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PREDICTION OF MICROSTRUCTURE DEVELOPMENT DURING RECRYSTALLIZATION HOT ROLLING OF Ti-V STEELS

William Roberts
Institutet för Metallforskning
S-114 28 Stockholm, Sweden

Alf Sandberg
Institutet för Metallforskning
S-114 28 Stockholm, Sweden

Tadeusz Siwecki
Institutet för Metallforskning
S-114 28 Stockholm, Sweden

Tommy Werlefors
Svenskt Stål AB
S-613 01 Oxelösund, Sweden

ABSTRACT

In recrystallization hot rolling of plate, the static recrystallization of austenite at moderate rolling temperatures is used to engender fine as-hot-rolled ferrite grain sizes and concomitantly attractive as-rolled properties. The advantages of the procedure in comparison with controlled rolling are shorter throughput times on conventional mills and lower rolling loads. In the present paper, the essence of a computer model for predicting microstructural evolution during recrystallization hot rolling of a Ti-V microalloyed steel is given. The model is shown to forecast a behaviour which is in acceptable accord with practical rolling experience. Furthermore, the effect of the principal rolling variables on the degree of microstructural refinement during processing via recrystallization rolling has been investigated theoretically in a systematic way. Low finish-rolling temperatures and heavy reductions, especially during the final passes, promote fine as-rolled ferrite grain sizes; the austenite grain size created during reheating has no bearing on the final as-rolled microstructure. In addition, a suggestion is given regarding the design of rolling schedules which fulfill the conflicting requirements of efficient microstructural refinement (large reductions in the final rolling passes) and satisfactory plate flatness (small reductions).

RISING ENERGY COSTS have motivated attempts to obviate normalizing as a step in the production of constructional steel plates for applications specifying moderate to high toughness levels. *Controlled rolling* is one possibility, but this is an expensive production method unless the plant available is specifically designed for the process; on conventional, low-capacity mills the production rate with controlled rolling is often unacceptably low. Another method of enhancing toughness of as-hot-rolled plates is via *accelerated cooling*; however, many mills lack the necessary facilities to take advantage of this procedure. In recent years, increasing interest has been centred upon obtaining small as-hot-rolled ferrite grain sizes after finish rolling at relatively high temperatures, so-called *recrystallization rolling*. This procedure, which constitutes the principal substance of the present paper, is attractive in that it is relatively uncomplicated and can be applied on conventional mills.

In recrystallization rolling, the basic philosophy is to make use of the grain refinement accruing from the static recrystallization of austenite at moderate finish-rolling temperatures (FRT), typically 900-1000°C. However, in order to maintain the fine recrystallized grain size during cooling between the FRT and the temperature at which the steel starts to transform to ferrite, it is necessary that grain growth be restricted in some way e.g. by particles. It is known from earlier work that TiN is a suitable candidate for such grain-growth inhibition. With continuous-casting technology, it is possible to induce a fine distribution of TiN-particles which are formed in association with and immediately following solidification of the steel. This dispersion is then extremely stable during reheating at normal temperatures and under rolling. The small TiN-inclusions existing after strand casting of Ti-microalloyed steels, not only engender effective grain-growth inhibition between rolling passes and during cooling following the termination of rolling, but are also useful for restricting HAZ-grain coarsening in association with subsequent welding.

Properly performed with an appropriate steel analysis, recrystallization hot rolling is capable of yielding as-rolled ferrite grain sizes small enough to meet moderate toughness requirements. In addition, the method offers a considerable potential for attaining high production rates on conventional mills combined with

relatively low rolling loads.

Obviously, the overall degree of grain refinement produced by recrystallization rolling depends sensitively on the rolling schedule adopted. In general, rolling schemes, are primarily designed from the point of view of rolling load as well as shape and dimensional control of the final product; this places constraints on the extent to which established procedures can be modified with a view to obtaining fine microstructure. The situation is therefore somewhat complicated and optimization of routines for recrystallization rolling by trial-and-error on a full-scale mill is both time consuming and expensive. In this context, an accurate model for microstructural evolution during rolling, which facilitates a careful theoretical assessment of the degree of microstructural refinement effected by various rolling schedules, is clearly of value. The substance of this report is a presentation of such a microstructure-evolution model for hot rolling and a discussion of its application to forecasting the microstructure engendered by recrystallization hot rolling of a Ti-V microalloyed steel.

MODEL FOR MICROSTRUCTURE DEVELOPMENT DURING HOT ROLLING

The principal material parameters governing the evolution of austenite microstructure in association with hot rolling are the *static recrystallization characteristics*, recrystallization kinetics and recrystallized grain size, and the extent of *normal grain growth* once recrystallization is finished. An additional factor which must be considered is *static recovery* which modifies the retained driving force (dislocation density) if recrystallization is not completed between passes. With sufficiently large reductions during the initial, high-temperature passes, *dynamic recrystallization* may occur under the very deformation; however, in such a case, the dynamically-recrystallized microstructure is inevitably modified as a result of so-called *metadynamic* (or quasi-static) processes once the stock has emerged from the roll gap. For the purposes of the present treatment, no attempt is made to distinguish between static and metadynamic processes; the relevant quantities are rather the rate at which the reactions proceed and the resultant grain size.

The extent of recrystallization and grain growth between rolling passes and after the termination of processing is, of course, strongly dependent on the various quantities which are collectively called the rolling schedule. In this context, the principal variables are:

i) the vector of pass reductions (N elements for N passes);
ii) the vector of pass temperatures (N elements);
iii) the vector of inter-pass times ((N-1) elements); and
iv) parameters defining the cooling of the stock between FRT and Ar_3 which in turn depend on the plate or strip thickness, application of accelerated cooling etc.

In addition to the rolling variables listed above, the thermal cycle during reheating will have bearing on the microstructural evolution during subsequent rolling; the net effect is incorporated in a single quantity, D_{start} which is the austenite grain size immediately prior to the first rolling pass.

Since in any practical situation, one is usually interested in elucidating the influence of rolling procedure on as-hot-rolled ferrite grain size (D^{α}), then quite apart from material parameters governing static recrystallization, static recovery and grain growth of austenite, it will be necessary to specify a relationship between the grain sizes of austenite and ferrite for various conditions of cooling.

Finally, it should be pointed out that the present model gives no cognizance whatsoever to any inhomogeneity of deformation or temperature during rolling; the effect of such phenomena is assumed to be of the second order. In theory, prediction of the concomitant microstructural gradients should be feasible but this would require detailed information pertaining to the distributions of temperature and strain after each pass. On the other hand, the influence of the homogeneous, adiabatic temperature increase due to the work of deformation can easily be taken into account via the pass-temperature vector.

STATIC RECRYSTALLIZATION KINETICS - The dependence of recrystallized fraction (X) on the time (t) during recrystallization is usually found to follow the *Avrami law* i.e.,

$$X = 1 - \exp(-kt^n) \qquad (1)$$

where k,n are constants. The Avrami exponent, n, may be regarded as a material parameter whereas k contains the dependence of X(t) on temperature, prior strain, grain size etc. For C-Mn and microalloyed constructional steels, the literature values for n relevant to the static recrystallization of austenite, lie in the range $1,5 \leq n \leq 2$; for the present purposes, a value of $n=1,7$ is adopted.

As discussed in detail by Sellars and coworkers (1,2), the influence of prior equivalent strain (assumed in the instance of flat rolling to be $1,15 \ln(H_{in}/H_{out})$), temperature (T) and pre-existing grain size(D_o) on the kinetics of static recrystallization is conveniently expressed in terms of the time required to induce some specified recrystallized fraction, say X=0,5. Experimental experience dictates that $t_{0,5}$ conforms to an empirical relationship of the form

$$t_{0,5} = C_1 \varepsilon^x D_o^y \exp(Q_{rex}/RT), \qquad (2)$$

where x,y are empirical power exponents and Q_{rex} is the activation energy for static recrystallization following hot deformation. Rewriting Eq. (1) in terms of $t_{0,5}$ instead of k gives

$$X = 1 - \exp\left[-0{,}693(t/t_{0,5})^n\right], \qquad (3)$$

which defines X(t) once $t_{0,5}$ is known. For n=1,7, the values tabulated below apply:

TABLE 1

X	$t_X/t_{0,5}$
0,1	0,329
0,3	0,675
0,7	1,385
0,9	2,030

The correlation between Eqs.(2) and (3) and experimentally-determined recrystallization kinetics for austenite is here restricted to investigations where direct metallographic techniques have been used to establish X(t). Many workers have studied static recrystallization by determining a softening index derived from the reduction in flow stress following a pause time in a two-step test. Such results have not been included in the correlation for $t_{0,5}$ because of the uncertain contribution from recovery to such a softening index. For Al-killed, C-Mn austenite, an evaluation based on data given in references (3-8) leads to the following equation:

$$t_{0,5}=5.10^{-21}\ \varepsilon^{-4} D_o^{\ 2} \exp\left(\frac{330.000}{RT}\right), \qquad (4)$$

where D_o is in μm and R=8,314 J/(°K.mol). This is very similar to the expression given by Sellars (2) for C-Mn austenite. Fig.1 demonstrates the measure of agreement between X(t) evaluated on the basis of Eqs.(3) and (4) and some of the experimental observations.

