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Recrystallization of Metals during Hot Deformation

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Recrystallization of metals during hot deformation

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[Plate 1]

Recovery processes tend to counteract the effects of work hardening during plastic deformation at high temperatures and at strain rates ranging from those of slow creep to those of rapid hot working operations. However, in metals in which recovery is relatively slow, sufficient stored energy can be accumulated to cause the occurrence of dynamic recrystallization during deformation once a critical strain is exceeded. This process then occurs repeatedly with continued straining. If any metal that has been deformed at high temperatures by a dislocation mechanism is held at temperature after deformation, static recrystallization tends to occur with time.

The effects of dynamic and static recrystallization on microstructure and on the flow stress or creep rate of the metal are considered in this paper and particular attention is given to the range of deformation conditions under which these recrystallization processes are expected to occur.

When metals deform plastically by crystal slip at elevated temperatures, the work hardening produced by deformation tends to be counteracted by recovery processes. These recovery processes cause rearrangement and annihilation of dislocations so that, as strain increases, the dislocations tend to form into two dimensional subgrain walls. In some metals and alloys the recovery entirely balances work hardening, and steady state is achieved and can be maintained to large strains before fracture occurs. In other metals in which recovery is less rapid, certain conditions of stress and temperature of deformation can result in the accumulation of sufficiently high local differences in dislocation density to nucleate recrystallization during deformation. This recrystallization is referred to as *dynamic* recrystallization to distinguish it from the *static* recrystallization that can occur in all metals when deformation is discontinued but the elevated temperature is maintained, or when deformation is carried out at low temperature and the metal is subsequently annealed.

In this paper, both types of recrystallization and their effects on deformation behaviour and microstructure will be outlined and the range of deformation conditions under which they are likely to occur will be considered.

Dynamic recrystallization

The occurrence of dynamic recrystallization was first observed and studied systemmatically during the creep of lead (Greenwood & Worner 1939; Andrade 1948; Gifkins 1958–9). Surveys of work carried out on other metals under creep conditions of constant stress of load (Hardwick, Sellars & Tegart 1961–2) and under constant strain rate deformation conditions (Jonas, Sellars & Tegart 1969) have shown that dynamic recrystallization of initially annealed metals also occurs in nickel, copper, gold, γ-iron, austenitic steels and high purity α-iron, but does not appear to occur in aluminium, zinc, magnesium, tin, low purity α-iron or ferritic steels.

Dynamic recrystallization has also been observed in ice (Steinemann 1958) and has recently been considered as a contributory mechanism in the deformation of quartzites (White 1976).

Effect on deformation behaviour

During creep deformation, the characteristic effect of dynamic recrystallization is to produce one or more transient periods of accelerated creep (Hardwick *et al.* 1961–2). In constant strain rate deformation, analogous fluctuations in flow stress are observed at relatively low strain rates (Rossard & Blain 1958; Jonas *et al.* 1969). The effect of dynamic recrystallization will be illustrated by results from the latter type of tests as these have frequently been carried out in torsion, which enables much higher strains than in tensile creep tests to be attained before fracture processes intervene.

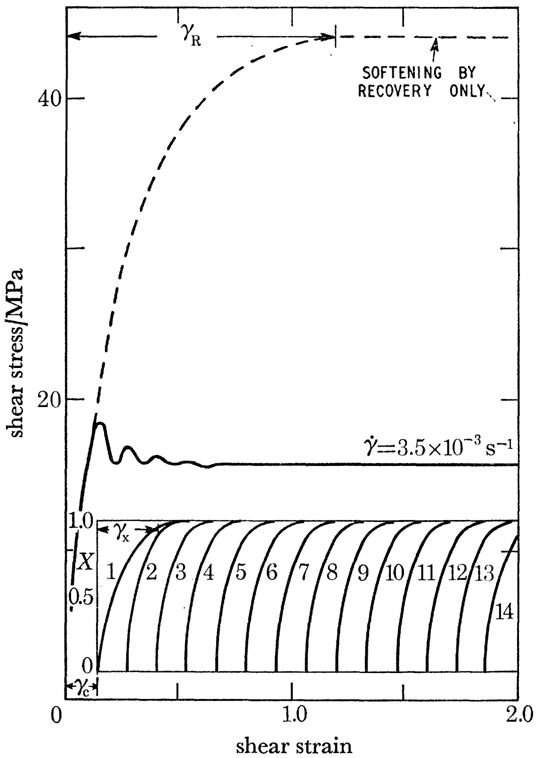

FIGURE 1. Shear stress–shear strain curve for 99.9% nickel deformed in torsion at a surface shear strain rate of 3.5×10^{-3} s^{-1} at 934 °C. The inset shows the cycles of dynamic recrystallization during deformation.

A typical stress–strain curve for nickel at 0.7 absolute melting temperature (Luton & Sellars 1969) is shown in figure 1. The curve rises initially as a result of work hardening and recovery processes, but at a critical strain, γ_c, dynamic recrystallization is nucleated and the curve passes through a maximum and then oscillates several times before settling to a steady state value. The broken curve indicates the stress–strain behaviour expected if recovery were the only operative dynamic softening process. The much higher steady state level is that deduced for the recovery creep mechanism by Ashby (1972, 1973) and is in accord with observations by Weertman & Shahinian (1956) on lower purity nickel in which dynamic recrystallization did not occur at this temperature. The important additional softening effect of dynamic recrystallization is thus clear, when the critical strain for recrystallization, γ_c, is much less than the strain expected for the onset of steady state by recovery softening only, γ_R.

The inset on this figure shows the recrystallization curves from which the observed stress–strain behaviour can be computed (Sah 1971). When the shear strain reaches γ_c, new strain-free grains are nucleated and with increasing strain (time) recrystallization proceeds along curve 1 leading to a decrease in flow stress. However, the concurrent deformation causes work hardening of these grains so that as the rate of recrystallization falls, the flow stress passes through a minimum and starts to rise again. When γ_c is again reached in the grains that recrystallized earliest in the first cycle, these grains recrystallize again along curve 2, leading to a second maximum in flow stress. This process is repeated each time γ_c is reached in the recrystallized grains, leading to further oscillations in flow stress until the process becomes sufficiently out of phase in different local regions of the material to make recrystallization effectively 'continuous'. This results in an overall steady state flow stress.

