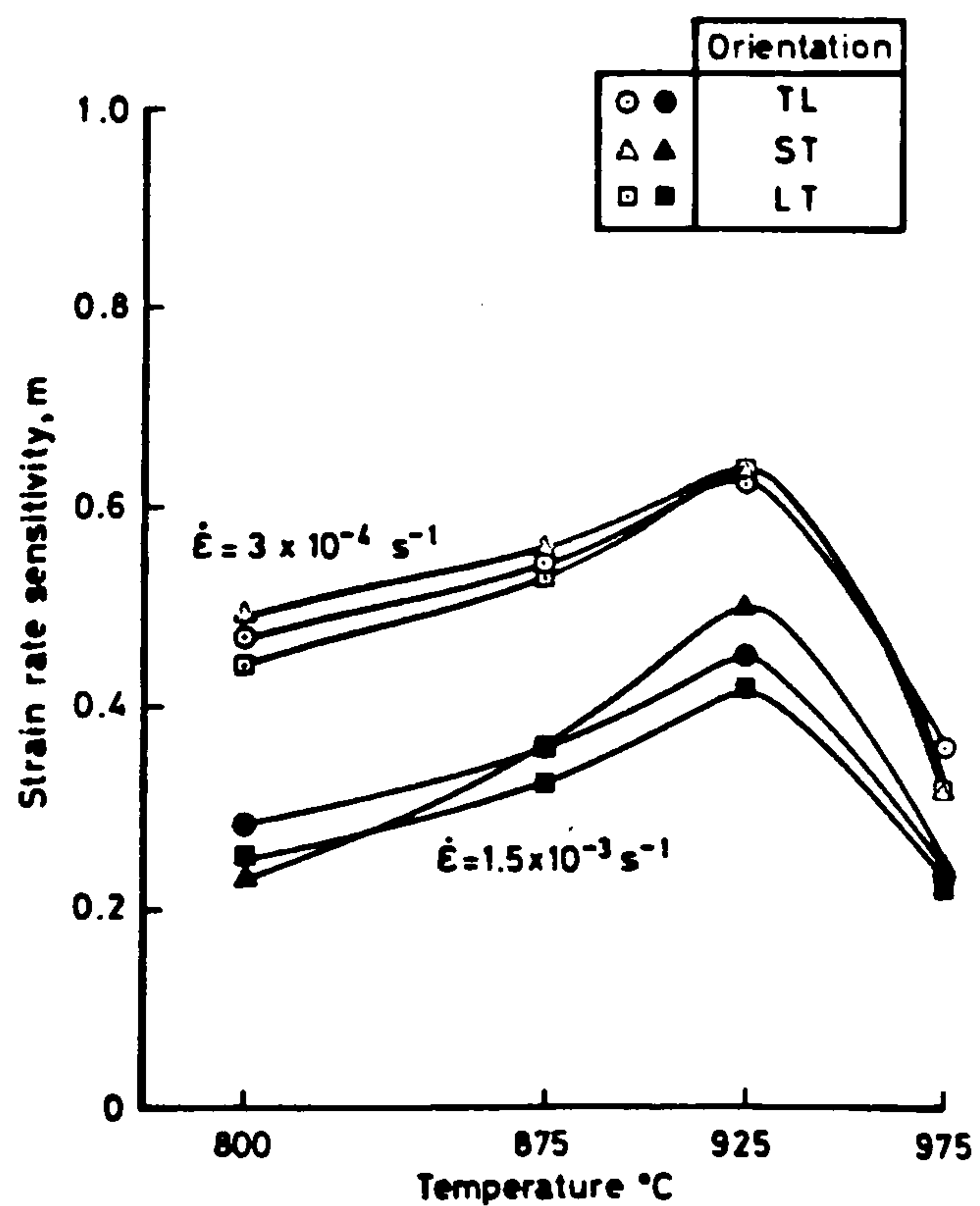


Fig 50 m values at maximum load for each orientation and strain rate for Ti-6Al-4V sheet test pieces machined from bar

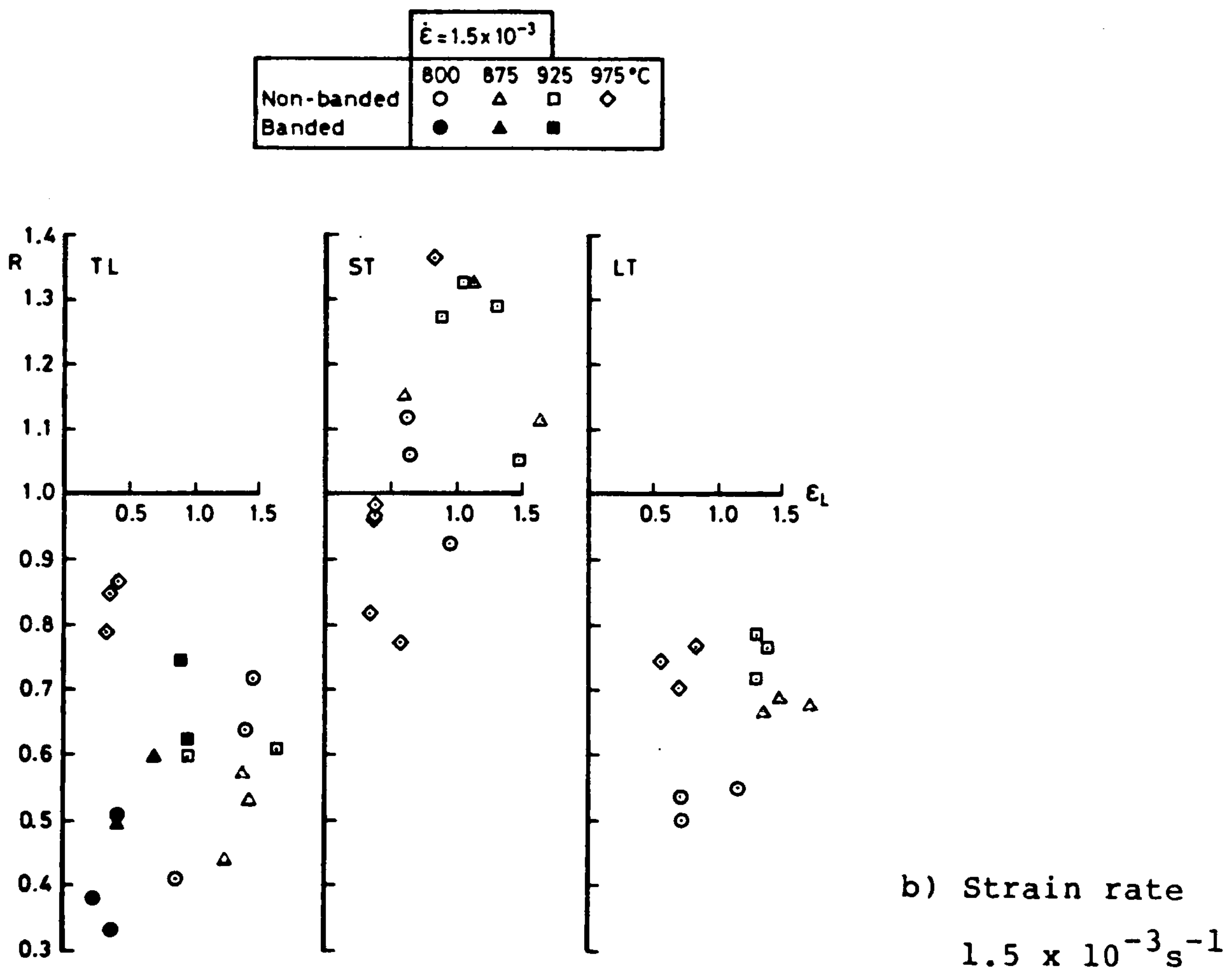
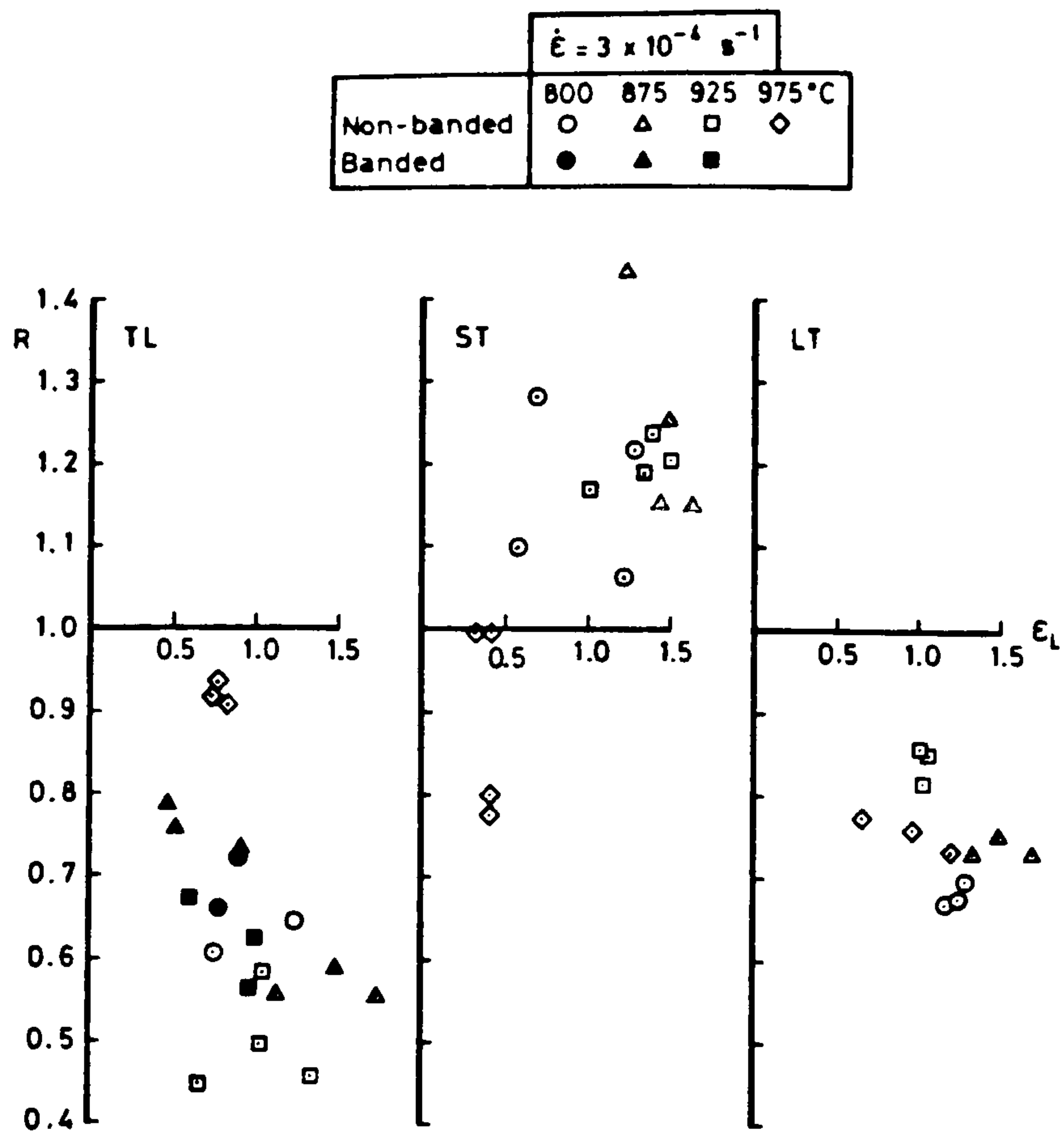
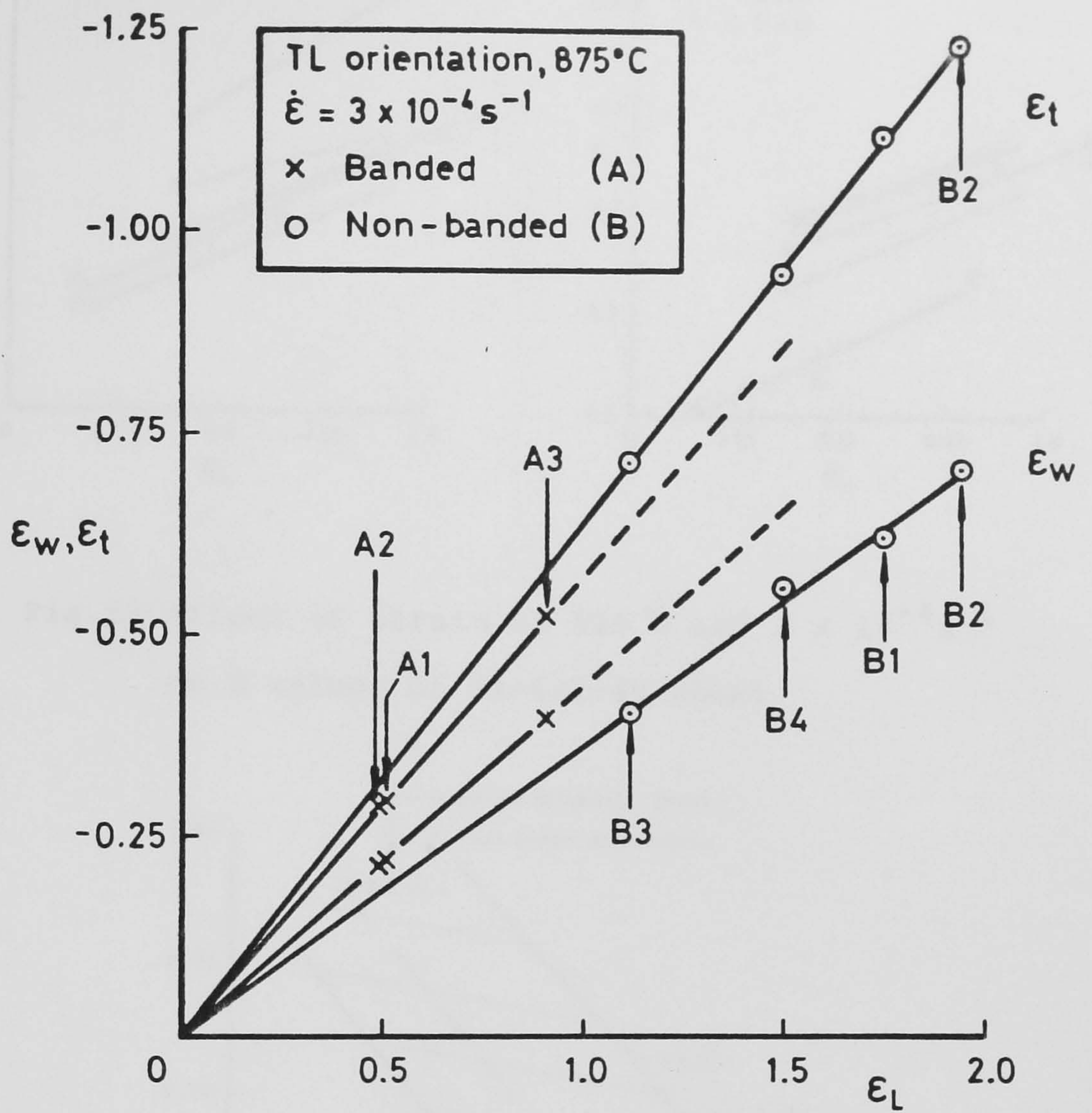
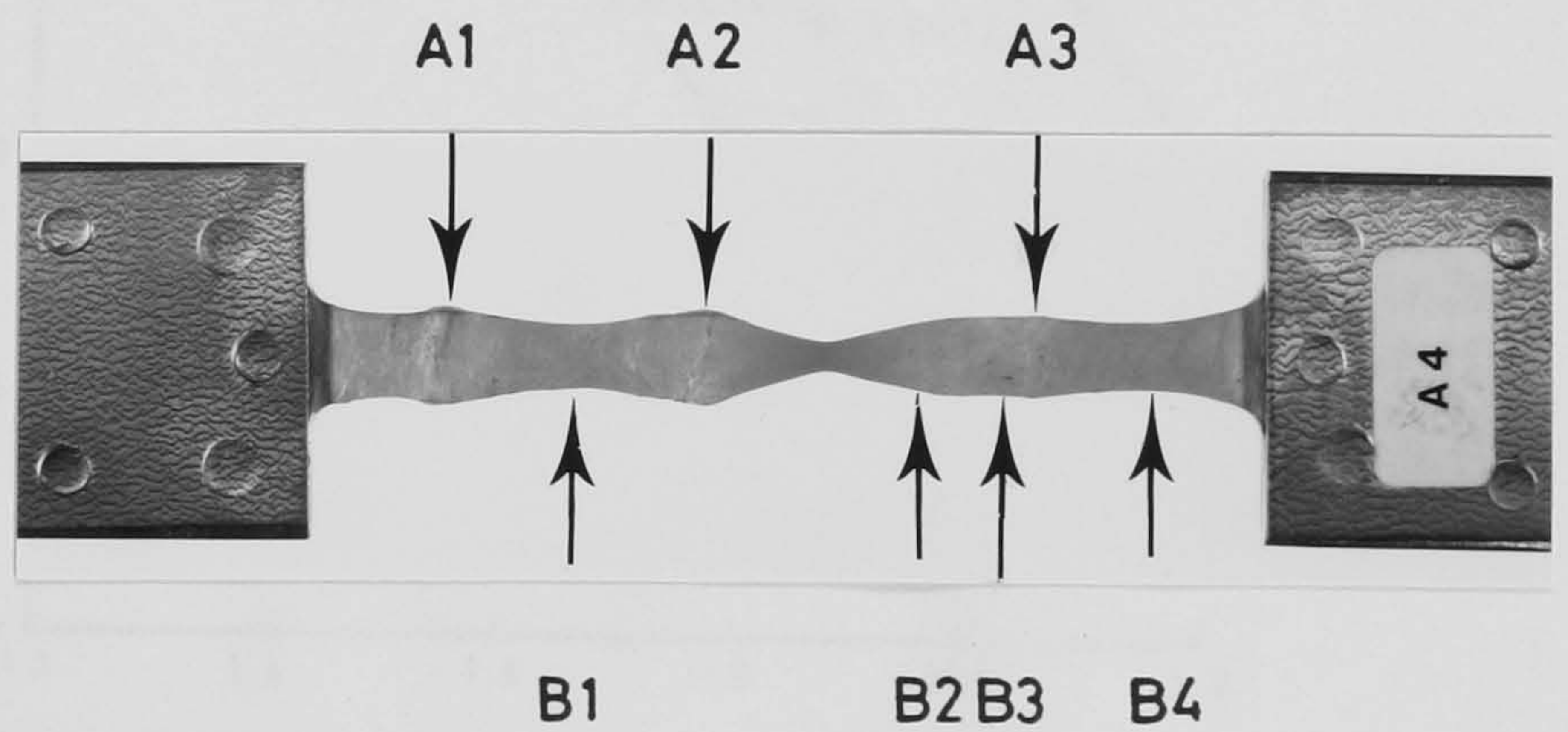