As stated in the introductory remarks, the present paper will be devoted mainly to a consideration of recrystallization rolling of a Ti-V microalloyed HSLA-steel. The grade of interest has the following typical analysis: C=0,13%, Mn=1,4%, V=0,04%, Ti=0,01%, N=0,01% (Al-killed). An extensive investigation of the static-recrystallization kinetics for this steel has been carried out via single-step hot-compression/holding experiments with a range of values for ε, D_o and T. It is quite clear from this work that the recrystallization kinetics of the Ti-V steel are very much slower for small prestrains than in the case of C-Mn austenite. For example, for a prior strain of 0,073 at 1200°C, $t_{0,5}$ is observed to be ∿50s (D_o= 17 μm) for Ti-V austenite; the corresponding C-Mn value is, from eqn.(4), ∿0,03s. The reason for the delayed recrystallization in Ti-V austenite is thought to derive from the drag force on moving grain boundaries due to the presence of TiN-particles. The best-fit relationship for the kinetic data on recrystallization of Ti-V austenite is

$$t_{0,5}=5.10^{-18}\ (\varepsilon-0{,}058)^{-3,5}\ D_o^2 \exp\left(\frac{280.000}{RT}\right). \qquad (5)$$

According to this, Ti-V austenite deformed to strains below 0,058 will not recrystallize at all. Below this strain, the drag force due to TiN evidently exceeds the dislocation-density driving force. In order make a rough check on the validity of this statement, we can set the dislocation-density (ρ_o) driving force as $\tau.\rho_o$ where τ is the dislocation line energy. For austenite at 1100°C $\rho_o(\varepsilon=0{,}058)$ is about $3.10^{12} m^{-2}$ and with $\tau=2{,}3.10^{-9}$ J/m, the driving force for recrystallization is $7.10^3 J/m^3$. The Zener drag from a random distribution of spherical particles (radius r, volume fraction f) is approximately $3Sf/4r(9)$ where S is the grain-boundary energy of austenite (0,8 J/m^2). Assuming 0,01%Ti combined as TiN, then $f=2{,}2.10^{-4}$ and with r= 10 nm, as typical for strand-cast steel, we expect a drag force of $1{,}3.10^4$ J/m^3. The close proximity of this figure to the approximate value of the dislocation-density driving force supports the contention that TiN delays recrystallization in Ti-V steels.

At large strains (ε>0,2), the values for $t_{0,5}$ predicted by Eqs.(4) and (5) are very similar (all other things being equal). This is to be expected since the drag derived from TiN-particles is then small in comparison with the dislocation-density difference driving recrystallization. The degree of accord between eqn.(5) and experimentally-determined recrystallization kinetics for Ti-V austenites is illustrated in Fig.2.

RECRYSTALLIZED GRAIN SIZE - The austenite grain size which prevails at the instant of completion of static recrystallization (D_{rex}) depends sensitively on strain and pre-existing grain size but is only a weak function of temperature; indeed, a number of authors have reported that the temperature dependence of D_{rex} (C-Mn steels; fixed ε,D_o) is virtually nil. For C-Mn austenite, a least-squares treatment of the available literature data (2,4,8,10-15) suggests the following relationship for D_{rex}(see Fig.3):

$$D_{rex}= 6{,}2+55{,}7.\ D_o^{0,5}\ \varepsilon^{-0,65}\ \left[\exp\left(\frac{350.000}{RT}\right)\right]^{-0,1} \qquad (6)$$

This is similar to the equation given by Sellars (2), but according to him the powers of D_o and ε should be higher and there is no temperature dependence at all.

For Ti-V austenite, with about the composition given previously, a systematic experimental study of the influence of D_o,ε and T on statically-recrystallized grain size leads to the following best-fit relationship,

$$D_{rex}=4{,}3+195{,}7.\ D_o^{0,15}\ \varepsilon^{-0,57}\ \left[\exp\left(\frac{350.000}{RT}\right)\right]^{-0,11} \qquad (7)$$

The experimental base for this equation is presented in Fig.4. Some uncertainty must be ascribed to the exponent for D_o because the range of values for this quantity has been very limited in the experiments, $17 \leq D_o \leq 40$ μm; (the variation in grain size attainable by altering the reheating

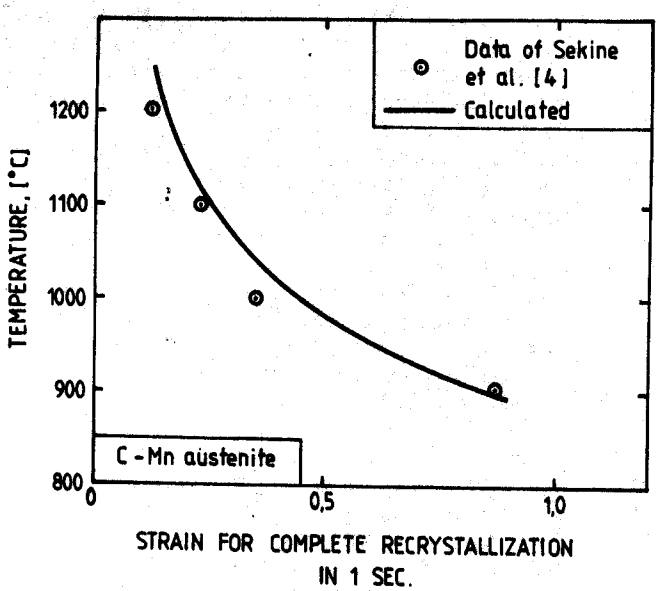
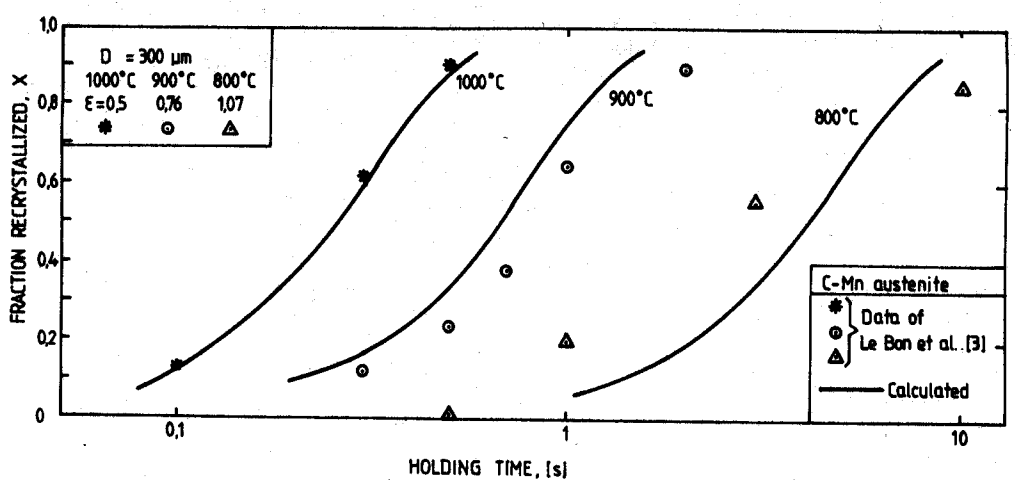


Fig.1 Illustrating the degree of correspondence between experimentally-evaluated recrystallization kinetics for C-Mn austenite and the behaviour predicted by Eqs.(3) and (4).