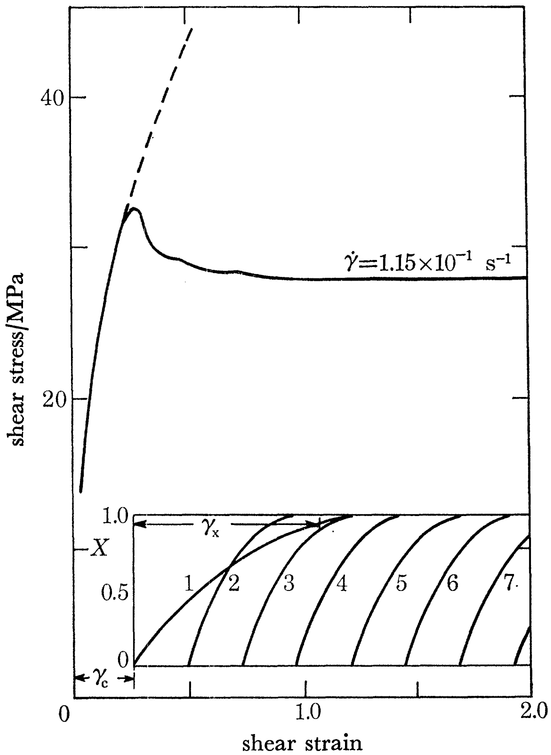

FIGURE 2. Shear stress–shear strain curve for 99.9% nickel deformed in torsion at a surface shear strain rate of 1.15×10^{-1} s^{-1} at 934 °C. The inset shows the cycles of dynamic recrystallization during deformation.

At higher strain rates, only one maximum in flow stress is observed (figure 2), but as shown by the inset, this again arises from the repeated occurrence of recystallization. The difference in this case is that the strain interval over which the first cycle of recrystallization occurs, γ_x, is now considerably larger than γ_c so that several cycles of recrystallization overlap, giving effectively 'continuous' recrystallization and a smoother fall in flow stress to steady state. It was originally suggested (Luton & Sellars 1969) that the changeover from initially 'periodic' to initially 'continuous' recrystallization, with the accompanying change in form of stress–strain curve, would occur when $\gamma_x \approx \gamma_c$. Developments of the early model (Sah 1971) to allow for differences in recrystallization kinetics in the first and subsequent recrystallization cycles, which are expected to arise from differences in grain size, show that the changeover probably occurs when $\gamma_x \approx 2\gamma_c$.

Microstructural changes

During deformation up to γ_c, work hardening and recovery lead to the development of a dislocation subgrain structure, but typically dislocations in the subgrain boundaries remain tangled (figure 3, plate 1), rather than forming the clean two dimensional networks observed in metals in which recovery is more rapid (Jonas *et al.* 1969). The existance of these higher energy tangled subgrain boundaries has been shown (Sandström & Lagneborg 1975) to be essential to obtain sufficient stored energy differences in local regions to nucleate dynamic recrystallization.

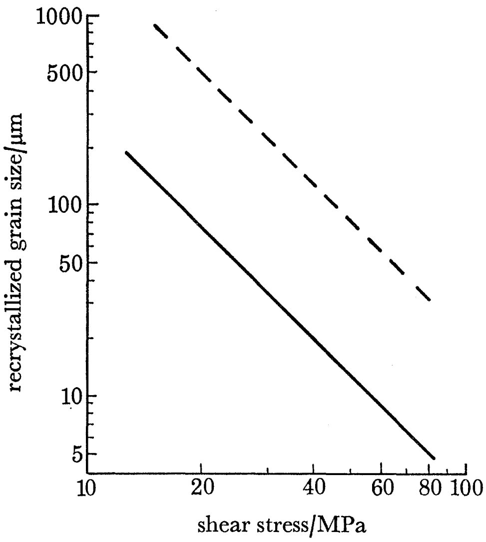

FIGURE 5. Stress dependence of recrystallized grain size produced by dynamic recrystallization (solid line) and by static (metadynamic) recrystallization (broken line) of the dynamically recrystallized structure in nickel (after Sah *et al.* 1974).

Optical microscopy reveals the progressive development of the recrystallized structure with increasing strain beyond γ_c (figure 4, plate 1), until it has entirely replaced the original grain structure (recrystallization 95 % complete at γ_x). With further strain there is no further apparent change in grain structure, and the grains always appear nearly equiaxed as a result of the repeated occurrence of dynamic recrystallization creating new grains of a size determined by the steady state deformation conditions. This contrasts with the situation in metals which do not recrystallize and in which the original grains become progressively more elongated with increasing strain.

Measurements of dynamically recrystallized grain size have been carried out on nickel (Luton & Sellars 1969; Sah, Richardson & Sellars 1974), copper (McQueen & Bergerson 1972; Bromley & Sellars 1973) and α-iron (Glover & Sellars 1973). In all cases it was found that the grain size is uniquely determined by the stress, independent of the temperature of deformation. As illustrated in figure 5, the relation between recrystallized grain size (d) and stress (τ) may be represented by an equation of the form

$$\tau \propto d^{-n}, \tag{1}$$

where n is a constant of value between $\frac{1}{2}$ and 1. The exact value is difficult to ascertain experimentally and it may vary with purity of the metal. Certainly, the level of the curve depends on

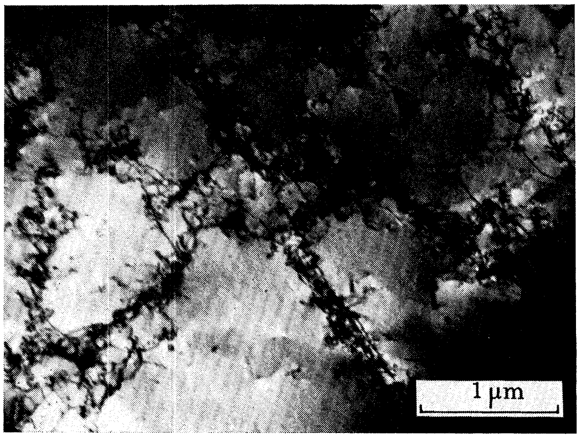

FIGURE 3. Electron micrograph of the subgrain structure in pure nickel deformed in creep at 20.7 MPa and 800 °C to a strain less than the critical value for dynamic recrystallization (Richardson *et al.* 1965).