Fig 51 Effect of strain on R value for Ti-6Al-4V
test pieces machined from bar



a)



b)

Fig 52 Strain distribution for TL test piece A4 pulled at 875 °C and $3 \times 10^{-4} \text{ s}^{-1}$. The indices A1-A3 and B1-B4 in b) each refer to a set of two points in a)

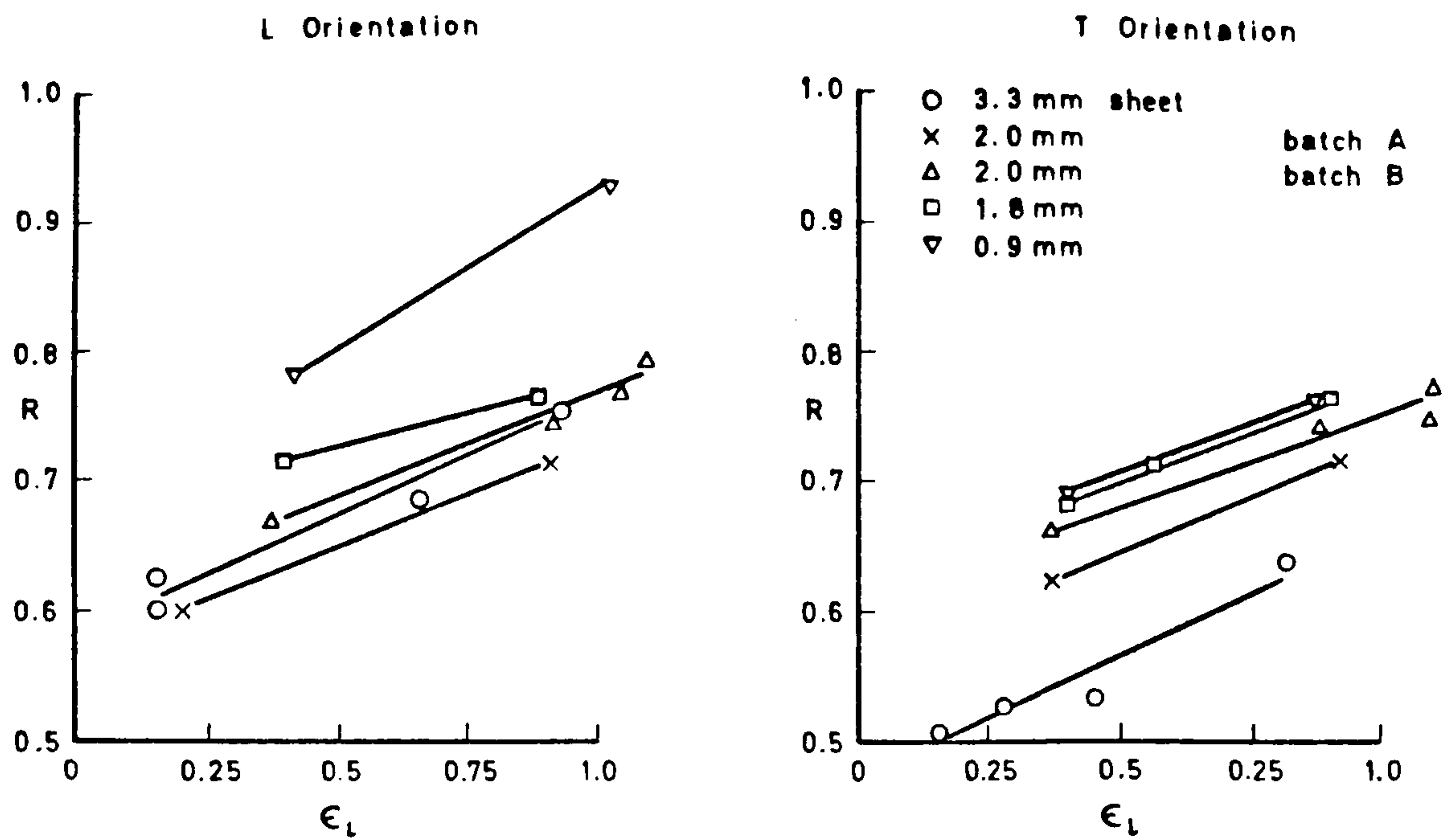


Fig 53 Effect of strain at 925°C and $3 \times 10^{-4} \text{s}^{-1}$ on R values of Ti-6Al-4V sheet