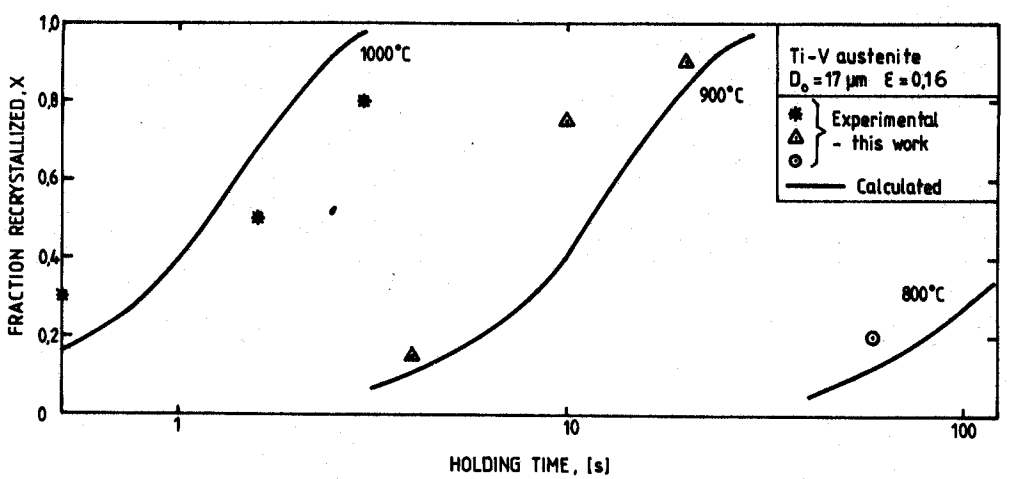


Fig.2 Illustrating the degree of correspondence between experimentally-evaluated recrystallization kinetics for Ti-V austenite and the behaviour predicted by Eqs.(3) and (5).

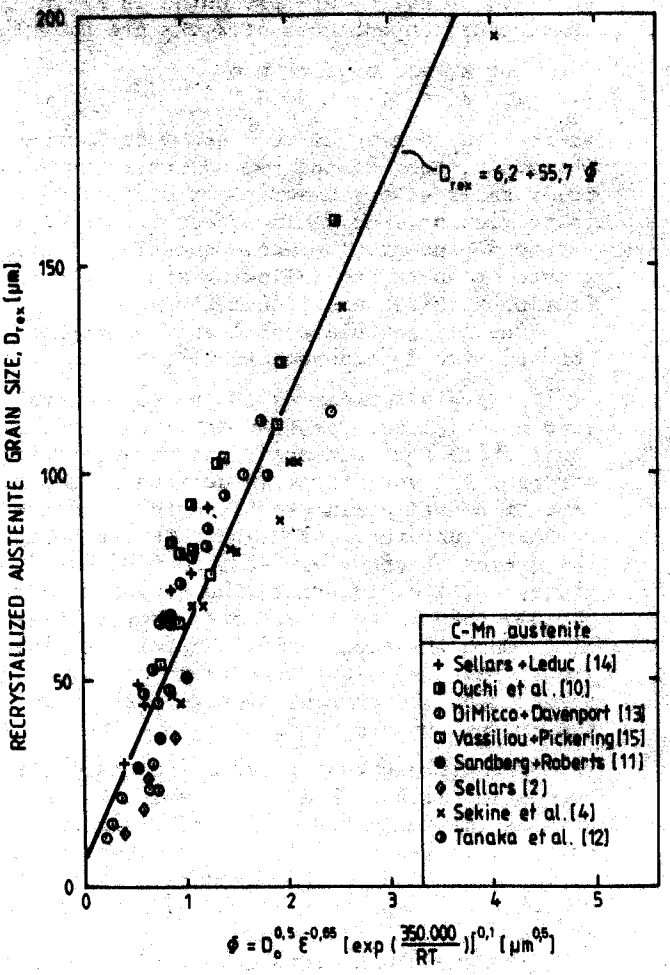


Fig. 3 Correlation between statically-recrystallized grain size and strain, pre-existing grain size and temperature for C-Mn austenite (literature data).

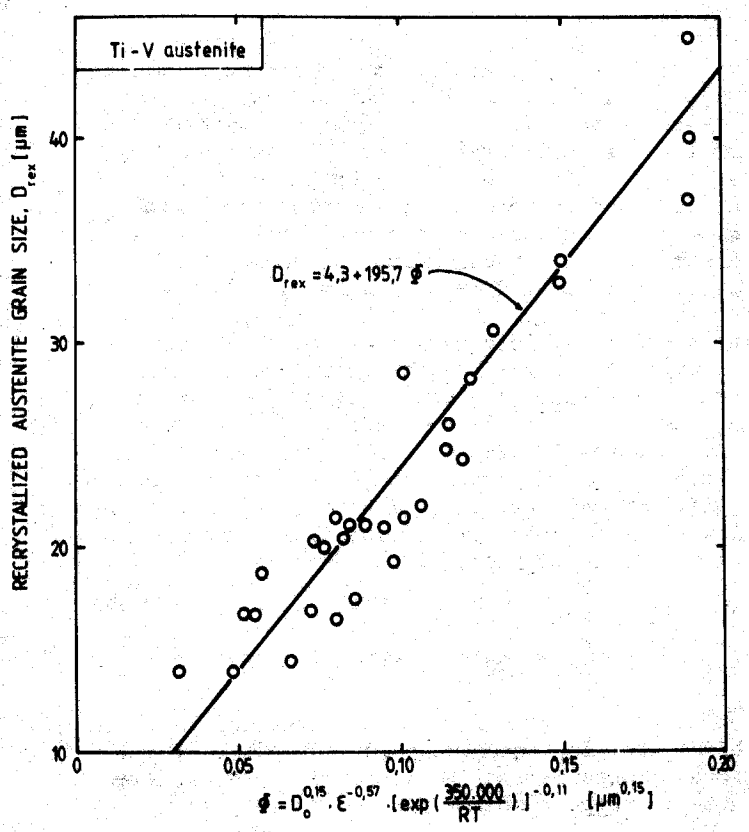


Fig. 4 Correlation between statically-recrystallized grain size and strain, pre-existing grain size and temperature for Ti-V austenite.

conditions is rather small for Ti-V steels, unless abnormal grain growth occurs). For small values of D_o(~20 μm) and otherwise equivalent deformation conditions, Eqs.(6) and (7) predict about the same D_{rex}. At larger D_o's, Ti-V austenite is evidently grain-refined more efficiently, an effect which may be of interest in a situation where abnormal grain growth of austenite has occurred during reheating.

Finally, it is worth pointing out that Eqs. (6) and (7) are by no means restricted to the case where static recrystallization effects grain refinement. For example, grain refinement or grain coarsening will occur in C-Mn austenite depending on whether D_o is respectively less than or greater than

$$3{,}1.10^3\ \varepsilon^{-1{,}3}\left[\exp\left(\frac{350.000}{RT}\right)\right]^{-0{,}2}.$$

The above expression (a similar one can be written for Ti-V steels) is not exactly correct because to obtain it, the first term in Eq.(6) is approximated to zero.

NORMAL GRAIN GROWTH — A correct description of austenite grain growth, both during inter-pass periods following the completion of recrystallization and after the termination of rolling, is an essential ingredient of a realistic microstructure-evolution model. For Ti-V steels, the grain-growth rate after complete static recrystallization is found to be very slow (as expected) even at temperatures as high as 1200°C (Fig.5). Hence, one makes little error in assuming that grain growth of Ti-V austenite is completely suppressed during the entire rolling sequence; this represents a considerable simplification.

Although the present report is concerned with Ti-V steels, grain growth of C-Mn austenite will be afforded brief attention since, later on, the two types of steel are compared with regard to the austenite grain sizes engendered during a given rolling schedule. As pointed out by Sellars and Whiteman (1), the isothermal grain-growth data after static recrystallization, which are reported in the literature, do not conform to a simple parabolic relationship, $d(D^2)/dt$ at short times being much too high relative to that at long times. This behaviour led these authors to suggest a power law for grain growth with a very large exponent (ten). However, a dependence as high as D^{10} leads to a growth rate which is unrealistically high at small grain sizes. Examination of the reported data suggests that the behaviour can instead be described by a parabolic law with different proportionality constants for short and long times, i.e.

$$D^2 = D_{rex}^2 + k_1 t \qquad (t>t_{bp}) \qquad (8a)$$

$$D^2 = D_{bp}^2 + k_2 t \qquad (t<t_{bp}), \qquad (8b)$$

where $k_1 >> k_2$. To a first approximation, the breakpoint for the change in $d(D^2)/dt$ is found to be independent of temperature, t_{bp}=20s. Some data in support of the type of grain-growth law embodied in Eq.(8) are presented in Fig.6. The temperature dependences of k_1, k_2 are given by

$$\log k_1 = 6{,}6 - 6200/6 \qquad (9a)$$

$$\log k_2 = 8{,}1 - 9000/T \qquad (9b)$$

where k_1, k_2 are in $\mu m^2.s^{-1}$ and T in degrees Kelvin. Obviously, Eqs.(8) and (9) do not represent a fully satisfactory description of grain growth; in particular, one would expect t_{bp} to be temperature dependent. The whole question of grain growth of austenite following static recrystallization requires additional study.