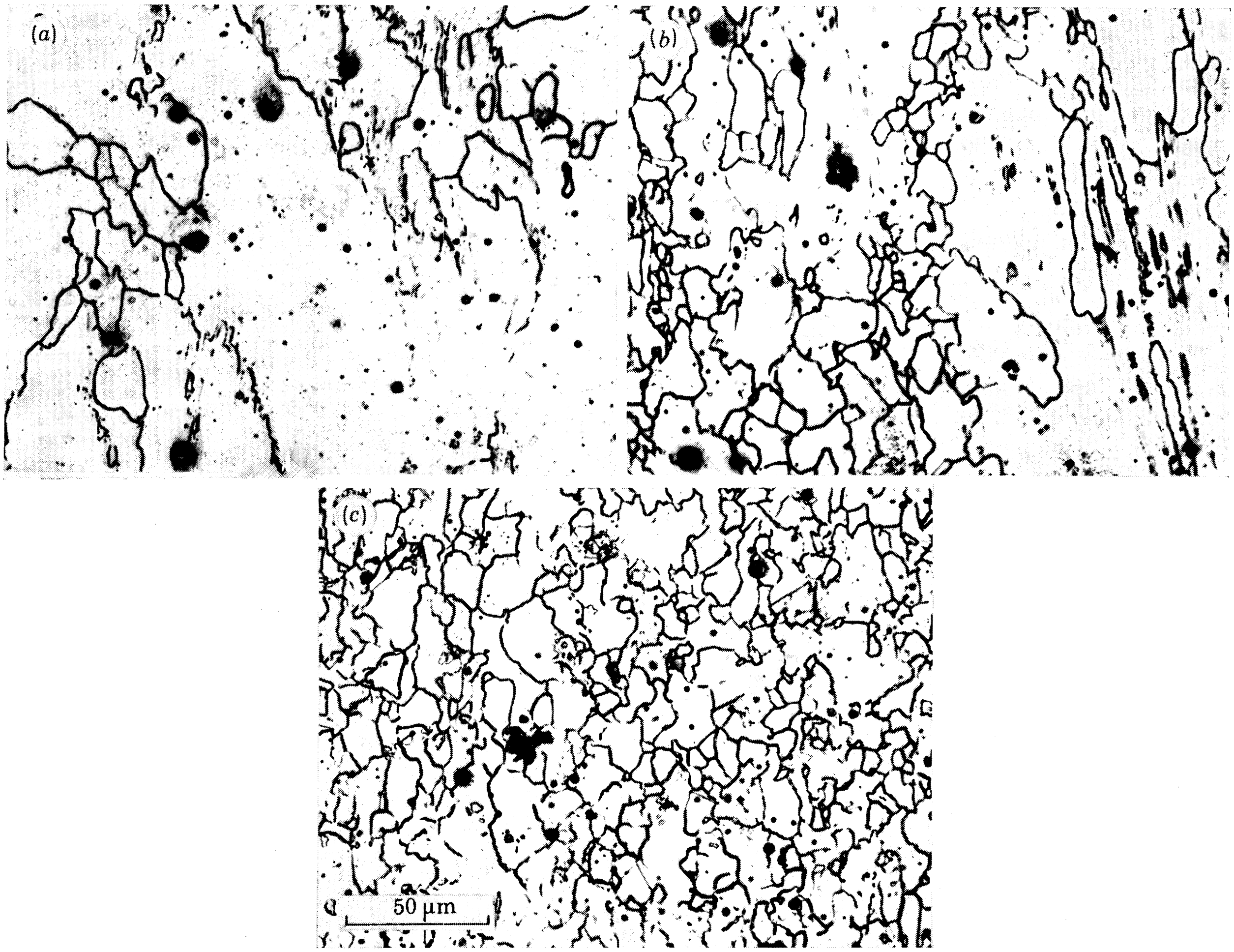

FIGURE 4. Optical micrographs of dynamically recrystallized grains developed in nickel deformed in torsion at a surface shear strain rate of $6.6 \times 10^{-2}\ s^{-1}$ at 880 °C to surface shear strains of (*a*) 1.3, (*b*) 4.7 and (*c*) 8.8. Critical shear strain for dynamic recrystallization 1.0 (Sah 1971).

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impurity or alloy content (Luton & Sellars 1969; Bromley & Sellars 1973; Glover & Sellars 1973), increasing levels giving a smaller grain size at a given flow stress. The dynamically recrystallized grain size does, however, appear to be independent of the original grain size of the material, although this influences the kinetics of the first cycle of recrystallization (Sah *et al.* 1974). Dynamic recrystallization, particularly under deformation conditions leading to high flow stresses, is therefore a very potent mechanism of grain refinement.

Electron microscope observations of dynamically recrystallized grains reveal that they contain tangled dislocation substructures analogous to those observed at strains less than γ_c (Luton & Sellars 1969; McQueen & Bergerson 1972). This distinguishes them from statically recrystallized grains and is to be expected as the grains are deformed as they develop.

It appears probable that, at least at fairly high strain rates, the concurrent deformation destroys the driving force for growth after an initial burst of growth from the nucleus (Sah *et al.* 1974). Recrystallization then proceeds by the continual formation of new nuclei and their restricted growth. This mechanism differs from that for classical (static) recrystallization, in which nucleation takes place initially and recrystallization proceeds by continued growth from these nuclei until impingement of the growing grains occurs.

The repeated nature of dynamic recrystallization means that during steady state the degree of development of the dislocation structure is heterogeneous and in different local regions there will be grains which have just recrystallized, grains which have been deformed by a further increment of γ_c and are just about to recrystallize again, and a spectrum of grains between these two limiting conditions.

This heterogeneity of substructure again contrasts with the uniform subgrain structure that is present during steady state when recovery is the only dynamic softening process. Under high-temperature deformation conditions, dynamic recovery involves subgrain boundary migration or 'repolygonization' which maintains an equiaxed subgrain structure with misorientations that reach a steady state value of only a few degrees (Jonas *et al.* 1969; Warrington 1976). However, in lower temperature deformation to high strains, there is evidence that subgrain misorientations may continue to increase with strain until they effectively form grain boundaries (Cairns, Clough, Dewey & Nutting 1971; Nutting 1974). The structure then has the appearance of being recrystallized, although only recovery has taken place and no separate nucleation and growth events are involved. This mechanism, which may be considered as 'recrystallization *in situ*' will not be discussed further in this paper.