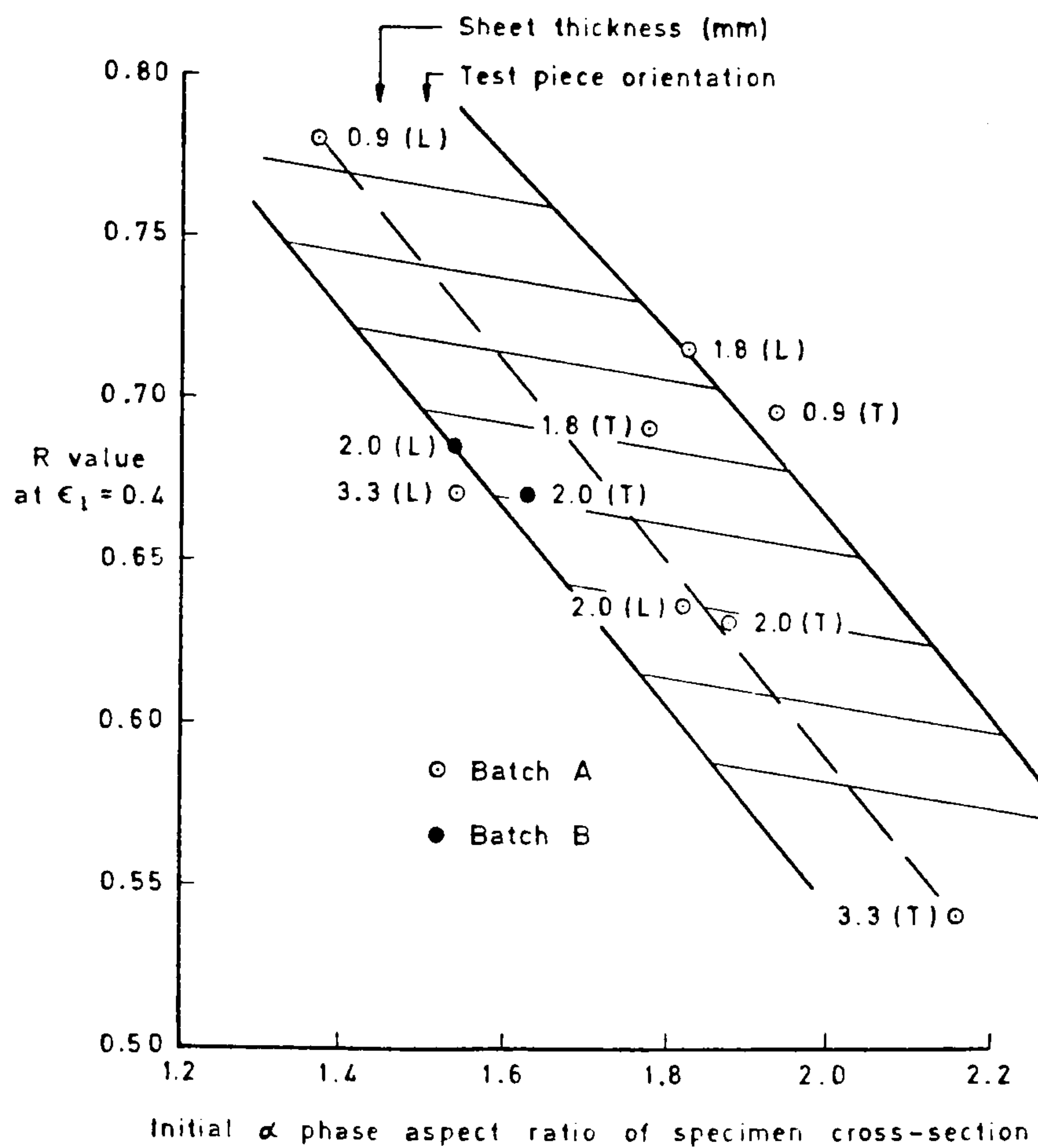
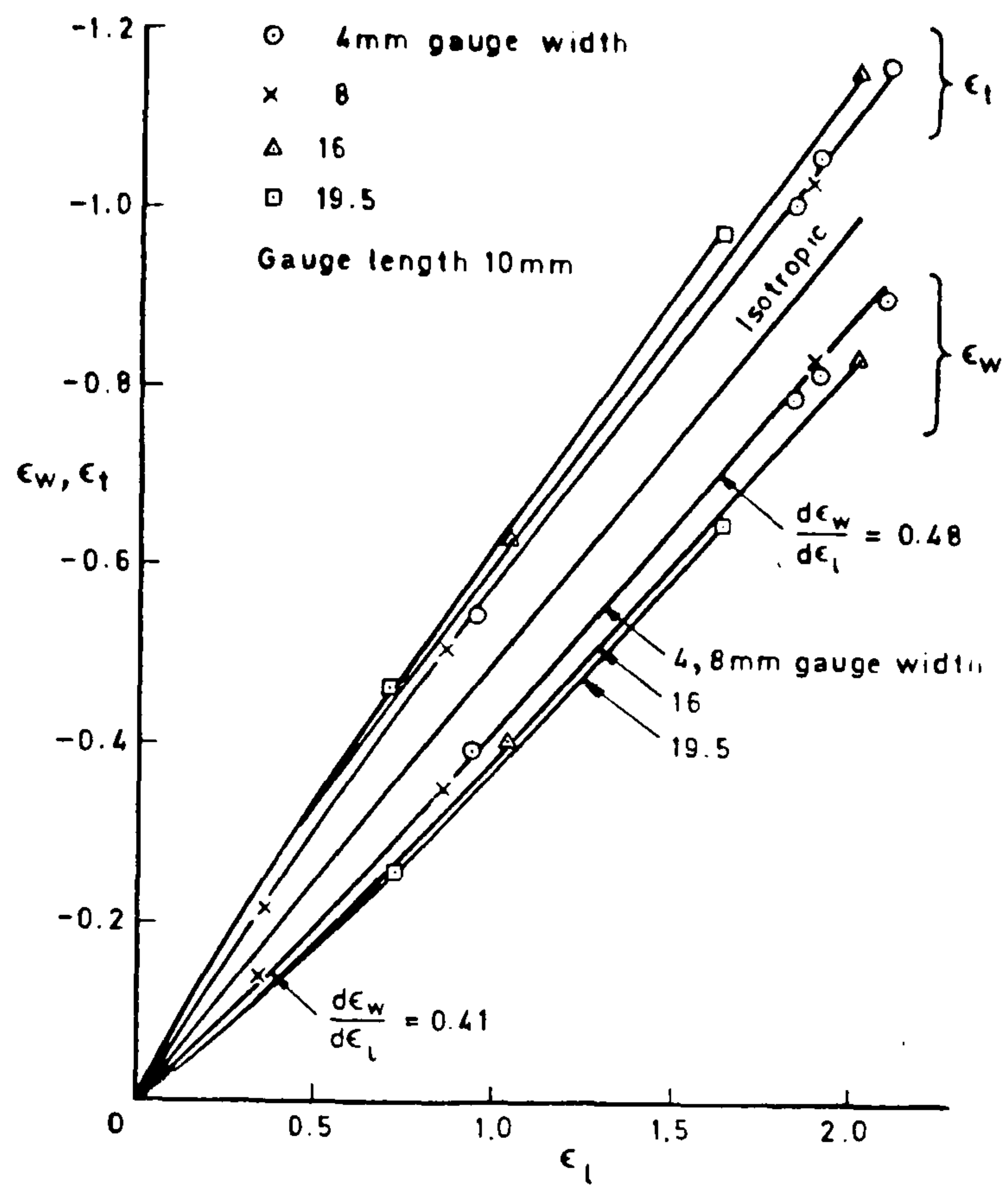
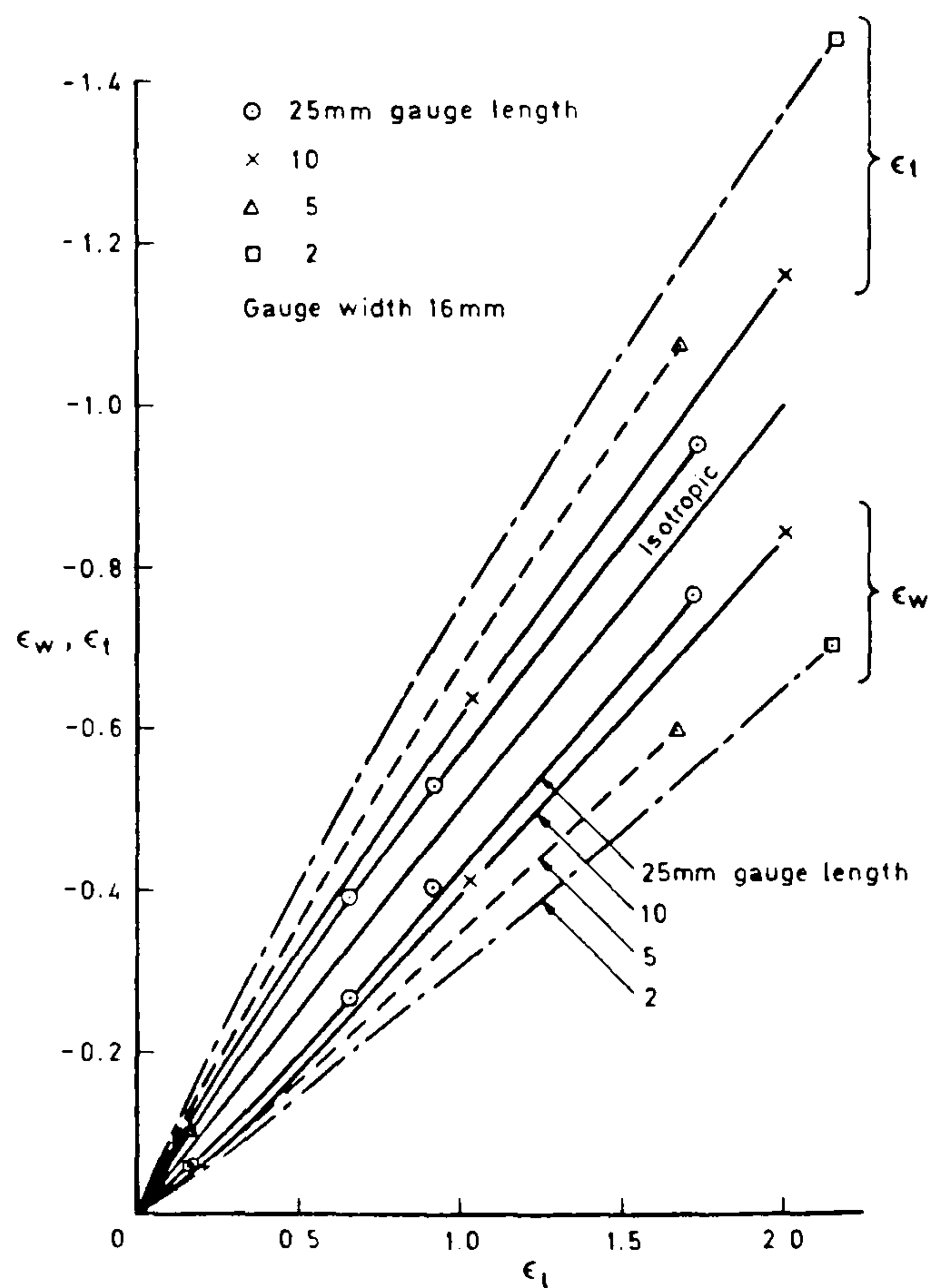


Fig 54 Relationship between alpha phase aspect ratio and R value for Ti-6Al-4V sheet after strain at 925°C and $3 \times 10^{-4} \text{s}^{-1}$

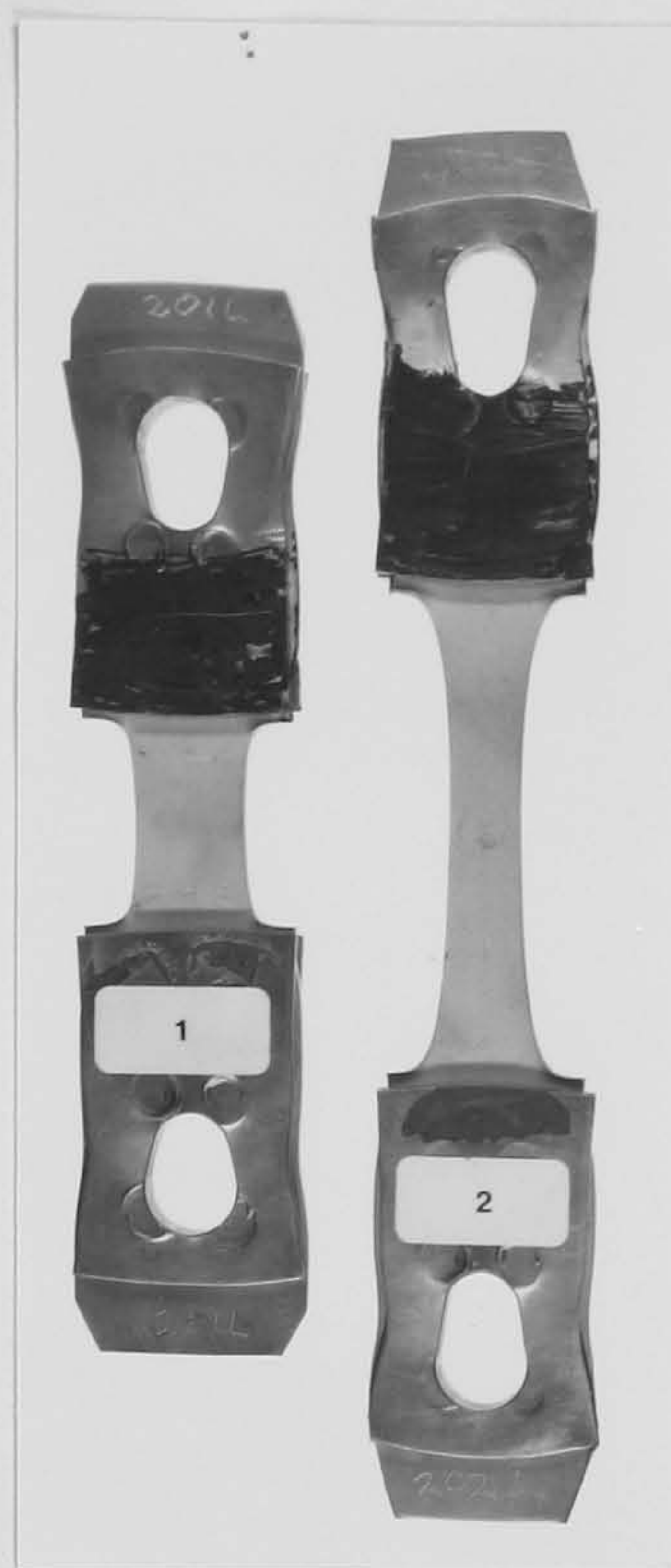


a) Effect of
 varying gauge
 width



b) Effect of
 varying gauge
 length

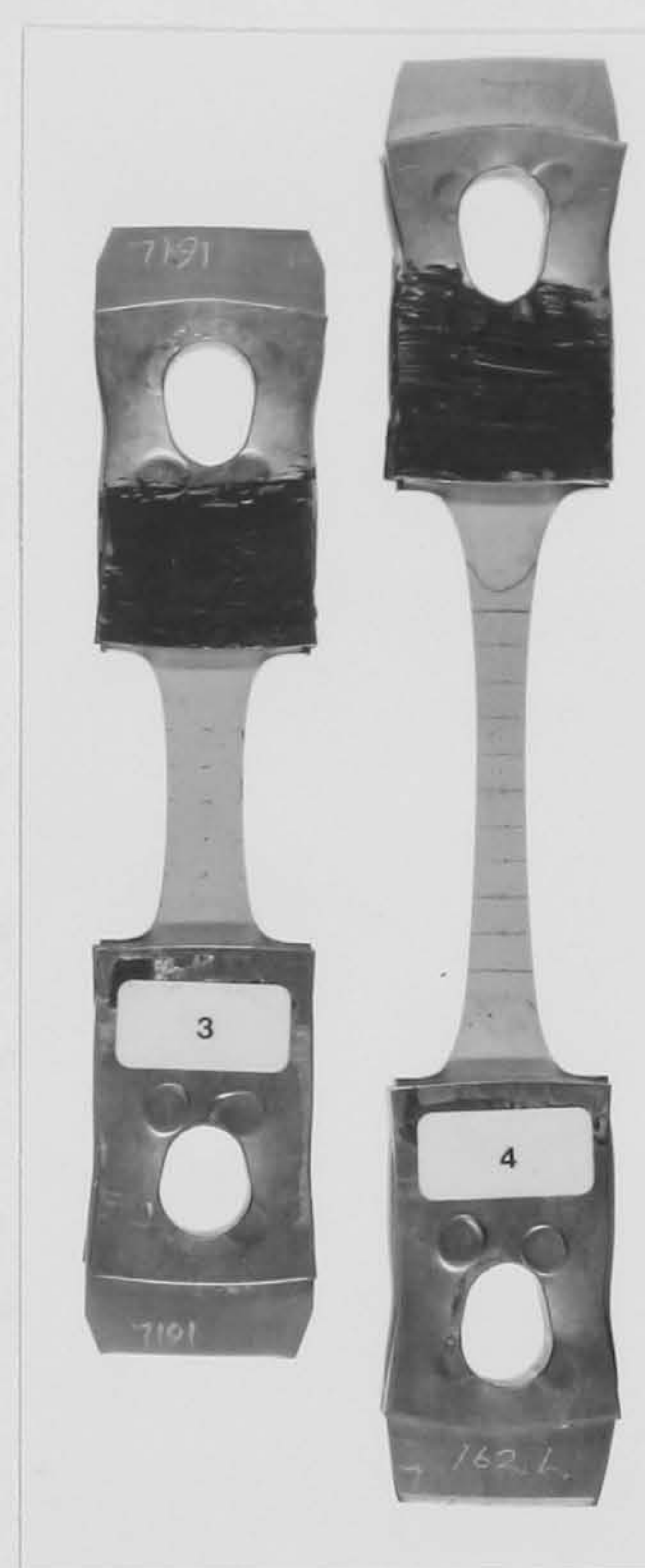
Fig 55 Effect of test piece geometry on superplastic
 strain anisotropy of Ti-6Al-4V sheet pulled at
 925 °C and $3 \times 10^{-4} \text{ s}^{-1}$



