# HIGH-TEMPERATURE MECHANISM OF DYNAMIC RECRYSTALLIZATION OF FERRITIC STEEL

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ABSTRACT.

Mechanical properties and microstructural changes during hot deformation in 25%Cr ferritic stainless steel have been studied. It was found that the dynamic recrystallization occurred in the material investigated at temperatures between 900 and 1100°C and strain rates of about $10^{-3}$ $s^{-1}$. The peculiarities of microstructure formation in the course of plastic flow have been considered. A model, describing the formation of recrystallized grains, has been proposed on the basis of experimental data.

INTRODUCTION.

As is generally known hot deformation in steels of different classes results in the formation of a fine-grained structure [1-5]. This seems to be associated with the development of dynamic recrystallization (DR) at low strain rates [1,2]. By the present time outer manifestations of DR and the relationship between deformation conditions and size of recrystallized grains have been studied rather well, while only few works have been devoted to the DR mechanism [6]. There are data in the literature describing some methods of formation of recrystallized grains in the course of the deformation process, connected with the local migration of sections of initial grain boundaries and with twinning at the early stages of plastic flow [7,9]. But, it is the DR mechanism, associated with dynamic recovery and subsequent transformation of subgrain boundaries into high angle ones of general type, that plays the main role [6,10-15].

There are several points of view on the processes resulting in the increase of the misorientation angle of grain low-angle boundaries during deformation. Thus, in works [10,12,13] there are given arguments in favour of a misorientation increase between the neighbouring subgrains resulting from lattice dislocation penetration into low-angle boundaries. Other authors [6,16] by the analogy with static recrystallization suppose that the coalescence of subgrains leads to the formation of DR nuclei. But the detailed experimental investigations of this process are not known. In the present study the mechanical properties and structural changes during hot deformation have been studied using the ferritic steel, the steel being the typical representative of body-centered cubic (BCC) metals with high stacking fault energy (SFE).

MATERIAL AND EXPERIMENTAL PROCEDURE

Specimens were cut from hot rolled rods of ferritic stainless steel (0,13%C-25%Cr-0,9%Ti-Fe) annealed at 1250°C for 1 hour with homogeneous microstructure (230$\mu$m mean grain size) to give the gauge dimension of $\phi$10x12mm. The specimens were deformed according to the axial compression scheme using an *Instron* 1185 testing machine within the temperature range of 900-1100°C. In order to fix the structure of specimens the latter were cooled by water spray immediately after the deformation completion. Metallographic analysis was performed using *Epiqant* automatic structure analyzer; fine structure was studied using BS-540 and JEM-2000EX electron microscopes.

RESULTS

The $\sigma$-$\epsilon$ curves for the steel, obtained as a result of the experiments, are analogous to the deformation diagrams given in a number of works [3,6,13,14]. Plastic deformation is accompanied with hardening, its value decreases with increasing temperature.

After small strain ($\epsilon$ =50%) the elongation of the initial grains in the direction of metal flow and the formation of network of boundaries take place. The depth of etching of these boundaries indicates that these are low angle boundaries [17]. The formation of recrystallized grains is observed with increasing deformation degree up to 75%. They take up about 40% of the specimen's volume, and their mean size amounts to be 40$\mu$m, which is 1, 2-1, 3 times as high as the size of the preceding subgrains established metallographically. New grains nucleate principally on the original boundaries and at triple junctions. The recrystallized grains are elongated in the direction of metal flow, the elongation ratio amounts to 1,67 on the average.