Deformation conditions for dynamic recrystallization

From the combined observations on nickel deformed in compression creep and in torsion (Richardson, Sellars & Tegart 1965; Luton & Sellars 1969) the critical strain for the onset of recrystallization, γ_c, is found to pass through a minimum as a function of stress (figure 6). The reasons for this are not yet fully understood, but the rising curve at low stresses appears to be associated with the decrease in dislocation density with decreasing stress. Higher strains are then necessary to give sufficient misorientation, and hence dislocation density, in the subgrain boundaries to provide the stored energy for nucleation. The rising curve at high stresses appears to result from the need for increasing stored energy with increasing strain rate to ensure that boundary migration is sufficiently rapid for growth of the nuclei to occur at all before the dislocation density behind the moving boundary has been increased sufficiently by concurrent deformation to destroy the initial driving force.

Over the same range of stress, the strain that takes place during recrystallization, γ_x, increases continuously. This leads to the changeover from periodic to 'continuous' recrystallization with increasing stresses.

The expected strain at which steady state would be obtained by recovery softening only, γ_R, is also shown by the broken curve in this figure. Observations on α-iron (Glover & Sellars 1973) have shown there to be an upper stress limit for the occurrence of dynamic recrystallization and have suggested that this occurs when $\gamma_c \approx \gamma_R$. This would arise because, when a steady state structure is obtained by dynamic recovery, the stored energy no longer rises with

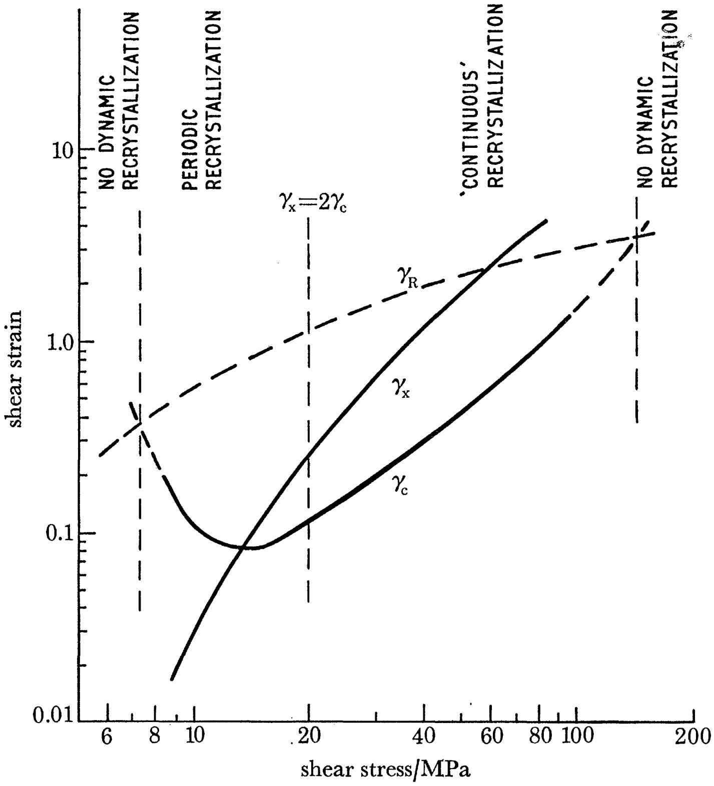

FIGURE 6. Relation between the critical strain γ_c for the onset of dynamic recrystallization, the strain interval γ_x for a large fraction of dynamic recrystallization to take place and the strain γ_R expected for the onset of steady state if softening were by recovery only, in 99.9% nickel of initial grain size 200 μm at *ca.* 930 °C.

strain so that if conditions for nucleation are not met at strains less than γ_R, they are unlikely to be met at higher strains. A cut-off for the occurrence of dynamic recrystallization at low stresses has also been observed during creep of impure nickel at 1100 °C and a shear stress of about 7 MPa (Weertman & Shahinian 1956). It is suggested in figure 6 that it also arises when $\gamma_c \approx \gamma_R$.

From diagrams of the form shown in figure 6 for different temperatures, the limits of dynamic recrystallization in pure nickel of initial grain size 200 μm have been deduced and are superimposed on an Ashby (1972, 1973) deformation map in figure 7. The lower temperature limit is uncertain, but has been made to be consistent with that observed on lower purity nickel of fine grain size (Jenkins, Digges & Johnson 1954).

From figure 7 it can be seen that the stress limits fall well within the area where recovery creep deformation is expected. These limits are expected to be sensitive to both alloy content and initial grain size as both influence γ_c (Gifkins 1958–9; Luton & Sellars 1969; Bromley & Sellars 1973; Glover & Sellars 1973; Sah *et al.* 1974). The lowest possible stress limit even with high purity materials of fine initial grain size will, however, be given by the lower bound of the area for recovery creep, which is itself sensitive to grain size (Ashby 1972, 1973), as diffusional flow does not involve dislocation movement and therefore cannot generate a driving force for recrystallization.