a) $w_o = 19.5\text{mm}$

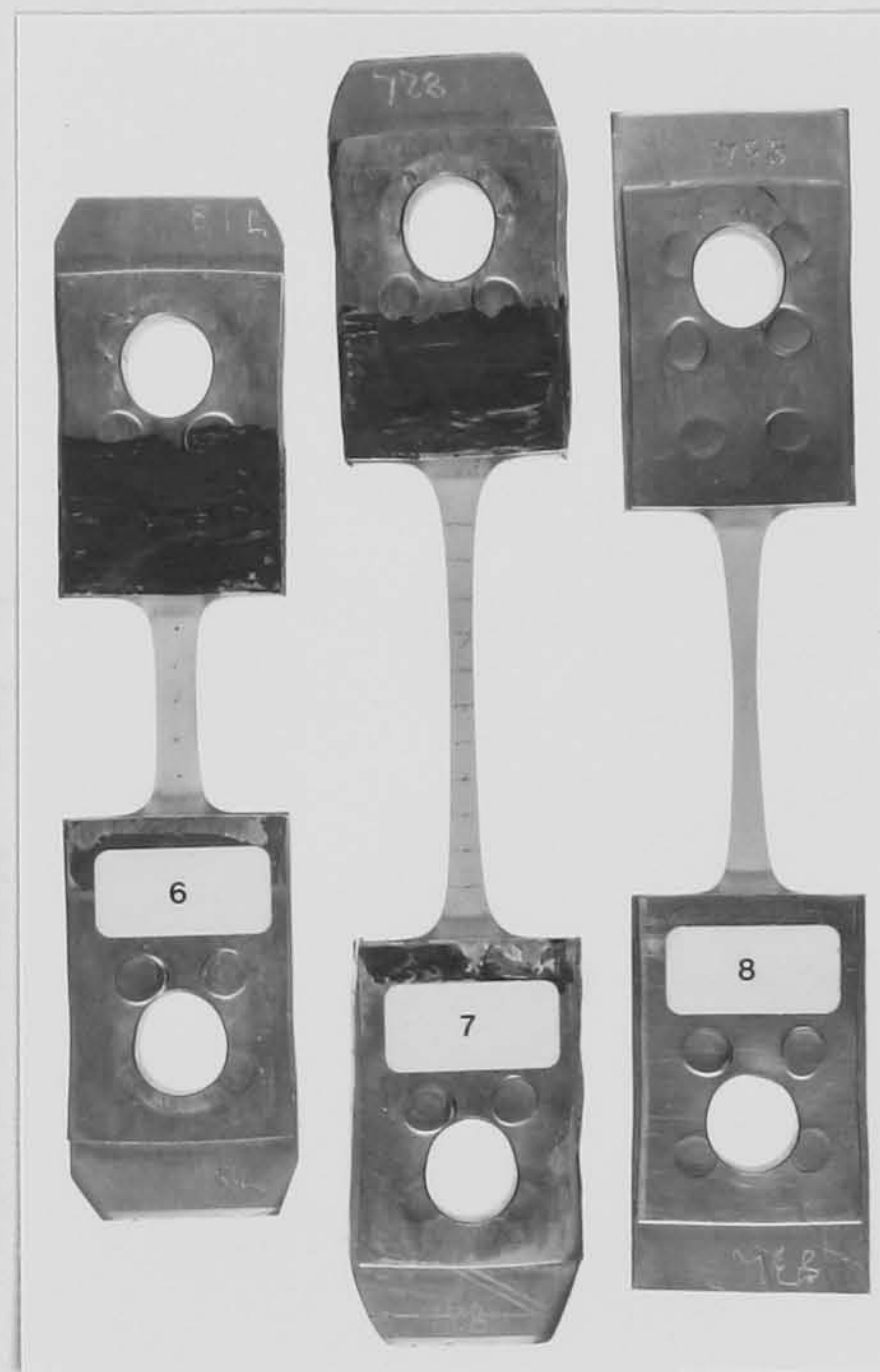
tp1 88%
tp2 320%

20 mm



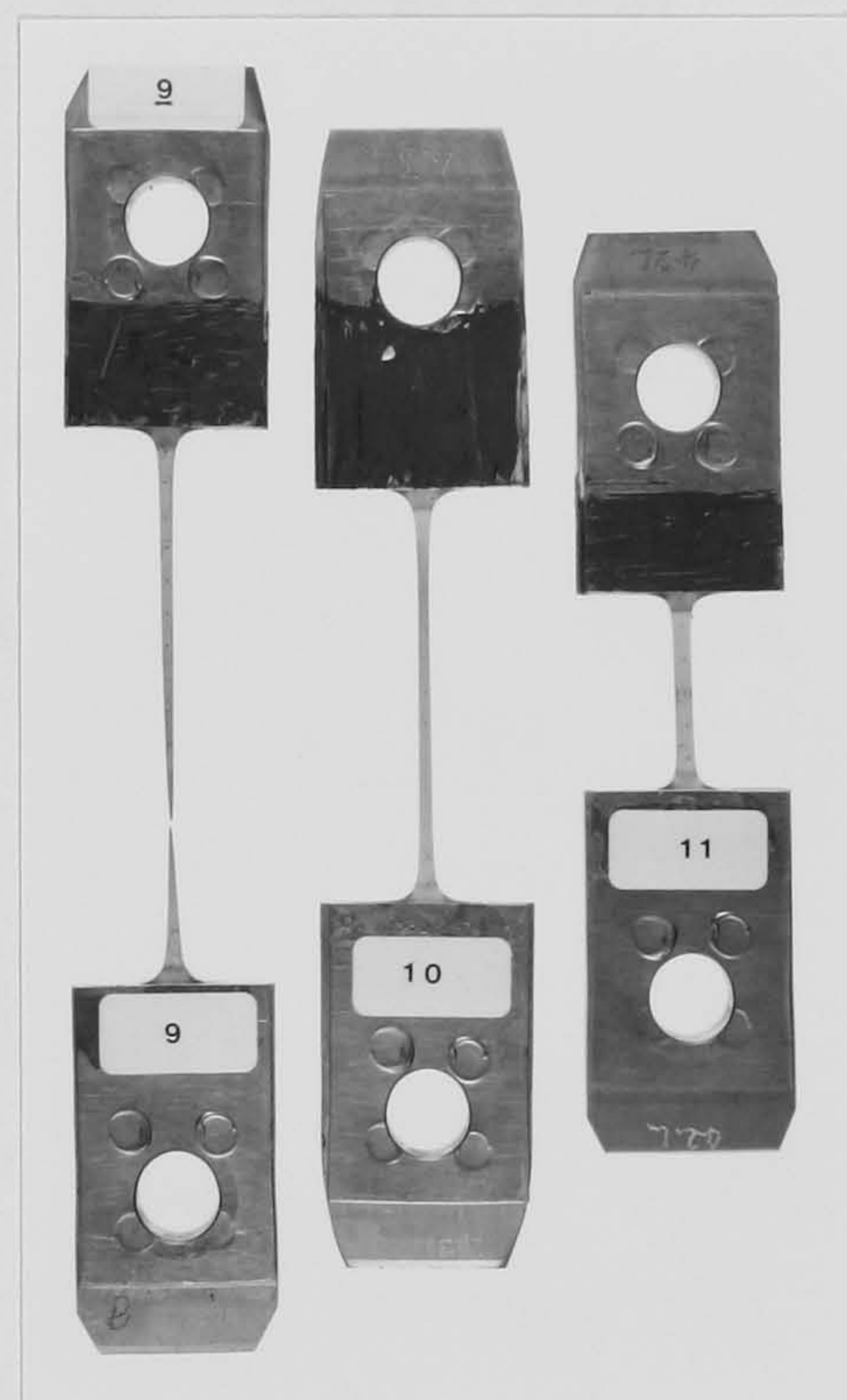
b) $w_o = 16\text{mm}$

tp3 252%
tp4 320%



c) $w_o = 8\text{mm}$

tp6 120%
tp7 400%
tp8 320%



d) $w_o = 4\text{mm}$

tp9 690%
tp10 460%
tp11 130%

Fig 56 Ti-6Al-4V sheet test pieces after strain at 925°C and $3 \times 10^{-4}\text{s}^{-1}$. Initial gauge length 10mm, initial gauge width w_o

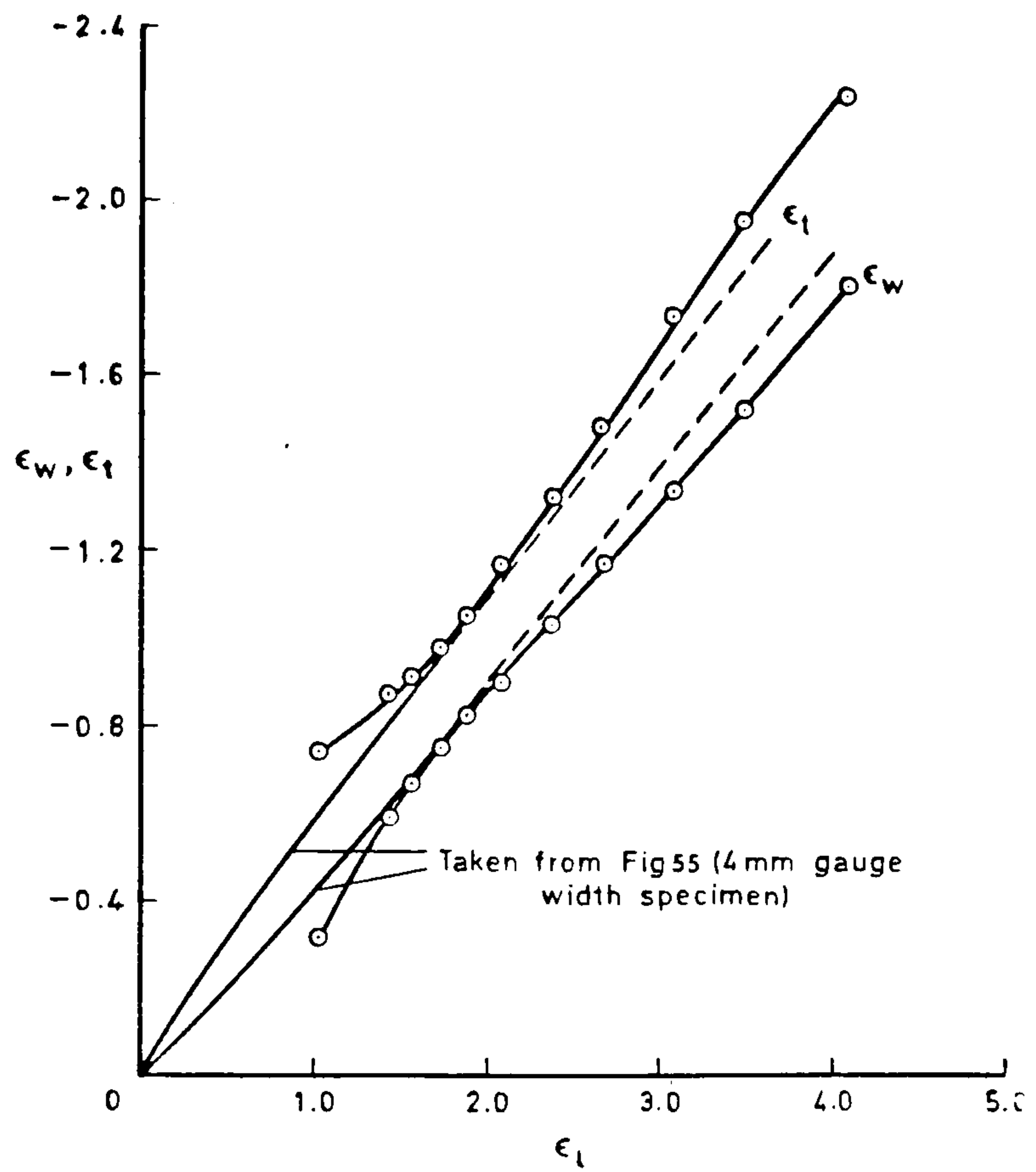


Fig 57 Strain distribution along the gauge length of a failed test piece (tp 9 in Fig 56d)