One can postulate at least two explanations for the time dependence of $d(D^2)/dt$.

i) The grain boundaries of freshly-recrystallized material are far from their gently-curved, equilibrium form but are rather characterized by strong local variations in curvature. Hence, the rate of normal grain growth, which is driven by boundary curvature, is high up to the point where the grains attain their equilibrium shape, or nearly so, i.e. initially, the grains grow more rapidly than is to be expected from their average size. Once the equilibrium shape is attained, the growth rate drops off to a value which is determined by the current average grain size and the grain-boundary mobility.

ii) Freshly-recrystallized grains have clean boundaries with high mobility which is consistent with an initially rapid rate of grain growth. However, as time goes on, the progressive segregation of solute to the boundaries causes the mobility to drop and with it the rate of increase of grain size.

Phenomenologically, both effects can be dealt with by introducing a time dependence of k in the normal grain-growth equation, i.e.

$$D^2 = D_o^2 + k(t).t.$$

In the very simple approach outlined above, k(t) is an elementary step function but one can envisage other forms for k(t); further work is necessary!

In microalloyed austenite containing particles, like TiN, Nb(C,N), the possibility of abnormal grain growth in association with rolling cannot be discounted. However, the kinetics of this process are usually so slow that it can hardly be expected to occur in connection with rolling unless the stock is held for a prolonged period at high temperature, e.g. holding during controlled rolling of Nb-steels. In the latter instance, the abnormal grain coarsening which has been reported (e.g.in(13)) is invariably associated with a light final predeformation pass and may in fact derive from a coarse *recrystallized* grain size associated with a small strain-energy driving force. For rolling schedules where all pass reductions at >1000°C exceed 10% and having times between passes restricted to 100s or less, the probability of abnormal grain growth is considered minimal; for this reason, no cognizance is given to this phenomenon in the present version of the microstructure-evolution model.

STATIC RECOVERY — When recrystallization does not proceed to completion between rolling

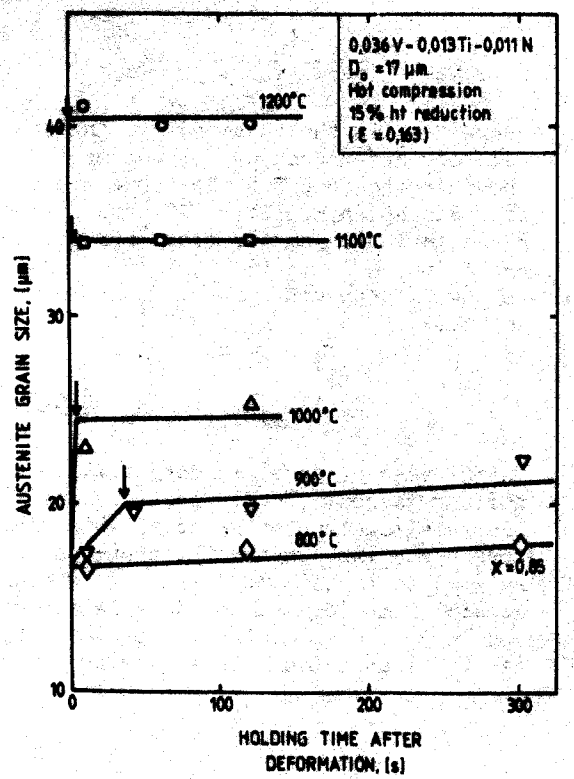


Fig.5 Grain growth in Ti-V austenite following static recrystallisation; arrows indicate $t_{0,9}$ evaluated from Eqs.(3) and (5).

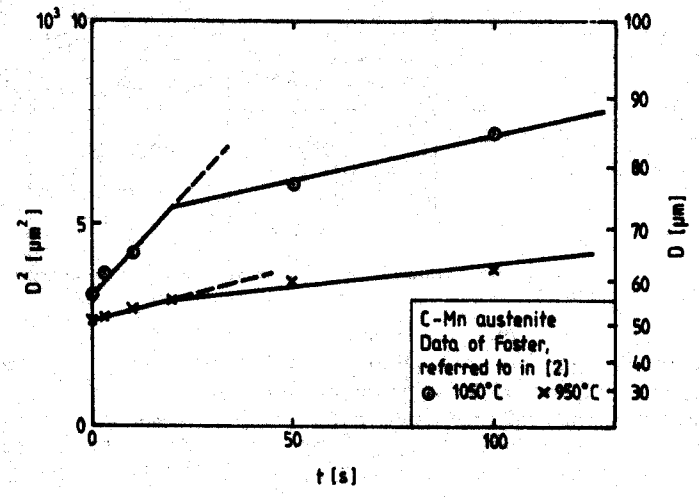
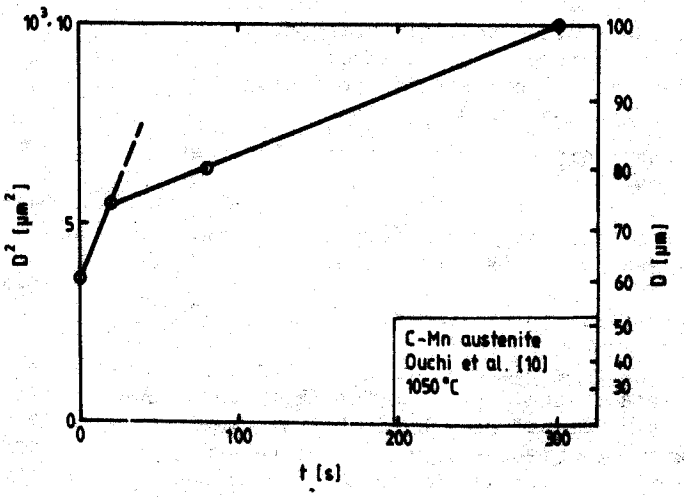


Fig.6 Illustrating the conformation to a double-parabolic behaviour of selected data for grain growth of C-Mn austenite following static recrystallisation.



passes, the unrecrystallized material will still undergo restoration via static recovery. The extent of recovery needs to be evaluated in order to define the accumulated recrystallization driving force after a given pass, in those regions which have remained unrecrystallized following the previous deformation. The theory of static recovery is relatively underdeveloped and the above effects can only be estimated in a rather crude manner.

Using the Friedel (16) approach for the growth of meshes in a three-dimensional dislocation network, the rate of decrease of dislocation density during static recovery can be expressed

$$\frac{d\rho}{dt} = -2M\tau\rho^2$$

where M is a dislocation mobility. Integration gives

$$1/\rho - 1/\rho_o = 2M\tau t \quad , \qquad (10)$$

ρ_o being the dislocation density engendered by prestraining (t=0). Presuming simple parabolic hardening, a reasonable approximation for small strains, i.e. $\rho_o = \psi\varepsilon$, then Eq.(10) is rewritten as

$$\varepsilon_{eff} = \frac{\varepsilon_o}{2M\tau\psi\ t\cdot\varepsilon_o + 1}$$

where ε_{eff} is the retained strain after static recovery has proceeded for a time t. Inserting known expressions for the temperature dependence of dislocation mobility (17) and dislocation accumulation parameter (18) appropriate to C-Mn austenite (M in $m^3/(J.s)$, ψ in m^{-2}), gives the following dependence for ε_{eff} on prior prestrain and time,

$$\varepsilon_{eff} = \varepsilon_o\left[1 + 0{,}023\ \exp\left(\frac{39.700}{T}\right).\right.$$
$$\left. 10\uparrow\left(10{,}8 + \frac{4020}{T}\right).\varepsilon_o t\right]^{-1} . \qquad (12)$$

In the case of partial recrystallization between rolling passes, the dislocation density driving recrystallization in material which has remained unrecrystallized after the previous pass, is related to the sum of ε_{eff} and the current pass strain. Some idea as to the kinetics of static recovery of C-Mn austenite predicted by Eq.(12) can be gleaned from the following table:

TABLE 2

t(s)	ε_{eff}			
	1000°C, ε_o=0,1	1200°C, ε_o=0,1	1000°C, ε_o=0,3	1200°C, ε_o=0,3
1	0,0991	0,086	0,294	0,201
5	0,0969	0,055	0,274	0,088
10	0,0940	0,038	0,252	0,051

An appraisal of the characteristics of recovery in Ti-V steels has not yet been made; consequently, the expression for C-Mn austenite has been used for the calculations reported later in the present paper, even though it is very likely that the kinetics of static recovery are affected by the presence of TiN-particles.

BRIEF DESCRIPTION OF COMPUTING ROUTINE — Input of the following basic data pertaining to the rolling schedule is required:

i) A reasonable estimate of the as-reheated austenite grain size (D_{start});

ii) A pass-temperature vector with N elements, one for each pass;

iii) A vector with (N+1) elements describing the scheme for thickness reduction. This contains a zero element which is the initial slab thickness (H_o);

iv) A vector of inter-pass times with (N-1) elements; and

v) A two-row matrix giving temperatures and times appropriate to cooling after the final pass down to the Ar_3-temperature. The number of elements depends on how precisely one is interested in approximating a non-linear cooling behaviour.