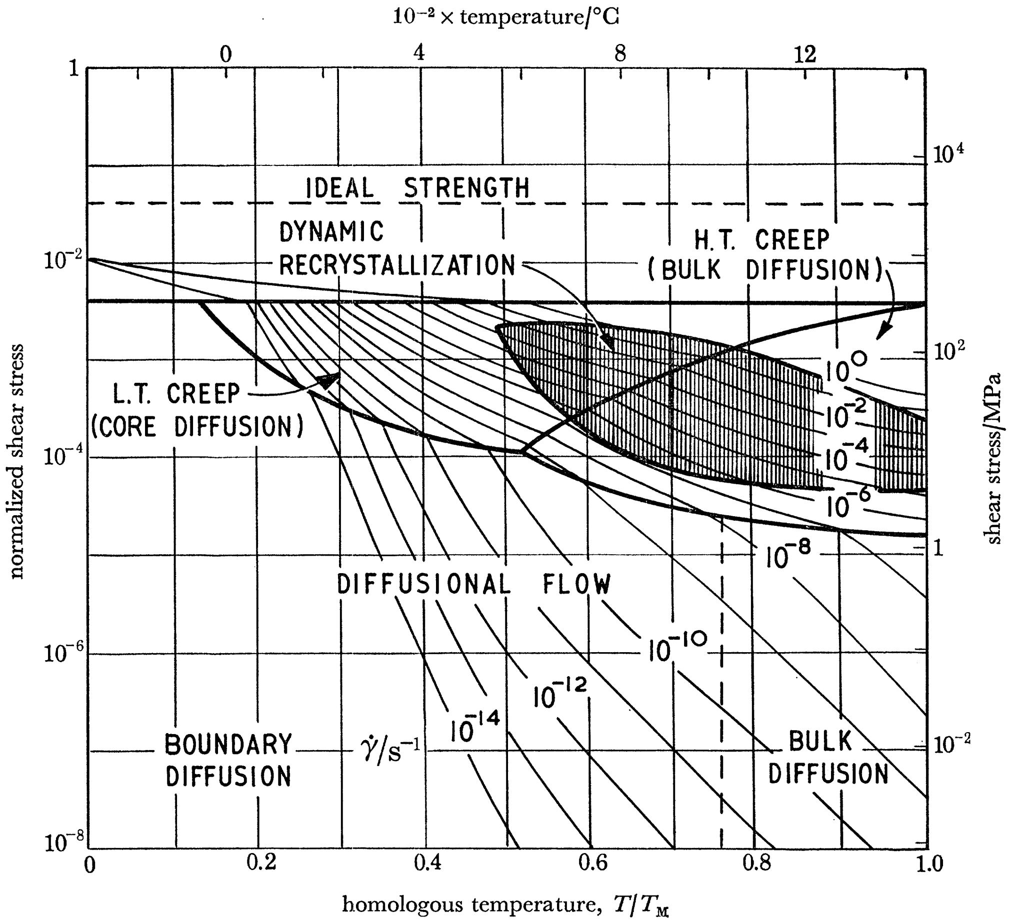

FIGURE 7. Ashby deformation map for 99.9% nickel of initial grain size 200 μm showing the limiting conditions for the occurrence of dynamic recrystallization.

Within the area where dynamic recrystallization does occur, deformation tends to be heterogeneous (Cottingham 1966; Sah *et al.* 1974). The softer recrystallized grains formed first cause concentration of strain within them which leads to favourable conditions for nucleation in adjacent regions where a strain rate gradient exists. This gives conditions for instability and localization of strain as recrystallization proceeds.

Static recrystallization

If deformation is discontinued but the material is retained at high temperature, the structure undergoes further change as a function of time by the occurrence of static recovery, recrystallization and grain growth.

These structural changes are reflected by softening of the material as a function of time after deformation. This is shown clearly by a change in form of the stress–strain curve obtained if deformation is continued after different rest periods, as shown in figure 8, which is taken from current work on austenitic stainless steel deformed in plane strain compression. Softening proceeds until recrystallization is complete, when the stress–strain curve is similar in form to that observed during the initial deformation. Minor differences remain when the recrystallized grain size differs from the initial grain size.

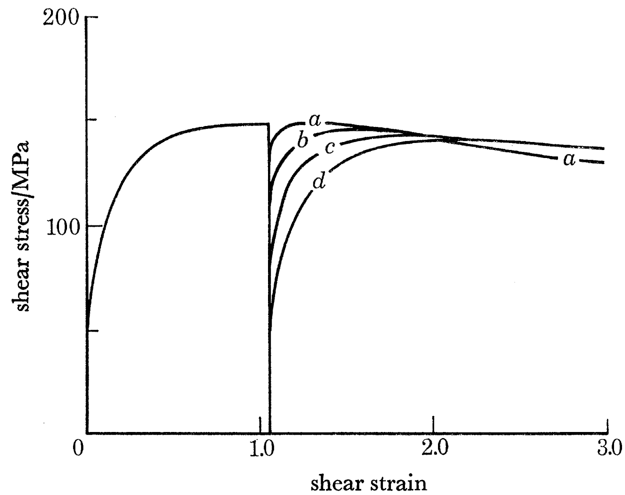

FIGURE 8. Effect of increasing rest period on the stress–strain curve for 18:8 austenitic stainless steel deformed in plane strain compression at an equivalent shear strain rate of 34 s^{-1} at 916 °C: *a*, 1 s; *b*, 20 s; *c*, 50 s; *d*, 300 s.

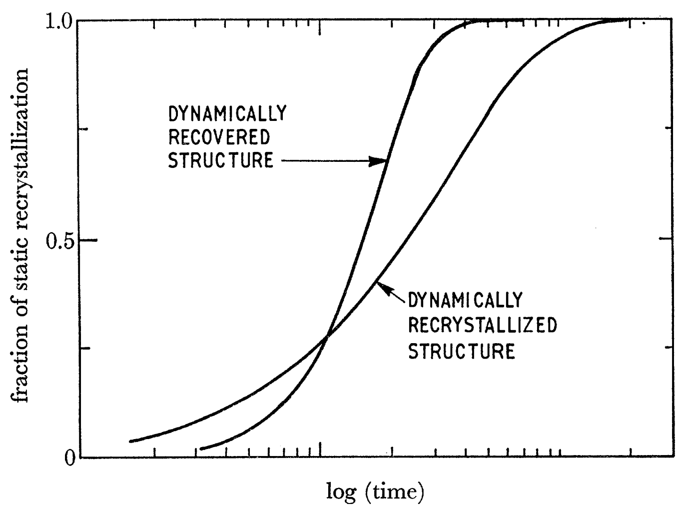

FIGURE 9. Time dependence of static recrystallization of dynamically recovered structures and of dynamically recrystallized structures produced during high-temperature prestrain.

Direct metallographic observations (Glover & Sellars 1972) and the technique of interrupting the deformation (Djaic & Jonas 1972, 1973; Petkovic, Luton & Jonas 1974) have established that the kinetics of static recrystallization differ fundamentally when the deformed structure has developed by work hardening and recovery only and when dynamic recrystallization has occurred (figure 9). The reasons for this are discussed later. Static recrystallization following dynamic recrystallization has been referred to as 'metadynamic' recrystallization (Djaic & Jonas 1972) to distinguish the process from the more 'classical' recrystallization of dynamically recovered structures which is closely related to the processes that occur on annealing metals after cold deformation.