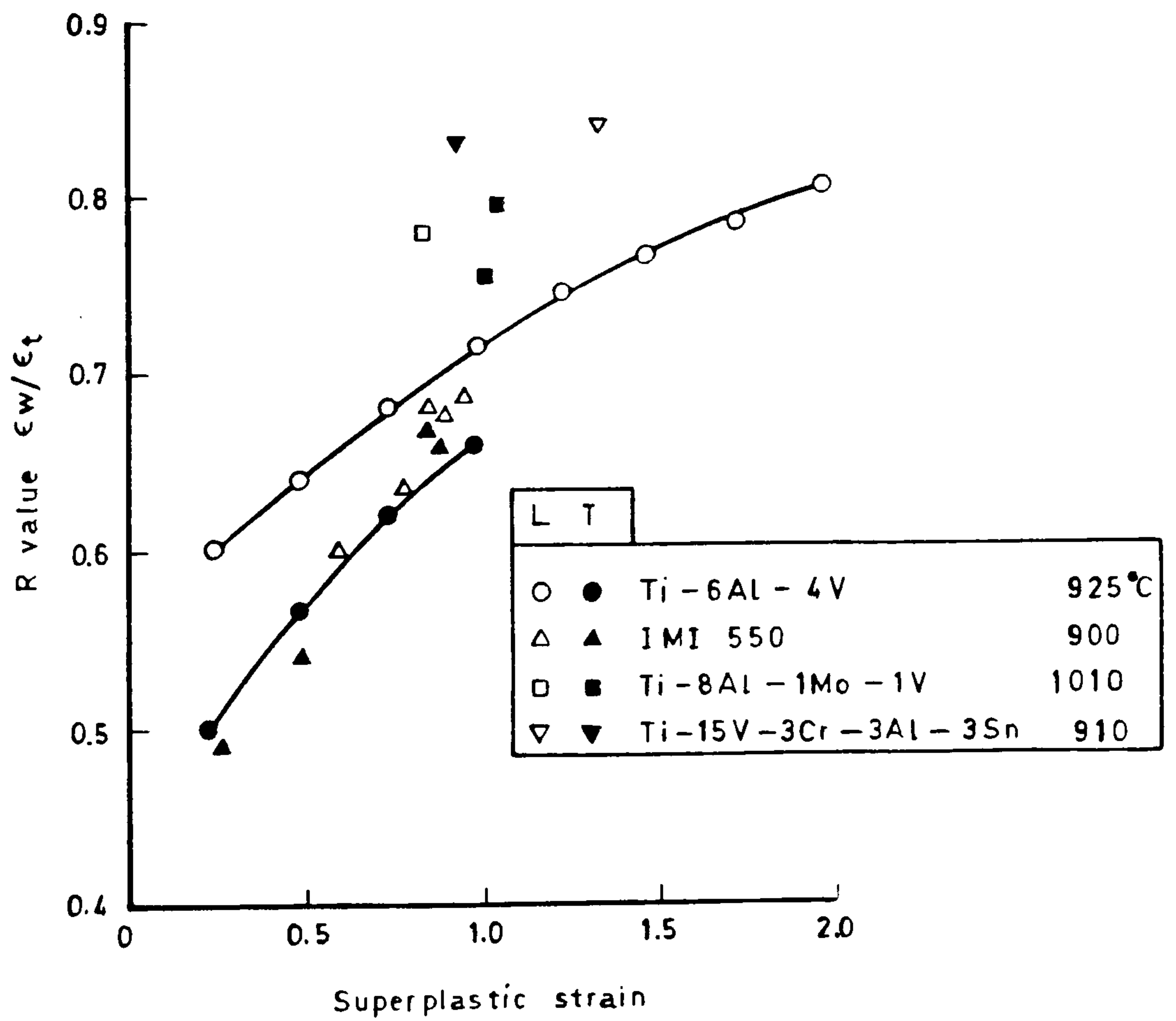


Fig 58 Effect of superplastic strain at $3 \times 10^{-4} \text{ s}^{-1}$ on R values of titanium alloy sheet test pieces

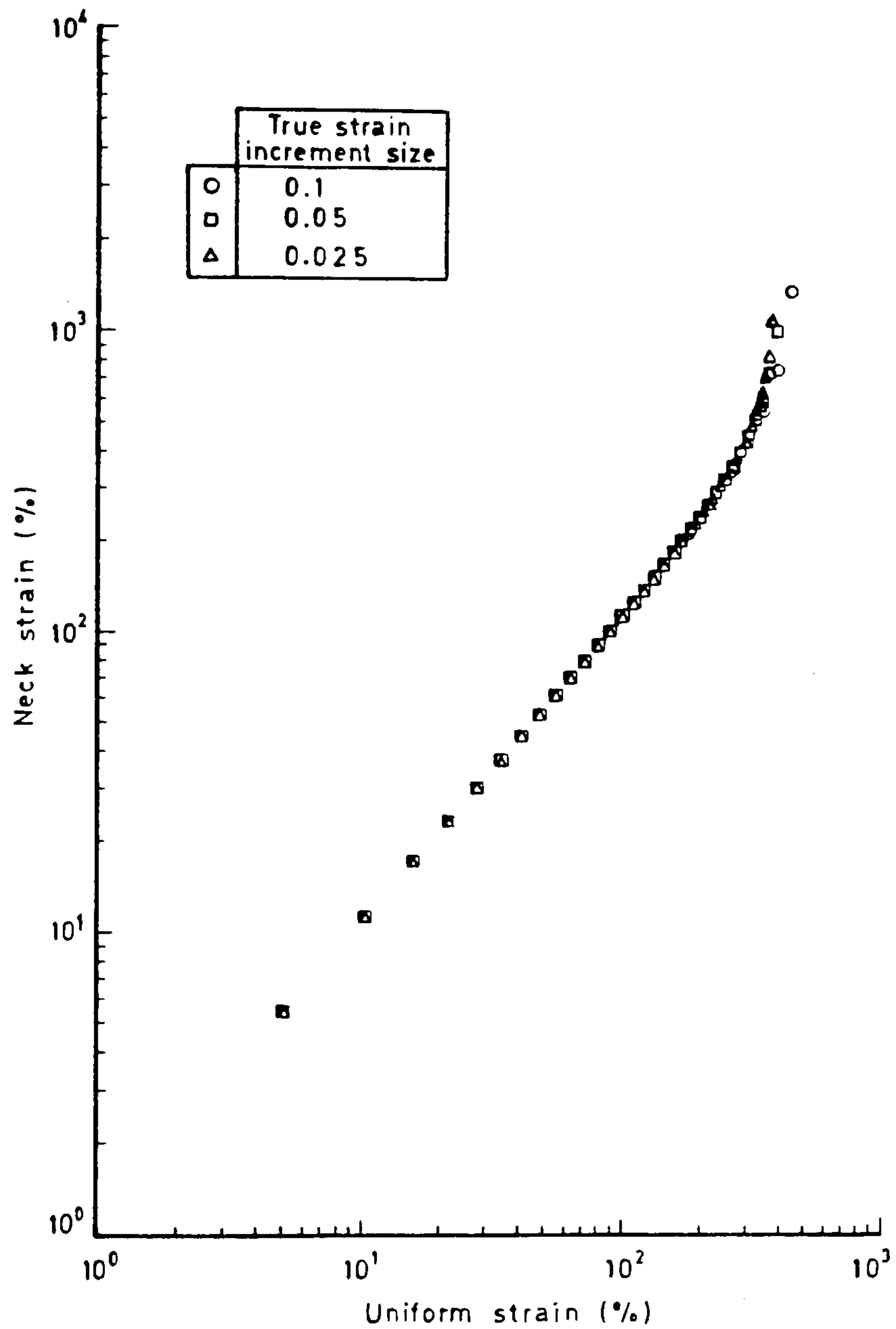


Fig 59 Effect of uniform strain increment size on predicted neck strain. Alternate points have been omitted for the 0.025 strain increment size for uniform strains $< 200\%$. Initial normalised neck area 0.95

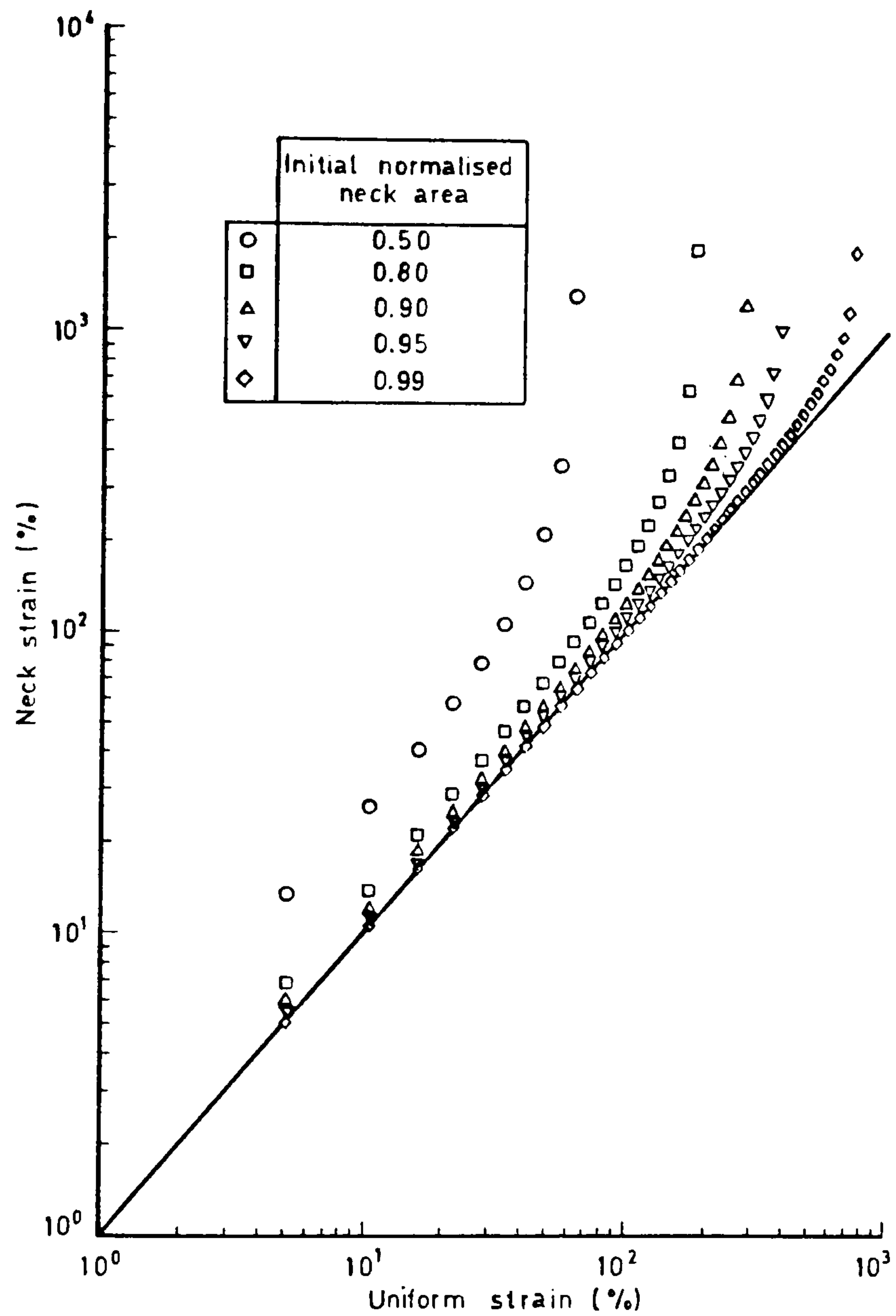


Fig 60 Effect of initial normalised neck area on predicted neck strain. Uniform strain increment size 0.05

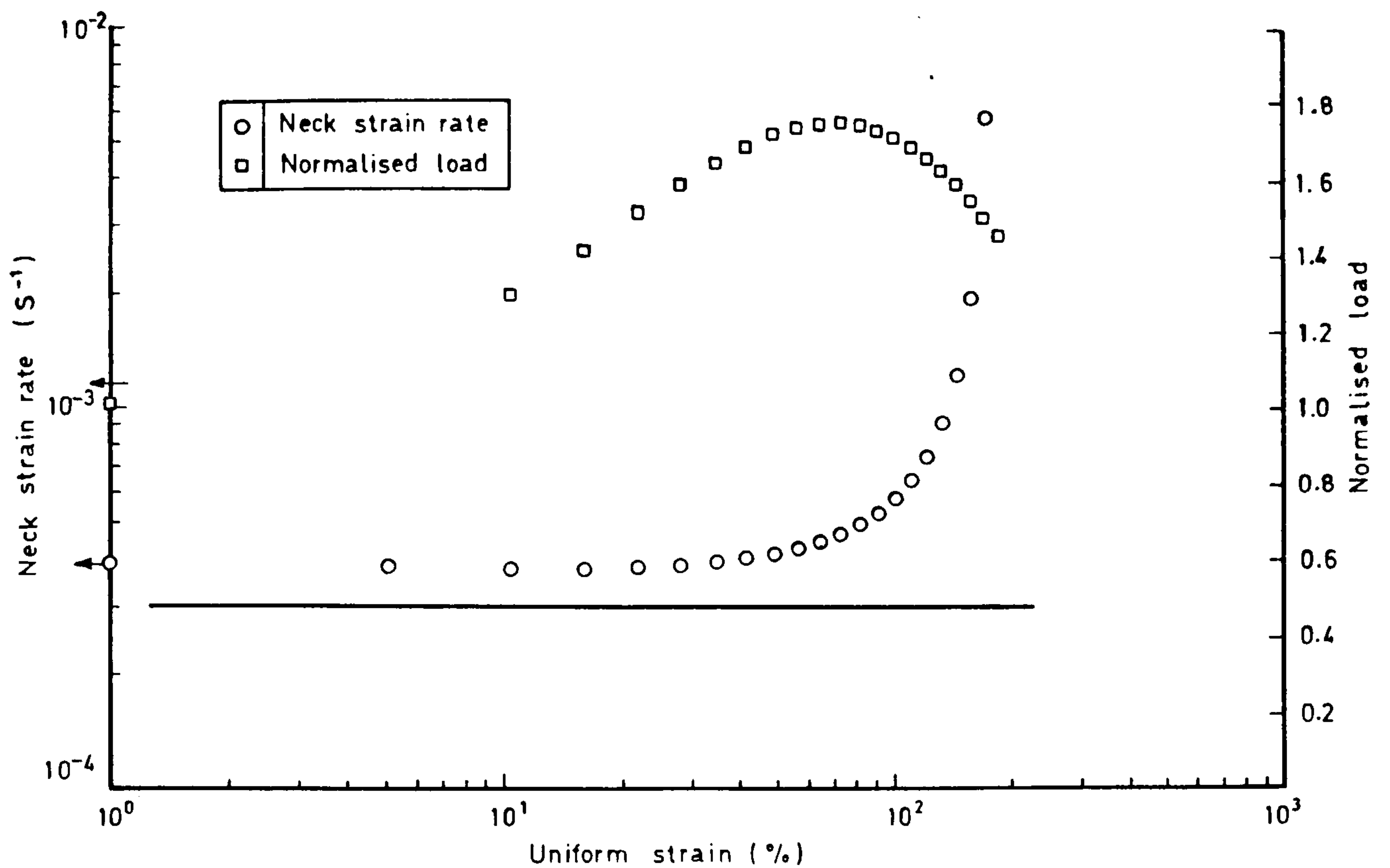


Fig 61 Predicted load and neck strain rate for an initial normalised neck area of 0.8