Complete inter-pass recrystallization — The computational routine is in this case fairly straightforward. The time required for complete recrystallization (X>0,9) between any two passes is evaluated (Eqs.(1-5)) together with the corresponding recrystallized grain size (Eqs.(6-7)). The increase in grain size due to grain growth between the finish of recrystallization and the next pass is then computed (Eqs.(8-9)). The grain size so calculated is taken as D_o for the subsequent pass and the procedure is looped (N-1) times. The time required for complete recrystallization together with the appropriate grain size after the final pass are then evaluated along with grain growth during cooling down to Ar_3. In all cases, the fact that the various microstructural changes proceed under conditions of falling temperature is taken into account via simple finite-difference methods.

Partial inter-pass recrystallization — When X<0,9 in the time available between two rolling passes, it becomes necessary to divide the material into recrystallized and unrecrystallized regions. Each of these is subdivided further after the next pass and so on. It will be appre-

ciated that if recrystallization is only partially completed following the Ith pass, then after the final pass the material will be divided into $2^{\uparrow}(N-I+1)$ components, each with its own recrystallization history. Hence, the amount of data will become unmanageably large if partial recrystallization occurs at the very beginning of rolling. Luckily, however, matters can be simplified because the contribution of many of the separate elements with different recrystallization histories to the total volume fraction is very small, say less than 0,01. By arranging that such very small material elements are combined with larger ones, in such a way that $\Sigma\Delta X$ is always unity, it is feasible to compute the microstructural changes, under conditions when static recrystallization is not completed between passes, in a manageable yet realistic manner.

In the computing routine, the material elements with different recrystallization histories constitute a recrystallized-fraction vector, for which the sum of the individual terms is unity. Grain sizes and accumulated strains are stored in a corresponding way. The accumulated strain in recrystallized material is obviously zero; that in an unrecrystallized element is modified to take account of recovery according to Eq.(12). The grain size of *recrystallized* material is evaluated straightforwardly from Eqs. (6-7). In *unrecrystallized* areas, the values used for grain size depend on whether D_{rex} is greater or smaller than D_o. In the former case, the grain size of unrecrystallized material is clearly D_o. Conversely, when grain refinement occurs, the size of the unrecrystallized regions is about

$$(1 - X)^{1/2} D_o.$$

This relationship is easily derived on the assumption that nucleation of new grains proceeds exclusively at pre-existing grain boundaries and that the recrystallized grains and the unrecrystallized regions are characterized by the same stereological coefficients relating linear dimensions to area and volume. In view of the uncertainty in the evaluation of D in unrecrystallized material, the latter is arbitrarily set to D_{rex} whenever these quantities differ by less than 20% of their value.

If for a *local* recrystallized/unrecrystallized pair in the recrystallized-fraction vector, the *local* X exceeds 0,9 then the effect of grain growth on the *local* grain size is evaluated. After each inter-pass period, the total fraction recrystallized and the average grain size are evaluated as

$$X_{tot} = \sum_{M=1}^{n} X_{2M-1}$$

and

$$\overline{D} = \left[\sum_{M=1}^{n} \frac{X_M}{D_M^2} \right]^{-1/2}$$

respectively, where n is the current number of elements in the recrystallized-fraction and grain-size vectors. If $X_{tot} > 0,9$, then the part of the programme treating complete inter-pass recrystallization is entered into once again with D_o set to $\overline{D}$.

Following the final pass, the evaluations of recrystallization time, recrystallized grain size and grain growth are performed for each element in the recrystallized-fraction vector. The total fraction recrystallized and the average grain size are computed from the above expressions. Furthermore, the final vectors for recrystallized fraction and grain size can be outputted if required, e.g. to examine the grain-size distribution engendered as a result of partial inter-pass recrystallization.

RELATIONSHIP BETWEEN AUSTENITE GRAIN SIZE AND FERRITE GRAIN SIZE — The austenite grain size (or grain-size distribution) at Ar_3, which is evaluated from the microstructure-evolution routine, can be used to determine the corresponding ferrite grain size (or grain-size distribution); any grain growth of ferrite is neglected. For the Ti-V steels of interest here, the following empirical expression has been found to fit experimental data ($D^{\gamma} > 8$ μm),

$$D^{\alpha} = 3,75 + 0,18 D^{\gamma} + 1,4 \left(\frac{dT}{dt}\right)^{-1/2}, \qquad (13)$$

where dT/dt refers to the average rate of cooling between 750 and 550°C. Actual experimental observations are compared with the values given by the above expression in Fig.7. There is no particular reason why other austenites, with the same carbon and manganese contents, should not conform to Eq.(13); some modification must, however, be made if transformation proceeds from heavily-deformed, unrecrystallized austenite as in low-temperature controlled rolling.

In the present paper, consideration is confined to the effect of rolling schedule on D^{α}. In principle, it is feasible to use this information and elucidate the relationship between rolling scheme and strength/toughness combination in a quantitative manner.

COMPARISON WITH LABORATORY HOT-COMPRESSION TESTING AND FULL-SCALE ROLLING

The predictions of the microstructure-evolution model described previously have been compared with results from laboratory hot-compression testing of a steel with 0,036%V, 0,013%Ti and 0,011%N. Samples were reheated for 5 min. at 1200°C and subjected to the following hot-deformation schemes (or parts of them):

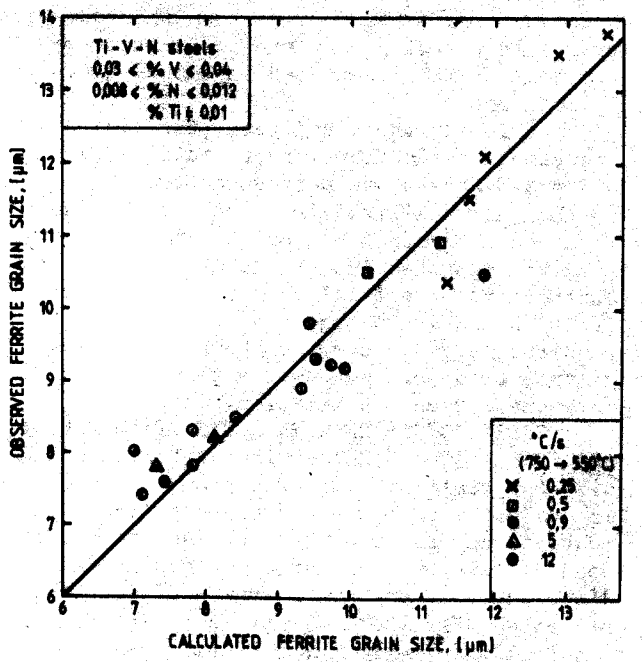


Fig. 7 Comparison of measured D^{α} in Ti-V-N steels and the value calculated on the basis of austenite grain size and cooling rate (Eq. (13)).

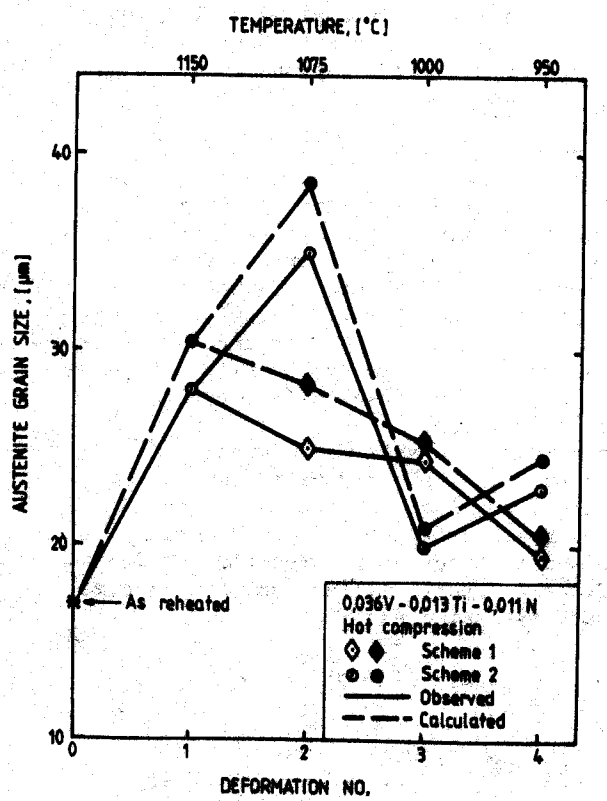


Fig. 8 Comparison of obeerved D^{α} with that predicted theoretically for hot-compression testing (four steps) of Ti-V austenite. For further details see Table 3 and text.