Interrelation of static softening processes

The way in which static softening occurs in metals which undergo dynamic recrystallization during deformation is shown schematically as a function of prestrain in figure 10.

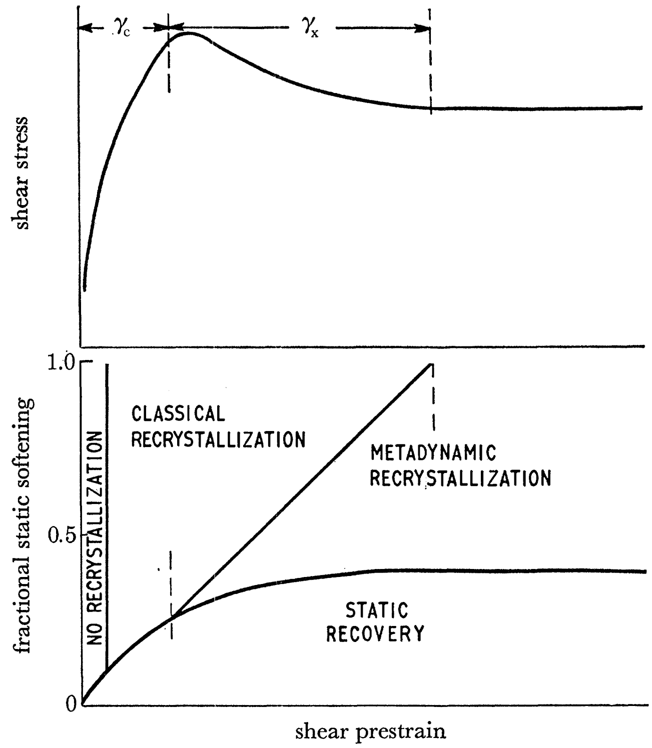

Figure 10. Schematic representation of the interrelation between the static softening mechanisms as a function of prestrain in a material that dynamically recrystallizes (after Djaic & Jonas 1973).

Static recovery

After all prestrains, static recovery takes place immediately the deformation is halted and proceeds at a decreasing rate with time. The recovery process involves the annihilation of dislocations in individual events and accounts for up to 40–50 % of the total softening at high strains (Evans & Dunston 1971; Petkovic *et al.* 1974). At low strains the stored energy is insufficient to cause static recrystallization and limited softening takes place by static recovery alone. The critical strain for static recrystallization after deformation at relatively high strain rates appears to be about 0.05–0.1, depending on deformation conditions (Djaic & Jonas 1973). This value is well below the critical strain for dynamic recrystallization, γ_c. The difference is consistent with the earlier conclusion that at relatively high strain rates (high stresses), excess stored energy, which requires higher strains, is necessary for dynamic recrystallization as growth of nuclei must occur more rapidly than some critical rate.

Equivalent measurements do not appear to have been carried out after deformation at low strain rates, but it would be expected that under these conditions the critical strain for static recrystallization will more nearly approach γ_c. Similar considerations to those discussed for γ_c will determine the absolute lower limit of strain rate (or stress) during deformation for static recrystallization to occur at all after any amount of prestrain.

Classical recrystallization

For the deformation conditions appropriate to figure 10, prestrains greater than the critical value for static recrystallization but less than γ_c result in classical recrystallization after an incubation period in which recovery processes create the recrystallization nuclei. The rate of recrystallization in a given material is determined by the stored energy, the density of favourable nucleation sites and the temperature. The stored energy increases with increasing strain rate and decreasing temperature of deformation and is strongly dependent on strain. All these factors therefore increase the recrystallization rate (English & Backofen 1964; Morrison 1972; Glover & Sellars 1972; Djaic & Jonas 1973). As nucleation takes place preferentially at grain boundaries, the density of favourable nucleation sites increases with decrease in grain size, leading to more rapid recrystallization in finer grained materials (Barraclough 1974). Recrystallization is a thermally activated process and so the rate is strongly dependent on the temperature of holding after deformation, e.g. with other conditions held constant the rate of recrystallization of α-iron decreases by about one order of magnitude for each 50 °C drop in temperature (Glover & Sellars 1972).

It is important to distinguish between the two independent rôles that temperature plays in influencing recrystallization after deformation at constant strain rate. Whereas decreasing the holding temperature decreases the rate of recrystallization, decreasing the temperature of deformation increases the stored energy at a given strain and therefore tends to increase the rate. This means that recrystallization occurs more rapidly when the temperature at which deformation has taken place is well below the temperature of holding.

Apart from the temperature of holding, the variables which increase recrystallization rate also decrease the recrystallized grain size (Glover & Sellars 1972; Barraclough 1974). The lack of effect of temperature is, however, often masked by the occurrence of grain growth after recrystallization is complete as this is a strongly temperature dependent process.

Metadynamic recrystallization

When the prestrain exceeds γ_c, an increasing fraction of the softening occurs by metadynamic recrystallization until at prestrains in the steady state range this becomes the only static recrystallization process.

During dynamic recrystallization at relatively high strain rates there are always nuclei present in the material and some grain boundaries are migrating. When the deformation is halted, these boundaries continue to migrate and nuclei continue to grow without the need for an incubation period. This form of recrystallization therefore proceeds very rapidly after deformation (figure 9), but the rate of recrystallization falls with time as the grains grow progressively into less dislocated material in the heterogeneous structure produced by the repeated cycles of dynamic recrystallization.

The overall rate of metadynamic recrystallization is influenced by the same factors as classical recrystallization, but after steady state deformation the stored energy is no longer dependent on strain and the dynamically recrystallized grain size is independent of the original grain size.

The important variables are therefore the strain rate and temperature of deformation and the temperature of holding (Glover & Sellars 1972; Djaic & Jonas 1972; Barraclough 1974).

The grain size produced by metadynamic recrystallization is uniquely determined by the steady state flow stress during the prestrain. As shown in figure 5, the relation follows the form given in equation (1), although the stress exponent is not necessarily identical to that for dynamic recrystallization (Glover & Sellars 1972, 1973). Because repeated nucleation does not occur during metadynamic recrystallization, the grain size is larger than that produced by dynamic recrystallization (Glover & Sellars 1972, 1973; McQueen & Bergerson 1972).