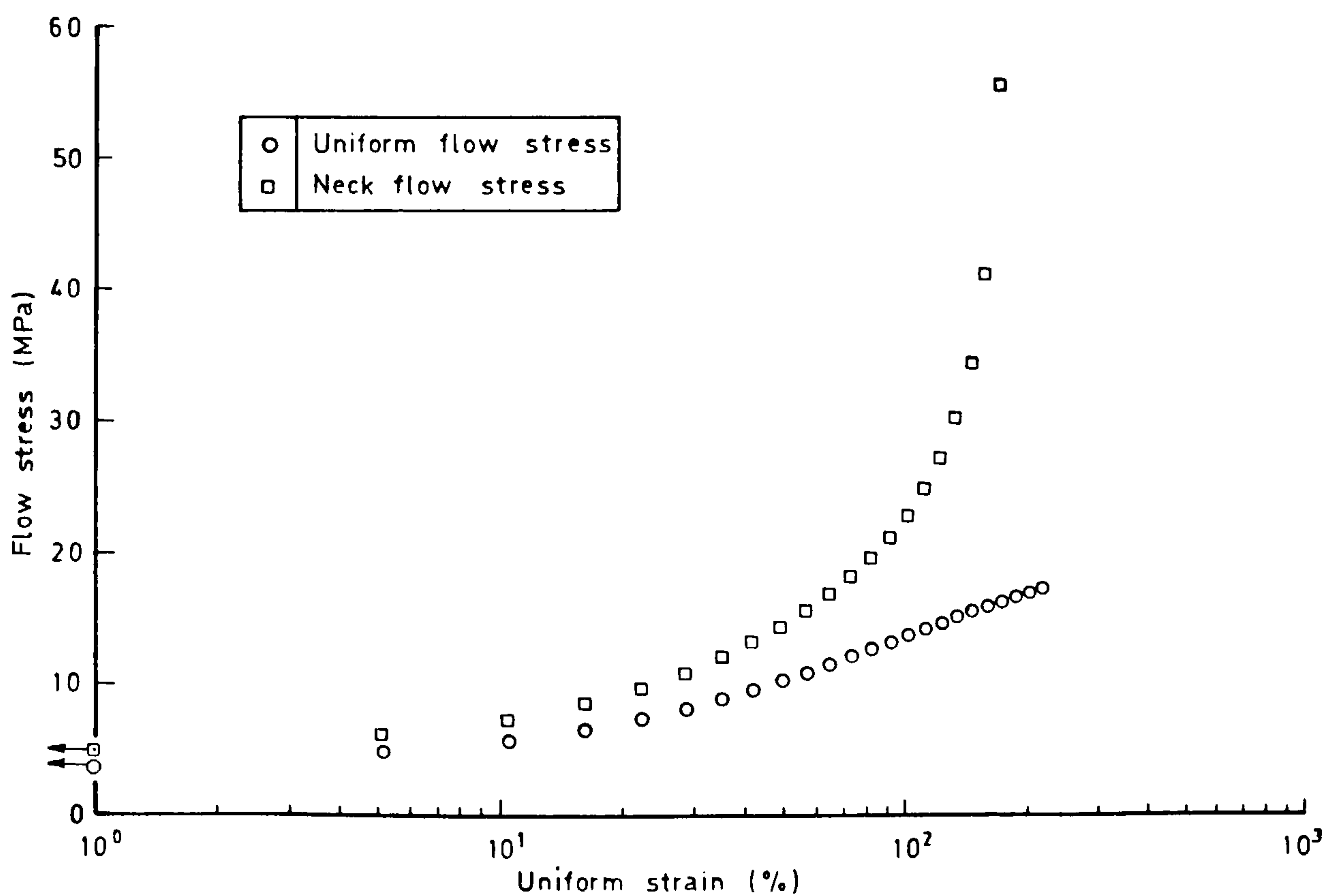


Fig 62 Predicted uniform and neck flow stresses for an initial normalised neck area of 0.8

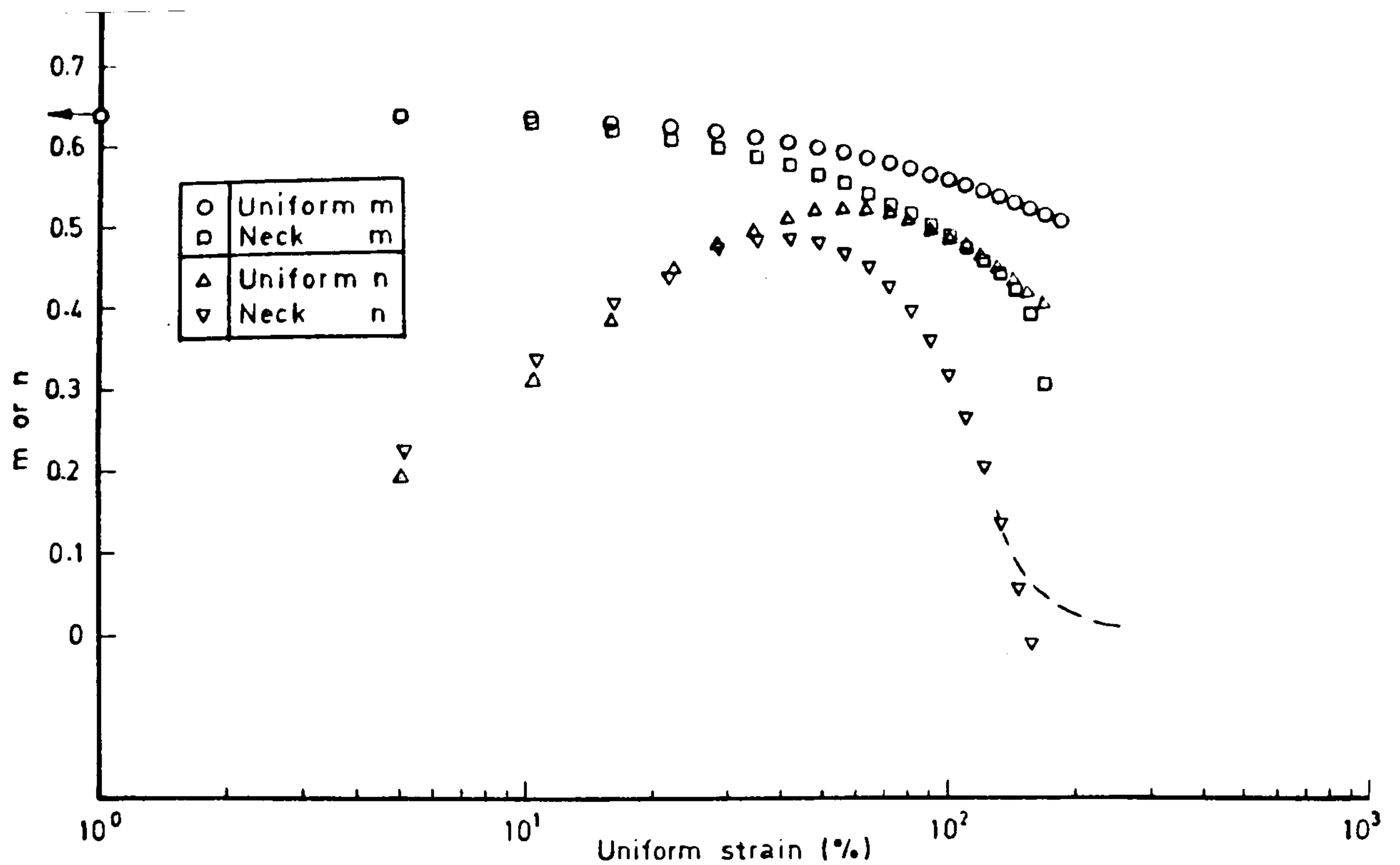


Fig 63 Predicted uniform and neck m and n values for an initial normalised neck area of 0.8

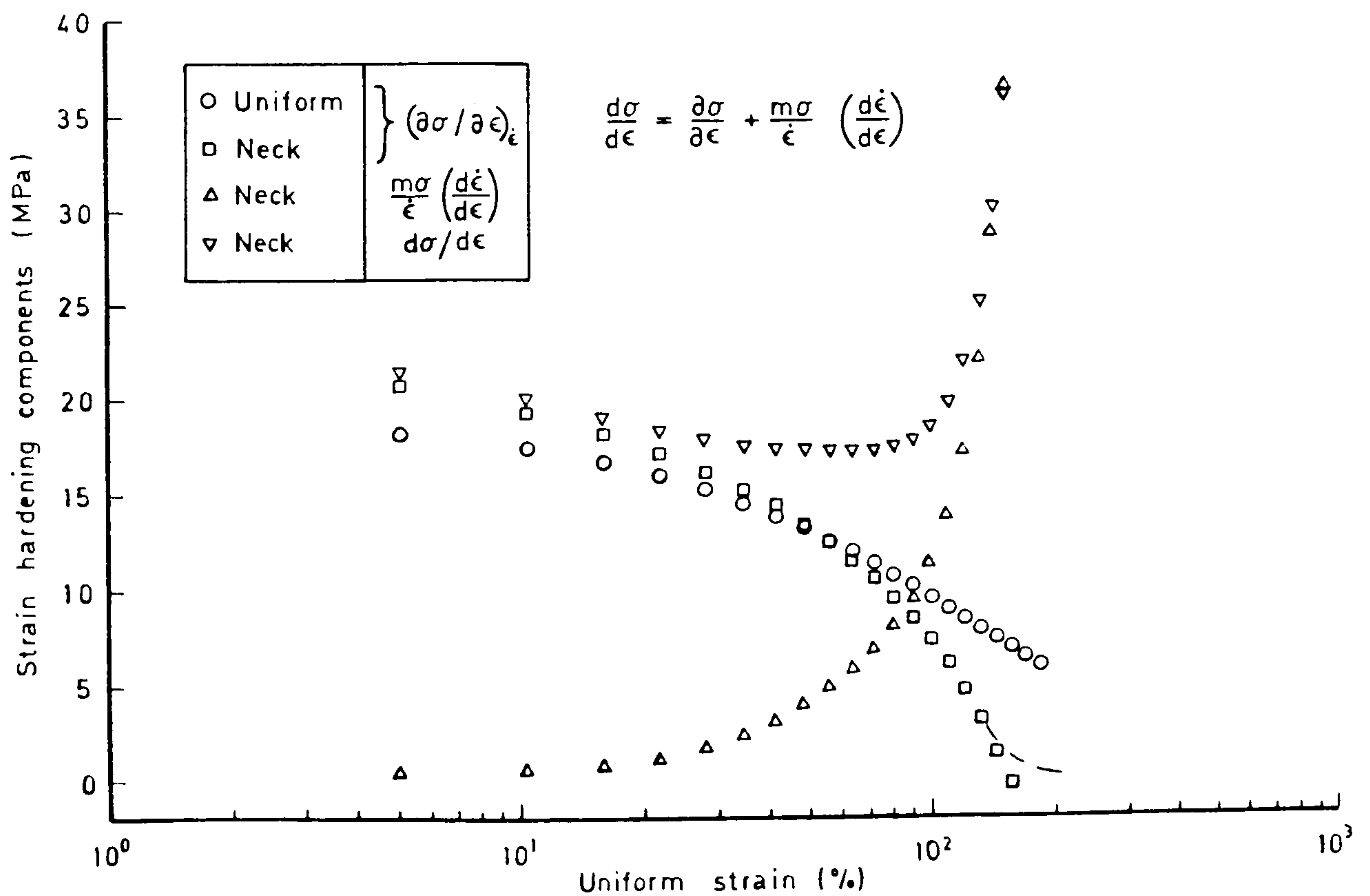


Fig 64 Predicted strain and strain rate hardening for an initial normalised neck area of 0.8