TABLE 3

T,(°C)	Height reduction , (%)	
	Scheme 1	Scheme 2
1150	20	20
1075	20	10
1000	20	30
950	20	10

The first specimen was quenched having undergone the initial reduction plus cooling to the temperature of the second deformation. The second sample was compressed at the first two temperatures before cooling to the third and quenching, and so on. In this way, the development of the austenite microstructure during the deformation schemes could be monitored. Two samples were subjected to all deformations, one being quenched from 800°C ($\sim Ar_3$) while the other was allowed to transform to ferrite. The cooling rate during the tests was maintained at 0,9°C/s; between 750 and 550°C, this corresponds to air cooling of 12 mm plate. The as-reheated austenite grain size was 17 μm. Fig.8 compares the austenite grain sizes found in the above experiments with those predicted from the microstructure model. The agreement is clearly as good as can be expected. The measured ferrite grain sizes were 9,3 and 9,8 μm for Schemes 1 and 2 respectively which compare with 9,0 and 9,7 μm evaluated theoretically.

The microstructure-evolution routine has also been used to compute ferrite grain sizes in thirty-one plates of Ti-V-N steel which have been processed by recrystallization hot rolling on the Oxelösund plate mill. The appropriate rolling scheme used (heights,temperatures, times) could in each case be logged out of the mill control computer, the schedule being worked out by the software on the basis of the required dimensions, mill loading and shape control. The temperature of the plate is measured directly after the final pass by a pyrometer and can be compared with the prediction from the temperature model which is incorporated in the control software. The computed D^{α} for each plate was compared with that evaluated metallographically and the results are summarized in Fig.9. The degree of accord is very satisfactory; indeed, for the plates showing the greatest discrepancy between observed and calculated grain sizes, there was invariably a considerable difference between the measured value of the temperature after the last pass and the one evaluated from the stock-temperature model, which suggests that in these instances the pass-temperature vector inputted might be somewhat in error.

The measure of agreement between the predictions of the microstructure-evolution model and results from laboratory compression testing and full-scale plate rolling, suggest that the former can be used with some confidence in elucidating the influence of the rolling schedule adopted for recrystallization hot rolling of Ti-V steels, on the overall degree of microstructural refinement which is achieved. This is the subject of the next section.

THE INFLUENCE OF ROLLING SCHEDULE ON THE DEGREE OF MICROSTRUCTURAL REFINEMENT

The philosophy behind recrystallization hot rolling of Ti-V steels is that a fine statically-recrystallized grain size be achieved by finish rolling at moderately low temperatures (900-1000°C) and that grain growth of the austenite is restrained, during cooling down to Ar_3, by TiN-particles. Clearly, these microstructural aspects should be borne in mind when designing the rolling schedule. However, with the exception of controlled rolling, commercial rolling schemes which take account of microstructural development are, at the present time, the exception rather than the rule. Cognizance is rather given to the following factors:

i) The capacity of the mill must under no circumstances be exceeded, high temperature and small reductions favour low rolling loads.

ii) The rolling operation should be effected as quickly as possible (high temperatures, small number of passes). This stipulation is often relaxed to some extent depending on current mill commitments and the gains as regards final properties which accrue from introducing hold times, e.g. controlled rolling.

iii) The shape and dimensions of the final product must lie within acceptable tolerances; in particular, plate and strip must be adequately flat. For example, poor flatness can be associated with high rolling loads (roll flexuring) and must then be counteracted by ensuring that the temperature range for rolling is high and that reductions are kept low during the final passes.

Without performing any calculations, one may state immediately that the above considerations impose constraints on rolling schedules which are in conflict with obtaining a high degree of microstructural refinement, the latter being favoured by large reductions at relatively low temperature. Hence, the problem with recrystallization rolling is to define rolling schedules combining acceptable microstructural evolution with low rolling loads, good flatness, adequate throughput etc.In this context, a model of the type described earlier can be extremely valuable; however, progress will be made only if the constraints on rolling procedure, which are determined by requirements such as flatness, can be clearly defined. One suspects that the restrictions in pass reduction etc. imposed by mill-control systems may be unnecessarily severe in many instances simply because the adopted shape-control model lacks sufficient precision.

INFLUENCE OF PLATE-ROLLING PARAMETERS ON MICROSTRUCTURAL REFINEMENT — This section gives

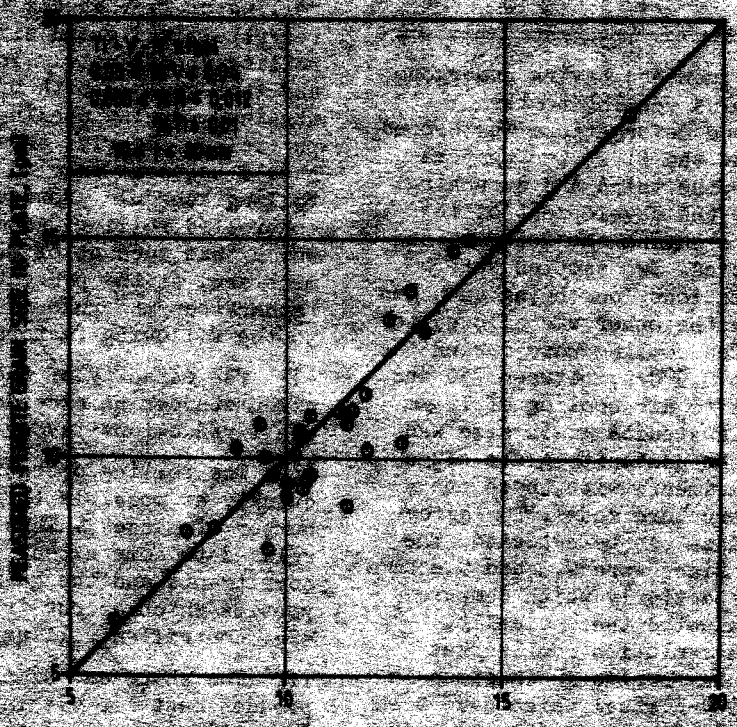


Fig. 9 Comparison between observed and calculated ferrite grain sizes for full-scale rolling of thirty-one plates of Ti-V-N microalloyed steel.

results from computations of microstructural evolution in Ti-V austenite engendered by a number of hypothetical rolling schemes. In the first instance, the main emphasis is placed on achieving a high degree of microstructural refinement; whether or not the schedules considered are suitable as regards shape control, rolling load etc. is of secondary importance.

The *thickness-reduction schemes* which have been examined are tabulated below (H_o = 220 mm).

TABLE 4

	Thickness (mm)					
	12 mm plate		25 mm plate		40 mm plate	
Pass nr	9 pass	12 pass	8 pass	11 pass	7 pass	10 pass
1	200	200	200	200	200	200
2	165	170	165	170	165	170
3	140	150	140	150	130	150
4	110	130	110	130	100	130
5	→ 82	112	→ 82	112	→ 74	112
6	57	93	60	→ 94	54	→ 95
7	33	→ 75	38	77	40	79
8	21	58	25	60		64
9	12	42		45		50
10		29		33		40
11		19		25		
12		12				

Quite obviously, the heavy final-pass reductions in the above schedules are not especially conducive to good shape control. These reduction schemes have been combined with various *rolling temperature ranges*, partly by altering the start temperature (1200, 1100 and 1050°C) and partly via introduction of hold times during which the stock can cool (60-80s); the points at which rolling is interrupted are indicated by arrows in Table 4. Apart from the hold periods, the inter-pass times have been set to 10s. Three starting temperatures with and without hold times give six FRT's for each reduction scheme. In every case, D_{start} was taken as 20 μm, i.e. no abnormal grain growth during reheating. The austenite grain size has been evaluated at Ar_3 (700-800°C depending on cooling rate) using the microstructure-evolution model; the resultant ferrite grain sizes have been calculated directly from Eq.(13). For each rolling schedule, D^{α} is given for two cooling rates, one corresponding to natural air cooling appropriate to the plate thickness and the other to a realistic cooling rate obtainable via resort to accelerated-cooling techniques.