Conclusion

This survey of recrystallization during and after high-temperature deformation of metals has shown that three types of recrystallization may occur. Their effects on microstructure and strength, or creep rate, have been considered, and the fact that the deformation must always take place by dislocation movement rather than by diffusion mechanisms has been emphasized.

The range of deformation conditions over which dynamic recrystallization is observed is generally more restricted than that for recovery creep and, even under favourable conditions, a critical strain must be exceeded before either dynamic recrystallization or static recrystallization after deformation takes place. Within these limitations, equivalent recrystallization processes would be expected in other crystalline solids in which dynamic recovery is relatively slow. They will be favoured when the material is also of high purity and fine grain size.

References (Sellars)

Andrade, E. N. da C. 1948 *Nature, Lond.* **162**, 410.
Ashby, M. F. 1972 *Acta metall.* **20**, 887–897.
Ashby, M. F. 1973 *Third Int. Conf. on Strength of Metals and Alloys*, vol. 2, pp. 8–42. London: Institute of Metals; Iron and Steel Institute.
Barraclough, D. R. 1974 Ph.D. Thesis, University of Sheffield.
Bromley, R. & Sellars, C. M. 1973 *Third Int. Conf. on Strength of Metals and Alloys*, vol. 1, pp. 380–385. London: Institute of Metals; Iron and Steel Institute.
Cairns, J. H., Clough, J., Dewey, M. A. P. & Nutting, J. 1971 *J. Inst. Metals* **99**, 93–97.
Cottingham, D. M. 1966 *Deformation under hot working conditions*, pp. 145–156. Iron and Steel Institute Publication, no. 108. London.
Djaic, R. A. P. & Jonas, J. J. 1972 *J. Iron Steel Inst.* **210**, 256–261.
Djaic, R. A. P. & Jonas, J. J. 1973 *Metall. Trans.* **4**, 621–624.
English A. T. & Backofen, W. A. 1964 *Trans. T.M.S.–A.I.M.E.* **230**, 396–407.
Evans, R. W. & Dunston, G. R. 1971 *J. Inst. Metals* **99**, 4–14.
Gifkins, R. C. 1958–9 *J. Inst. Metals* **87**, 255–261.
Glover, G. & Sellars, C. M. 1972 *Metall. Trans.* **3**, 2271–2280.
Glover, G. & Sellars, C. M. 1973 *Metall. Trans.* **4**, 765–775.
Greenwood, J. N. & Worner, H. K. 1939 *J. Inst. Metals* **64**, 135–158.
Hardwick, D., Sellars, C. M. & Tegart, W. J. McG. 1961–2 *J. Inst. Metals* **90**, 21–22.
Jenkins, W. D., Digges, T. G. & Johnson, C. R. 1954 *J. Res. natn. Bur. Stand.* **53**, 329–352.
Jonas, J. J., Sellars, C. M. & Tegart, W. J. McG. 1969 *Metall. Rev.* **14**, 1–23.
Luton, M. J. & Sellars, C. M. 1969 *Acta metall.* **17**, 1033–1043.
McQueen, H. J. & Bergerson, S. 1972 *Metal Sci. J.* **6**, 25–29.
Morrison, W. B. 1972 *J. Iron Steel Inst.* **210**, 618–623.
Nutting, J. 1974 *Eighth International Conference on Electron Microscopy*, vol. 1, pp. 580–581.
Petkovic, R. A., Luton, M. J. & Jonas, J. J. 1974 *Int. Symposium on the hot formability of steels*, Strbske Pleso, Czechoslovakia.
Richardson, G. J., Sellars, C. M. & Tegart, W. J. McG. 1965 *Acta metall.* **14**, 1225–1236.
Rossard, C. & Blain, P. 1958 *Rev. Metall.* **55**, 573–594.

Sah, J. P. 1971 Ph.D. Thesis, University of Sheffield.
Sah, J. P., Richardson, G. J. & Sellars, C. M. 1974 *Metal Sci.* **8**, 253–331.
Sandström, R. & Lagneborg, R. 1975 *Acta metall.* **23**, 387–398.
Steinemann, S. 1958 *Beitr. Geol. Schweiz* (*Hydrologie*) **10**, 1–22.
Warrington, D. H. 1976 Computer simulation for materials science applications. In *Nuclear metallurgy*, vol. 20, pt 2, pp. 672–683. National Bureau of Standards.
Weertman, J. & Shahinian, P. 1956 *Trans. A.I.M.E.* **206**, 1223–1226.
White, S. 1976 *Phil. Trans. R. Soc. Lond.* A **283**, 69–86.