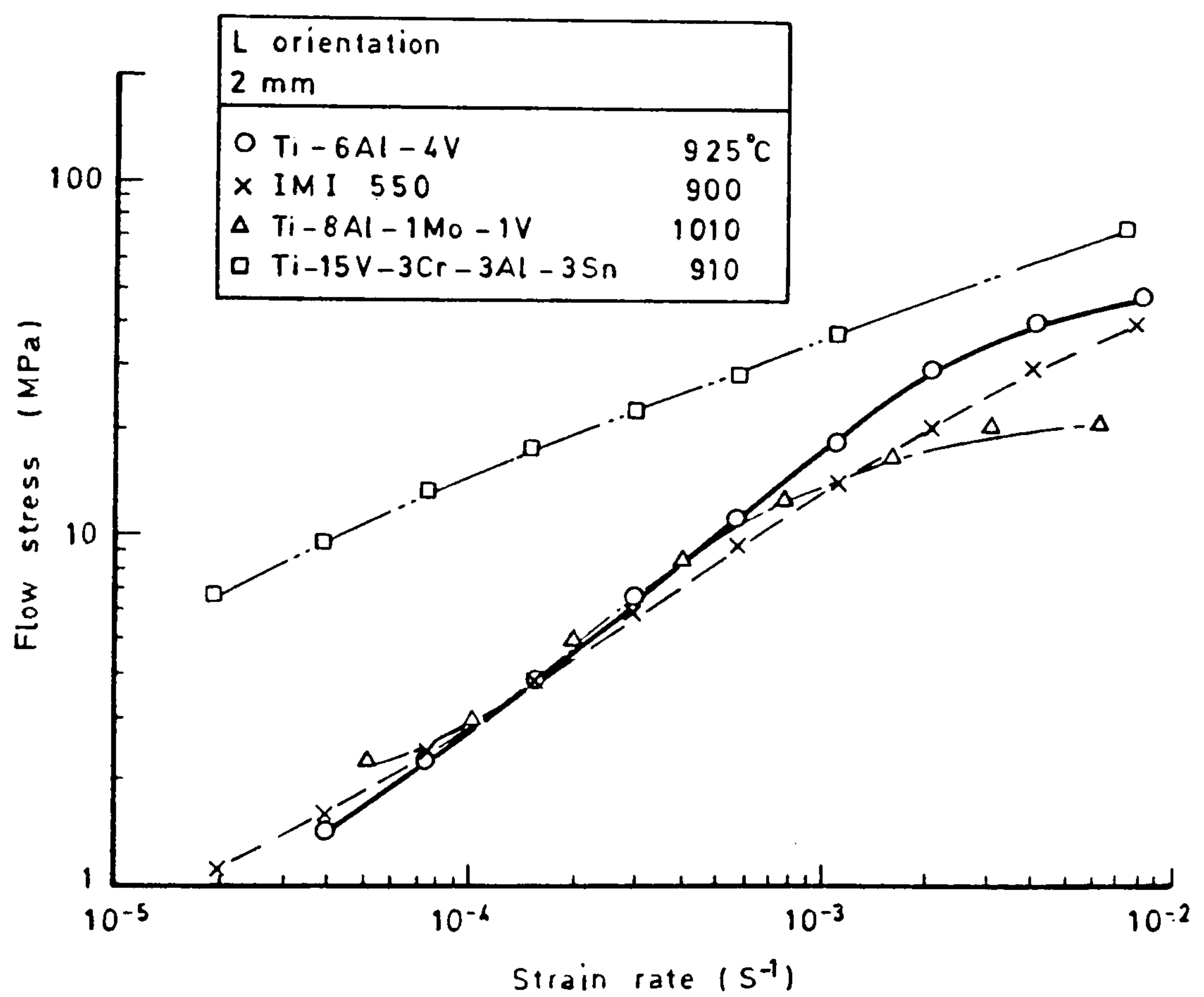
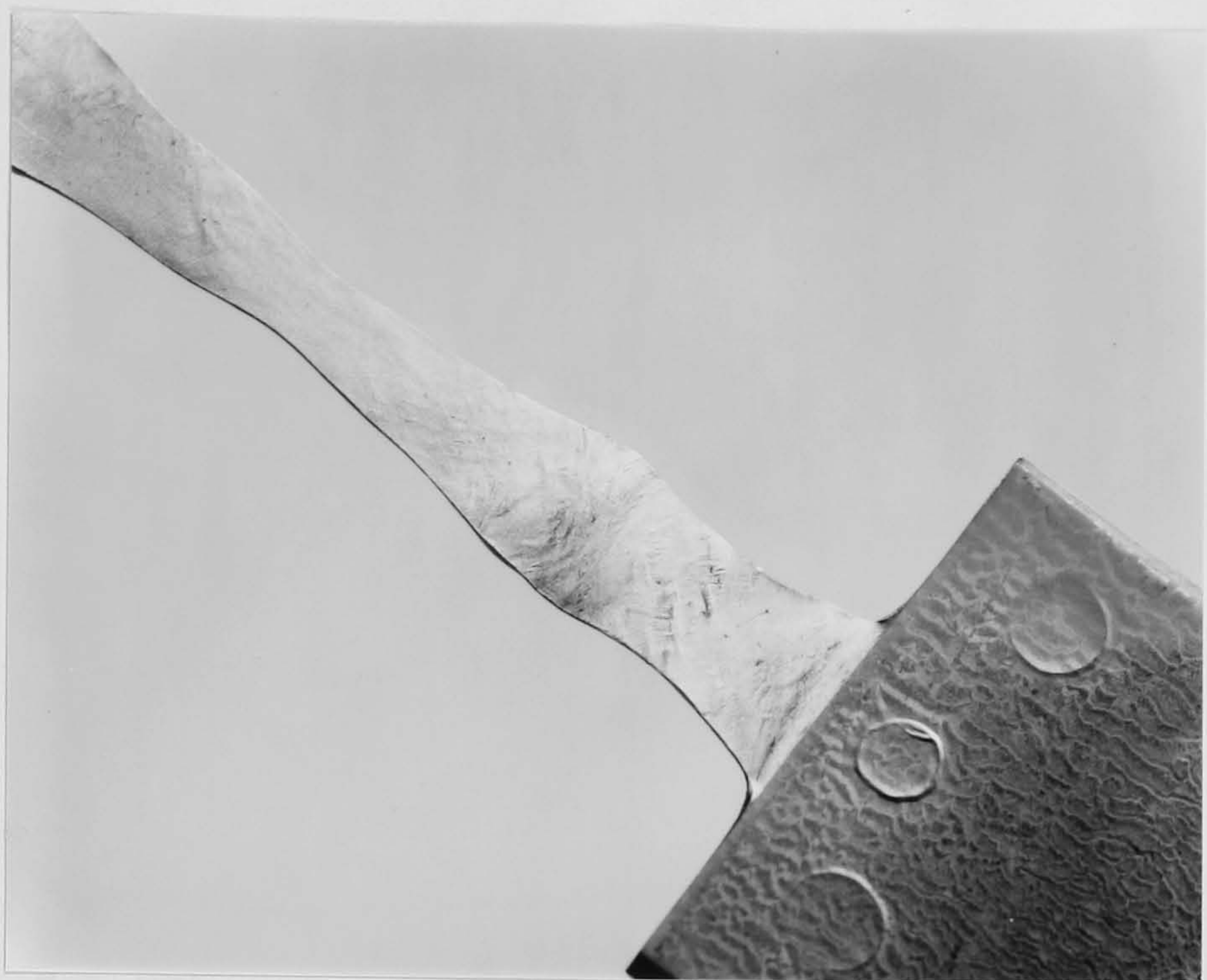
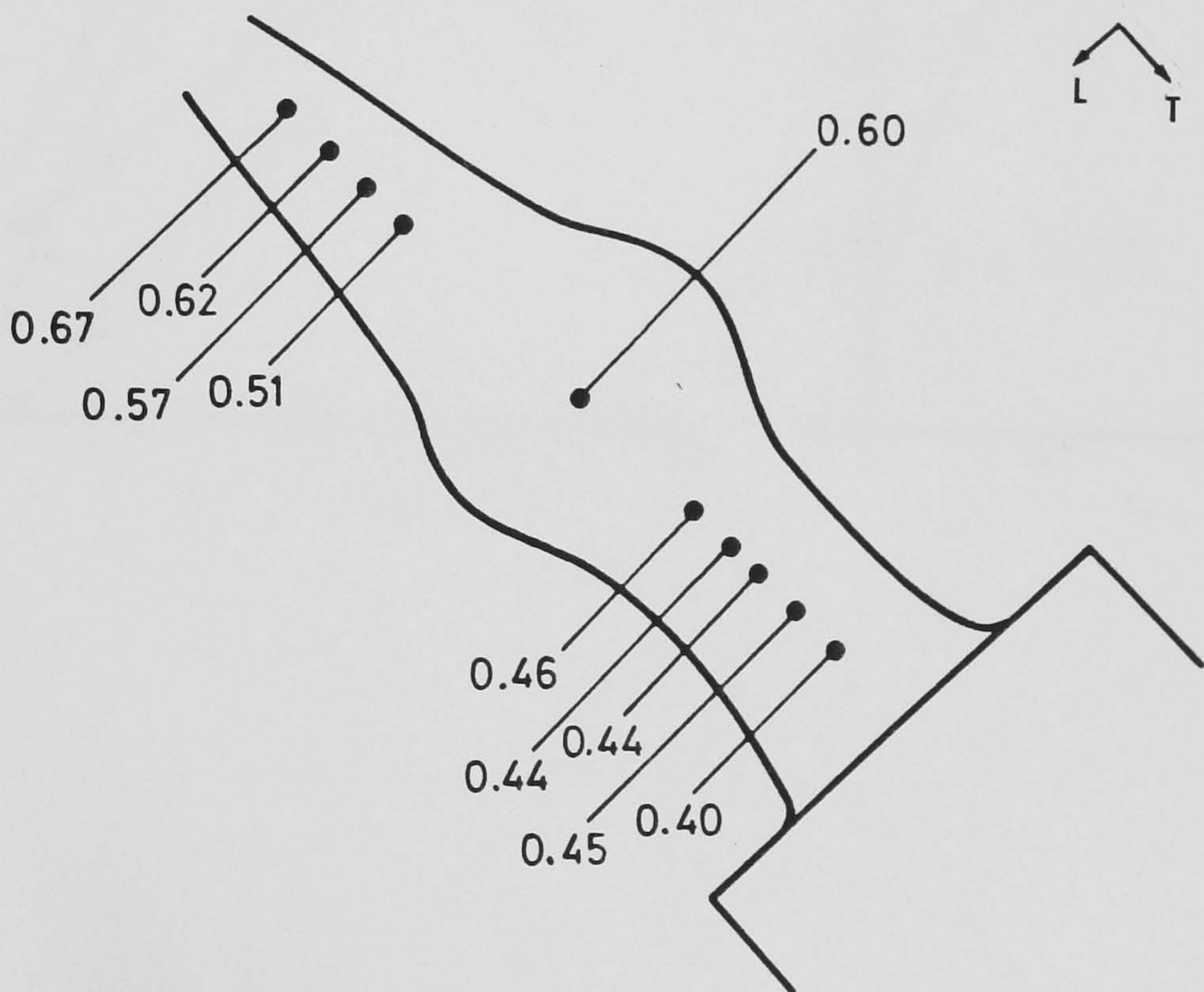


Fig 65 Effect of strain rate on flow stress for each of the alloys



a) Test piece appearance

5 mm



b) R value measurements

Fig 66 TL orientation test piece A9 after strain
at 875 °C and $1.5 \times 10^{-3} \text{ s}^{-1}$

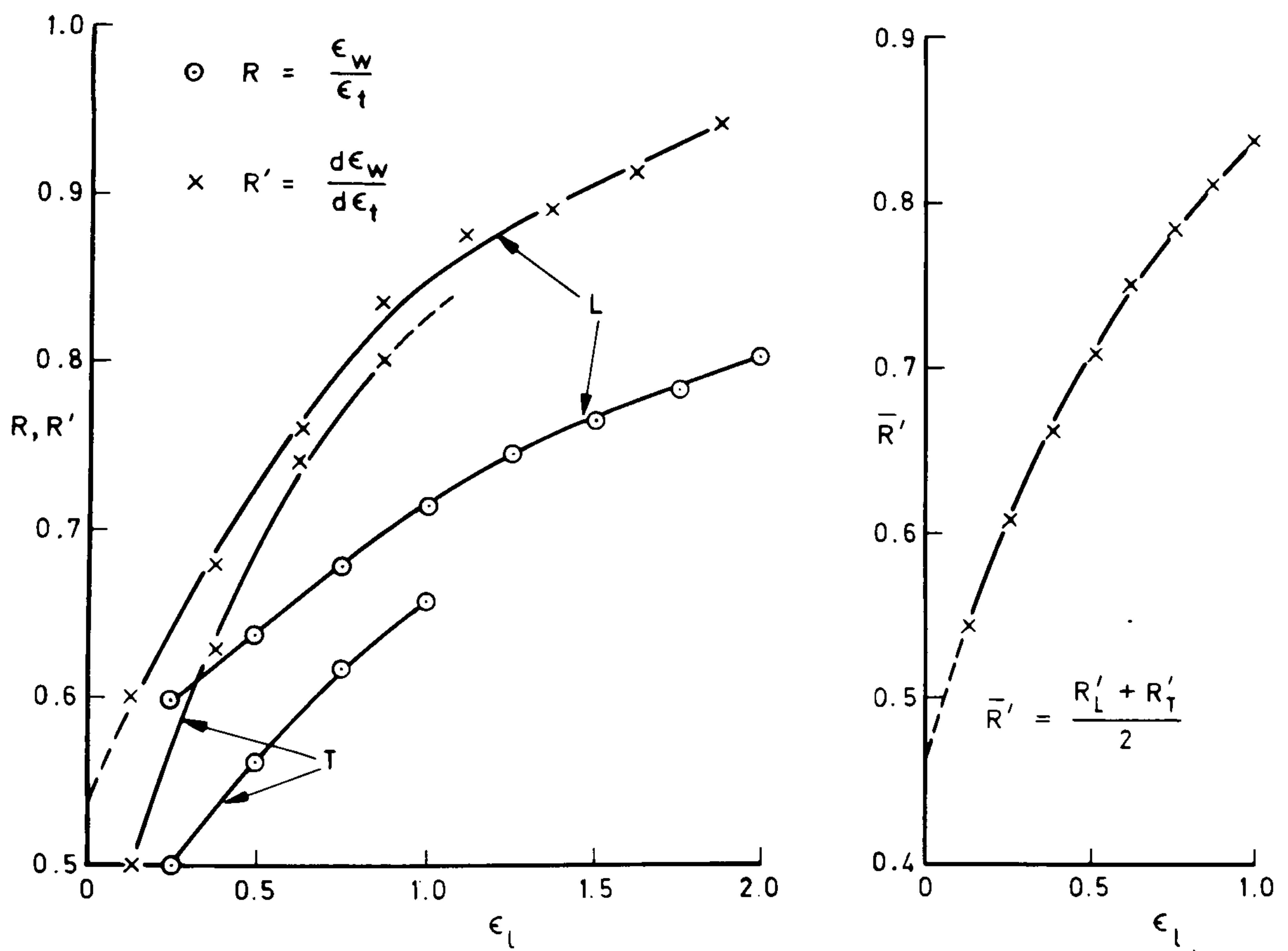


Fig 67 Effect of superplastic strain at 925°C and $3 \times 10^{-4}\text{s}^{-1}$ on strain ratios R , R' and $\bar{R}'$ for 3.3mm Ti-6Al-4V sheet