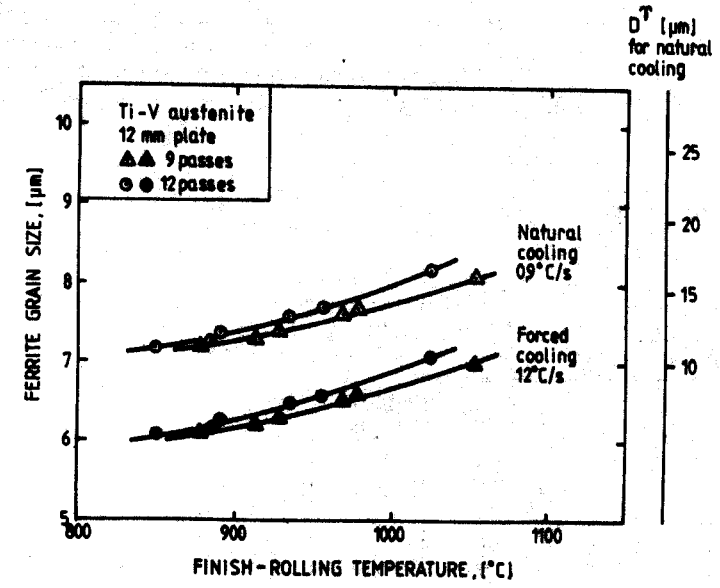
The results of the evaluations are summarized in Fig.10 as plots of D^{α} versus finish-rolling temperature; corresponding austenite grain sizes are also indicated on the diagrams. The following features are salient.

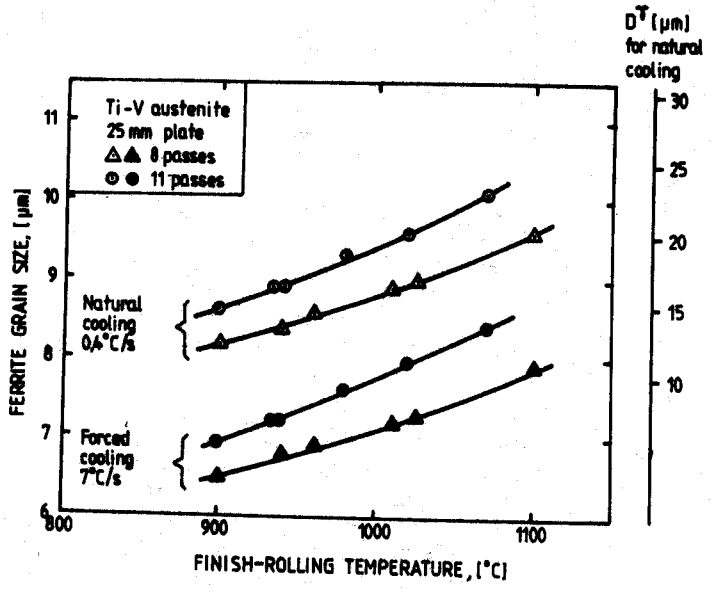
i) For a given pass-reduction scheme and cooling conditions, D^{α} is a unique function of FRT, independent of whether the latter is varied by altering the start temperature or via the inclusion of hold times or both.

ii) Fewer passes with a greater average reduction per pass is concomitant, naturally enough, with more efficient microstructural refinement. However, harder reductions are not as effective as, say, reducing the FRT by 100°C. The positive effect of greater average pass reductions is to some extent counteracted by the higher finishing temperature associated with a smaller number of passes.

iii) Even when the reductions during the final rolling passes are large and the FRT's are low, it does not seem possible, for the Ti-V steel of interest in this paper, to refine D^{γ} to below ~10 μm. The corresponding lower limit on D^{α} for accelerated cooling of the thinnest gauges (12 mm) is ~6 μm.

All results given in Fig.10 are calculated on the basis of an as-reheated austenite grain size (D_{start}) of 20 μm. Fig.11 shows the predicted effect of varying D_{start} on microstructure evolution during an otherwise identical rolling schedule (25 mm plate, start-rolling temperature = 1100°C, 11-pass schedule, 10s between passes with no hold). The large D_{start} (500 μm) is intended to simulate the situation where abnormal grain growth has occurred during reheating of a Ti-V steel (too high temperature and/or too long time at temperature). Clearly, the widely differing D_{start}'s have no bearing whatsoever on the final austenite grain size. Indeed, the latter is virtually identical in the two cases after as few as four passes. This means that for rolling schedules comprising more than 4-5 passes, the final austenite grain size is predicted to be insensitive to the extent of austenite grain growth during reheating. This does not mean to





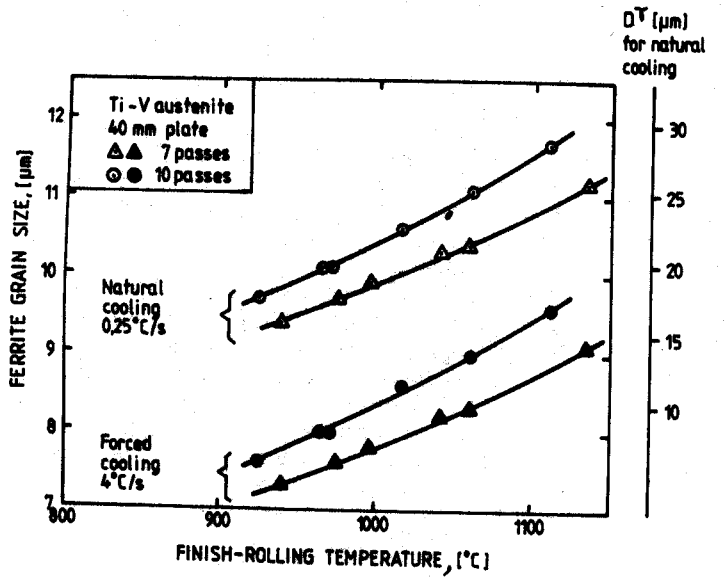


Fig.10 Computed dependence of D^{α} on finish-rolling temperature for a number of different rolling schedules (for further discussion, see text).

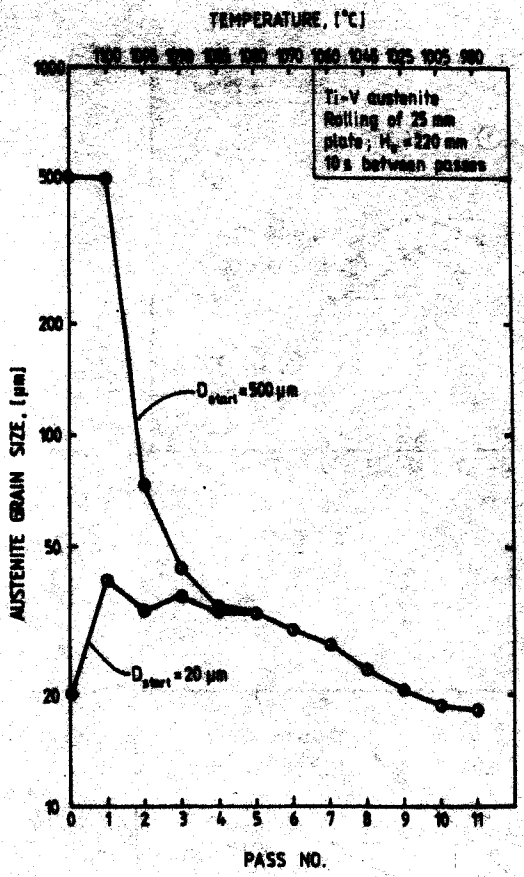


Fig.11 Illustrating the effect of as-reheated grain size on the microstructure evolution of Ti-V austenite during an otherwise identical rolling schedule (25 mm plate, pass-reduction scheme given in Table 4).

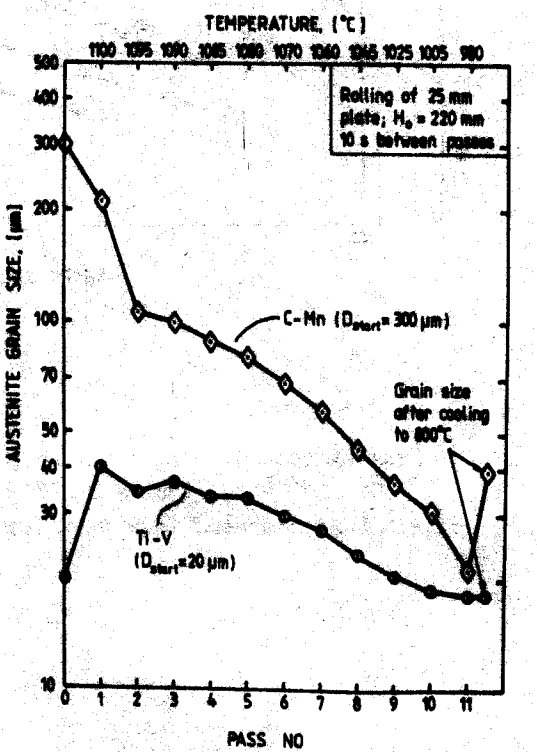


Fig.12 Comparison of microstructure evolution in C-Mn and Ti-V austenite subjected to the same rolling schedule (25 mm plate; pass-reduction scheme in Table 4).