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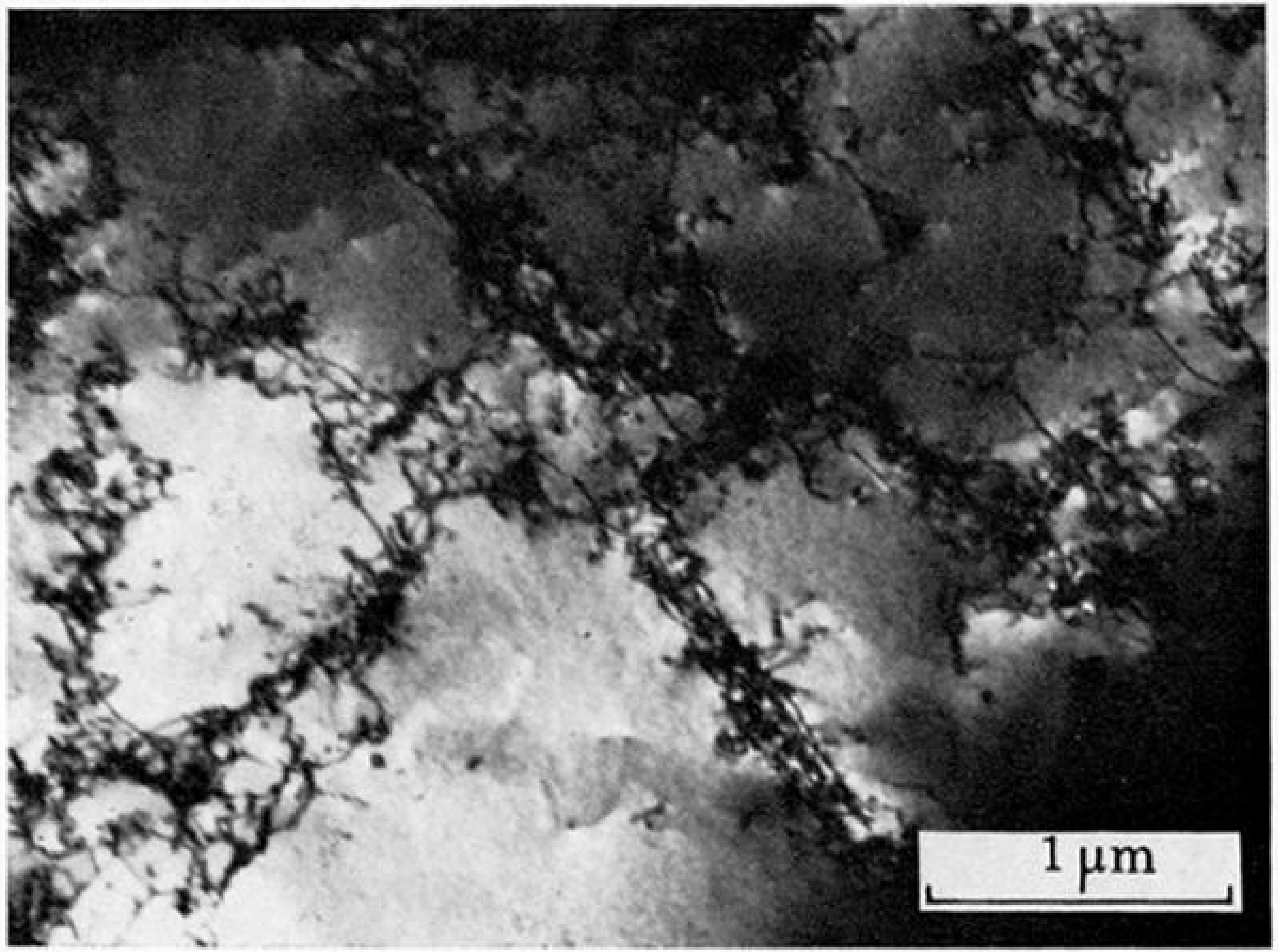

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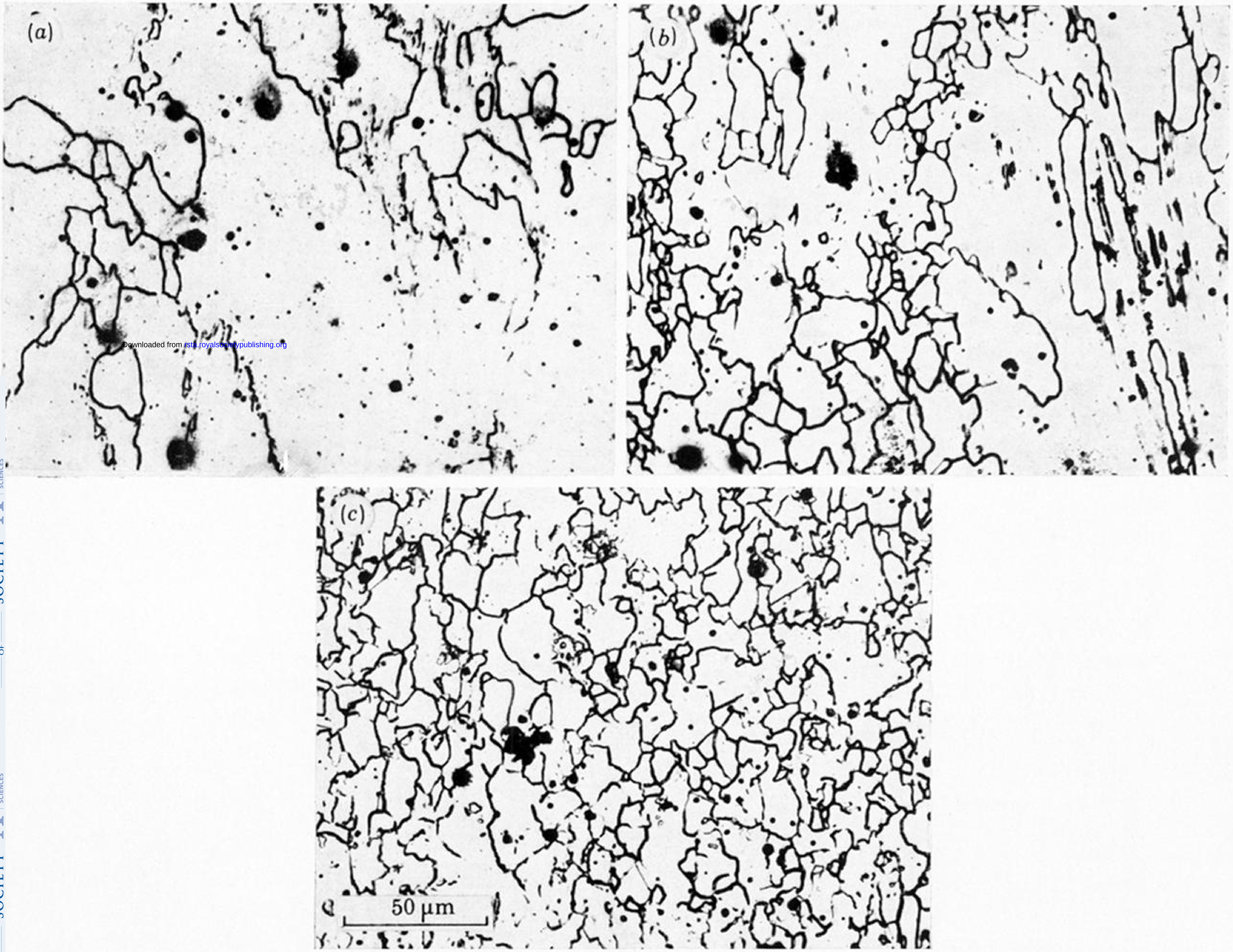

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