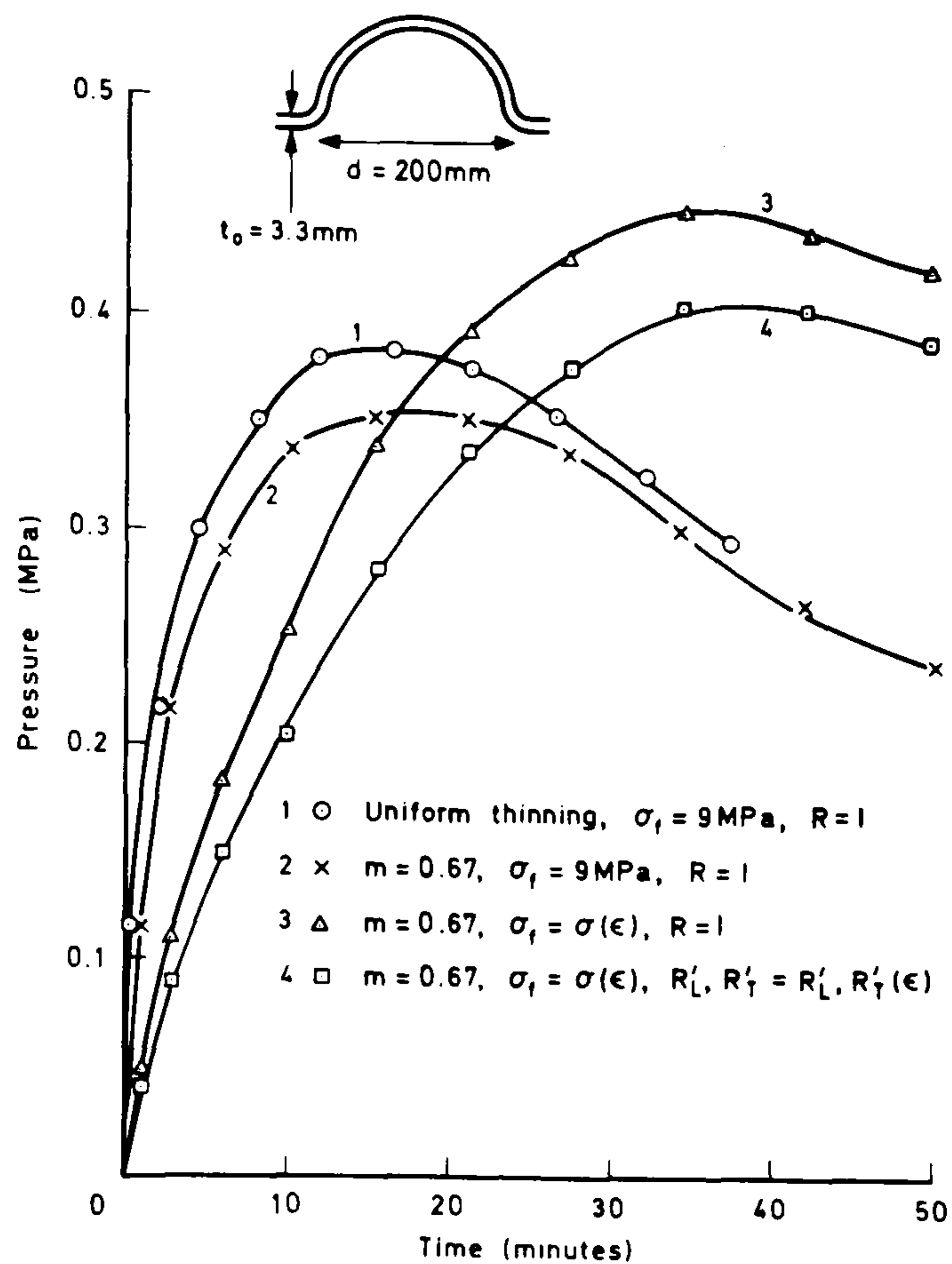


Fig 68 Calculated optimum pressure forming cycles for a 200mm diameter hemisphere from 3.3mm Ti-6Al-4V at 925 °C

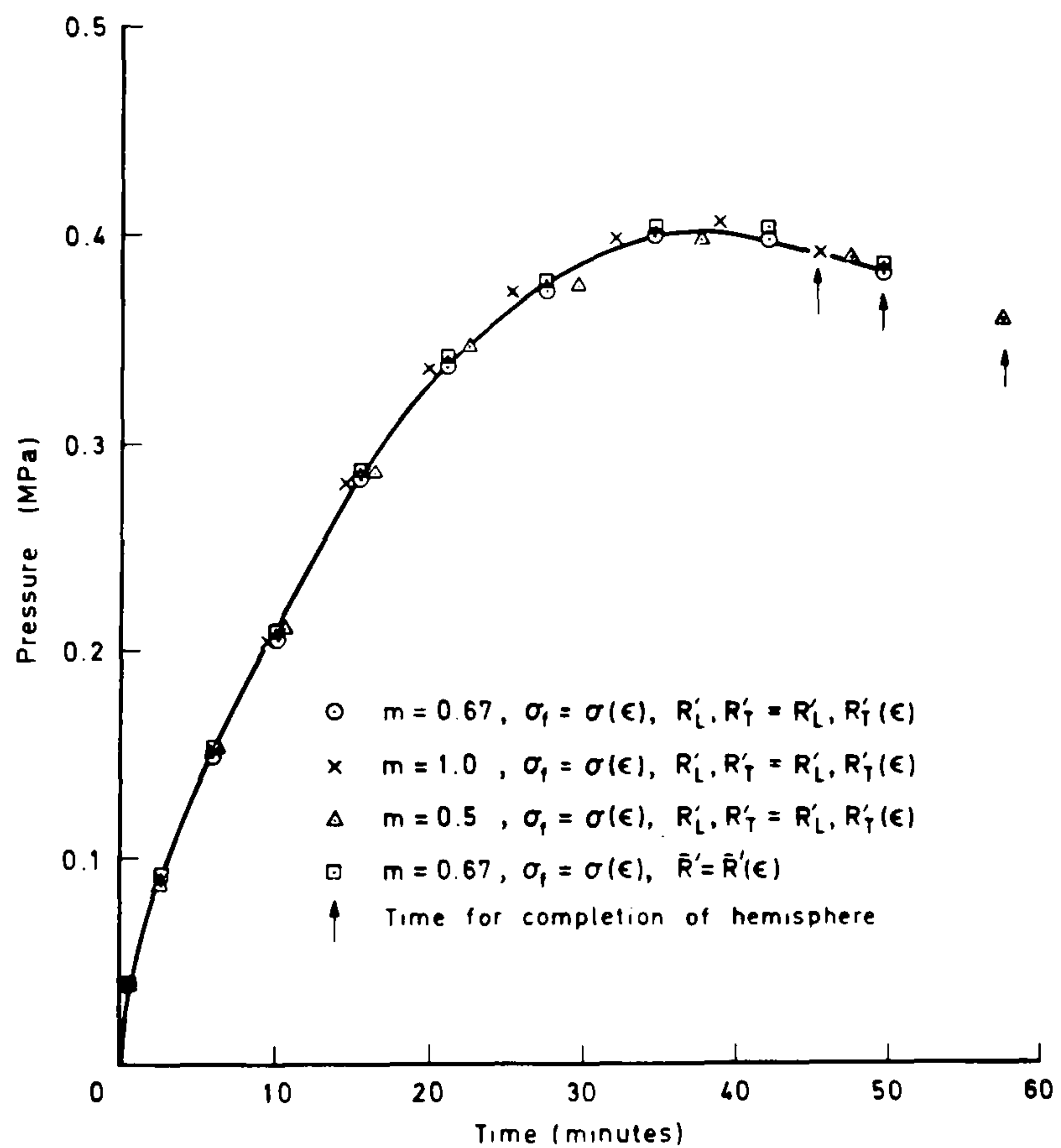


Fig 69 Calculated optimum pressure cycles for various m values

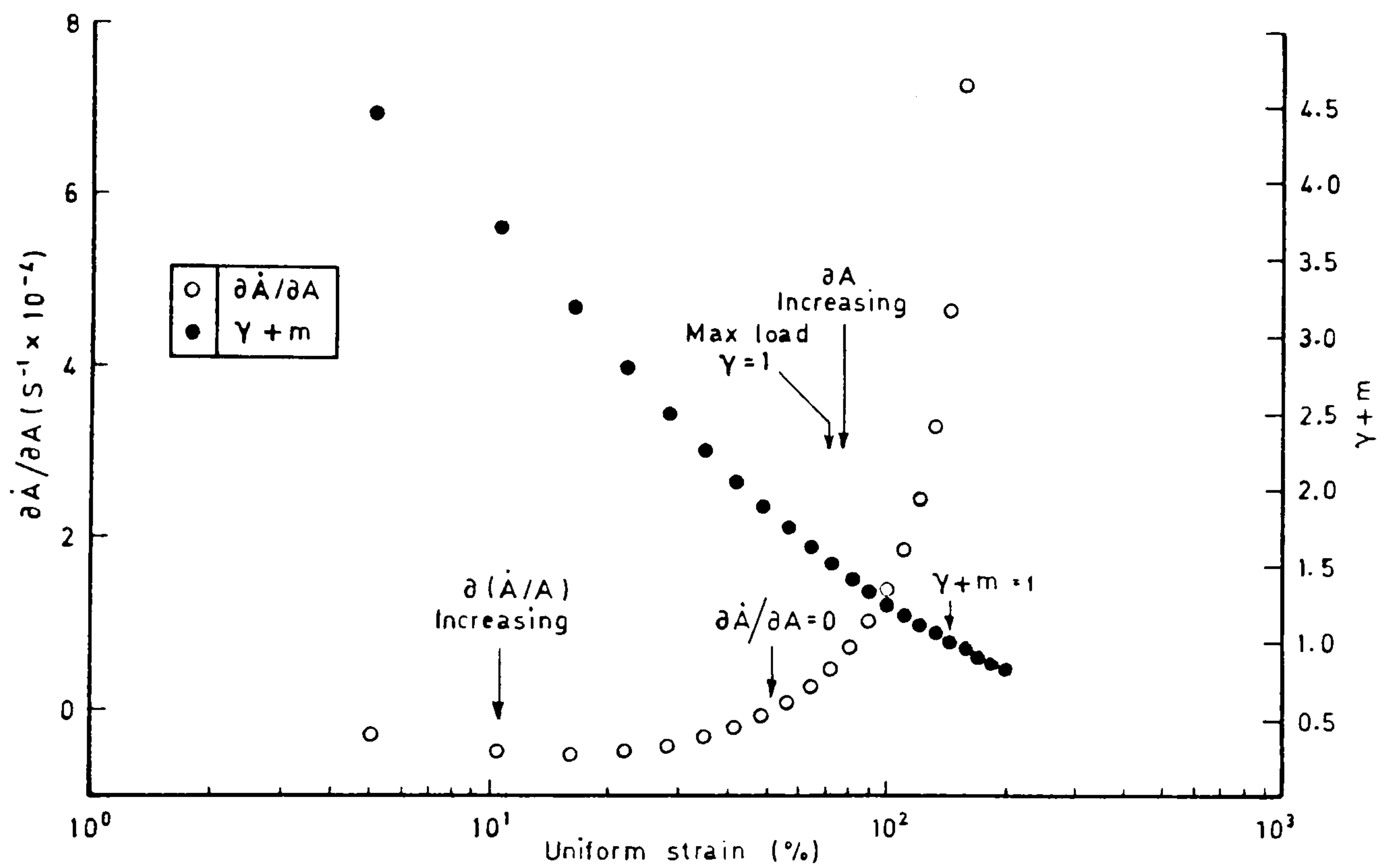


Fig 70 Predicted tensile instability points for
an initial normalised neck area of 0.8