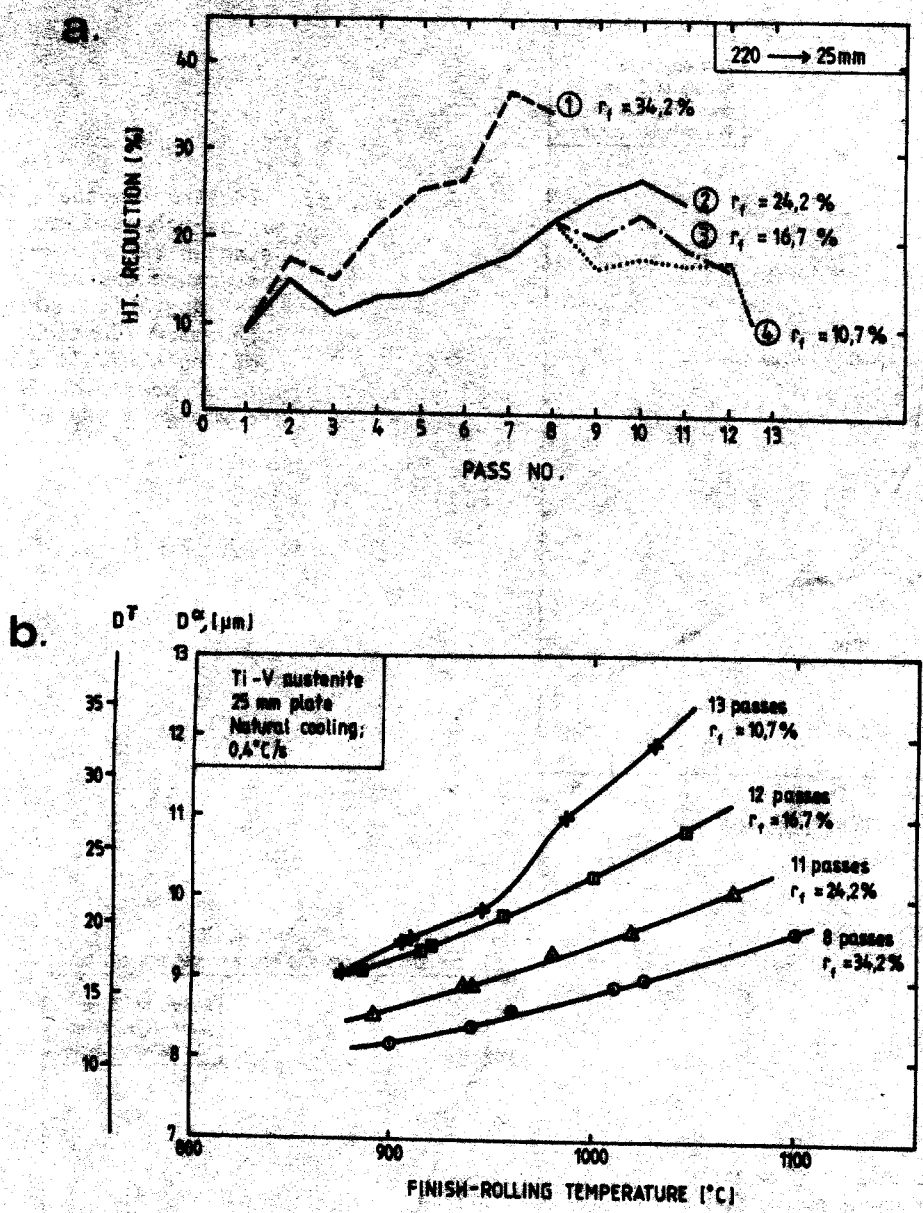


Fig. 13 a) Pass-reduction schemes (25 mm plate) for which the dependence of D^α on finish-rolling temperature shown in

b) has been computed (see text for further details).

say that the choice of reheating temperature is unimportant since, among other things, it determines the start temperature for rolling (at least in conventional practice). Note from Fig.11 that recrystallization of the austenite with fine D_{start} is predicted to engender grain coarsening at first and it is not before pass 10 that any refinement of the initial grain size is attained; on the other hand, the initially coarser-grained austenite is refined continuously during rolling.

It is instructive to compare the theoretical forecasts as regards microstructure development in C-Mn and Ti-V austenites subjected to the same rolling schedule. Such a comparison is given in Fig.12, which again refers to the rolling of 25 mm plate in 11 passes starting from 1100°C. D_{start} is taken to be 20 μm for the Ti-V steel and 300 μm for C-Mn austenite. After the final pass, the recrystallized grain size is actually not much larger in the latter instance than in the case of the Ti-containing austenite. However, for finishing at 980°C, the grain size of the C-Mn steel is increased appreciably during cooling to 800°C as a result of grain growth; in the Ti-V grade, on the other hand, the fine grain size from recrystallization after the final pass is retained down to Ar_3.

PLATE SCHEDULES CONCOMITANT WITH EFFECTIVE MICROSTRUCTURAL REFINEMENT AND ADEQUATE FLATNESS — As stated earlier, one method of achieving good shape control in plate rolling is to limit the rolling load during the final passes (small reductions). However, a high degree of overall microstructural refinement is favoured by large pass strains, especially in the final stages of rolling. Fig.13b summarizes the computed dependence of D^{α} on FRT for Ti-V steels subjected to rolling schedules which differ principally in the pass-reduction scheme during the later passes (details of the latter are given in Fig.13a). As a whole, the schedules refer to rolling of 25 mm plate with the FRT being varied in the some manner as before. The values for D^{α} given in Fig.13b are those evaluated assuming natural air cooling (0,4°C/s between 750 and 550°C). The curve for the smallest r_f (= reduction in last rolling pass) exhibits a distinct 'knee' in the temperature range 950-990°C. For finishing above 990°C, the austenite recrystallizes fully on cooling between FRT and Ar_3. It does not do so when finishing is effected below 950°C, being 50% recrystallized when FRT = 905°C and only 20% for 875°C finishing. In this latter instance, therefore, the final austenite grain size is rather determined by the situation in the last-but-one pass, for which the reduction is appreciably greater. The resulting D^{γ} and D^{α} are accordingly smaller than would be obtained by extrapolating the curve above 990°C to lower FRT's.

The behaviour described above derives from the rather slow recrystallization kinetics for Ti-V austenite after deformation to small strains (Eq.(5)) and suggests a means of combining fine microstructure with acceptable flatness after recrystallization rolling of these grades. The idea is to arrange the rolling schedule in such a way that the final passes are so light and effected at so low a temperature that recrystallization proceeds to only a limited extent between the FRT and the transformation start temperature. Obviously, it is, at the same time, necessary to ensure that the last pass in the rolling schedule for which recrystallization *does* go to completion during the inter-pass period, is sufficiently heavy that the resulting grain size, which will be very close to the final one, is fine. The subsequent light deformations, after which recrystallization is not completed, can thus be regarded as calibration passes aimed at attaining good dimensional tolerances and satisfactory flatness.

The examples given in this section serve to demonstrate the usefulness of microstructure-evolution models of the type described earlier, in designing schedules for recrystallization rolling capable of inducing optimal microstructural refinement.

CONCLUSIONS

i) Fine-grained microstructures in the as-hot-rolled condition can be produced in Ti-V microalloyed steels via the technique of recrystallization rolling, whereby static recrystallization of austenite between 900 and 1000°C is used to engender an effective microstructural refinement.

ii) The evolution of microstructure in Ti-V austenite during recrystallization rolling can be described quite accurately by means of a model based on the characteristics of recrystallization, recovery and grain growth which can be gleaned from simplified laboratory simulations.

iii) The kinetics of static recrystallization for Ti-V austenite are appreciably slower than for C-Mn steels under corresponding conditions; this effect, which is especially pronounced at small strains, is ascribed to Zener drag on moving grain boundaries from TiN-particles.

iv) The rate of normal grain growth after static recrystallization of Ti-V austenite is very low indeed over the entire hot-working temperature range.

v) For a given pass-reduction scheme and cooling rate following rolling, the ferrite grain size after recrystallization rolling is a unique function of finish-rolling temperature.

vi) In general, a deformation scheme involving fewer passes with greater average reduction per pass is not as conducive to ferrite grain refinement as a substantial reduction in finishing temperature.

vii) The grain size existing after reheating has no effect on the final austenite grain size

engendered via recrystallization rolling, as long as the rolling schedule involves more than 4-5 passes.

viii) The theoretical predictions from the model indicate that good shape control combined with fine microstructure is attainable via a judicious choice of the finish-rolling temperature and the reduction scheme adopted for the final 3-4 rolling passes.

ACKNOWLEDGEMENTS

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