Electron-microscopic investigations confirmed these results. At the early stages of deformation ($\epsilon$ =15%) subgrain boundaries in the form of mutually intersecting dislocation networks are formed inside the initial grains, dislocation networks being both rhomblike (with an angle of dislocation intersection of about 71°), and hexagonal. Their misorientation angles are rather small (Fig.1). Most subboundaries, formed after attaining this degree of deformation, have misorientation angles of 1-2° and less. There is a small maximum corresponding to boundaries of $\theta$ =4-8°

The increase of deformation degree up to $\epsilon$ =50% results in the growth of misorientation angles of low-angle boundaries. The spectrum of misorientation angles of low-angle boundaries of deformation origin is shifted in the direction of high angles (Fig.1). Boundaries with $\theta$ > 16° which may be classified as high-angle boundaries of recrystallized grains, appear [18]. Misorientations between four sections, laying on one straight line, have been measured in one of the non-recrystallized grains. The distance between the neighbouring points has been chosen to be about 20$\mu$m. The dislocations are distributed uniformly along the chosen direction and their density is not changed. The following results have been obtained: $\theta_{1\text{-}2}$=0,47°; $\theta_{2\text{-}3}$=0,14°; $\theta_{3\text{-}4}$=0,19°; $\theta_{1\text{-}4}$=0,34°. It is evident that the vectorial sum of misorientations between neighbouring sections while moving from the first point is by more than 2,3 times less than the scalar one.

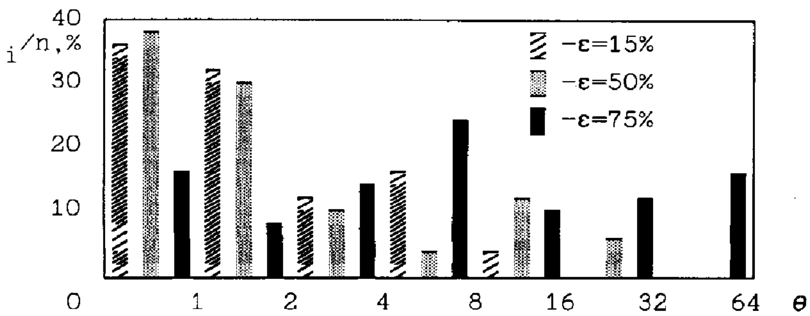


Figure 1.
Distribution of grain boundaries according to $\theta$ misorientation angles.

Subsequent deformation ($\epsilon = 75\%$) leads to the formation of high-angle boundaries with $\theta > 16°$ (Fig. 1). The microstructure of steel acquires the shape of interlacement of low-angle and high-angle grain boundaries. At the same time dislocation boundaries with $\theta = 1°$, which intensively migrate, are preserved in the material (Fig. 2a,b).



a) b)

Figure 2. TEM (transmission electron microscopy). $\epsilon = 75\%$, T=1100°C.

Let us consider it in detail. From the presented sections of fine structure it is evident that subboundaries move during deformation. The direction of migration leading to the mutual merging of subboundaries, can be determined by bulging. In the first case (Fig.2a) subboundaries form a triple junction. The migration of a subboundary with $\theta = 0{,}6°$ results in the change of $\theta$ of the boundary, it joins from 2,9 to 3°. In the second case (Fig.2b) there is a section limited by subboundaries with misorientations of 9,1°; 1,2°; 0,9°. In the given case a tendency is observed to tighten the marked section to the effect of the formation of a triple junction of boundaries with misorientations: 9,4°; 8,5°; 13°. The migration of subboundaries results in the smooth increase of the middle misorientation of the boundaries of deformation origin in both cases under consideration.

## DISCUSSION

As seen from the data presented the phenomenology and kinetics of DR development in ferritic steels are not typical of most materials [1,2,6-11]. The shape of $\sigma$-$\epsilon$ curves in the general case is characteristic of dynamic recovery [2], while the formation of recrystallized grains after large strain degrees simultaneously in great amounts and all over the volume of initial grains is observed only in materials with BCC lattices [3,19,20].

The analysis of the results of microstructural studies has shown that this is associated with the effect of the DR specific mechanism in ferritic steel during deformation in the high temperature region. At the early stages of plastic flow single-row accumulations of screw dislocations are formed in the vicinity of grain boundaries in the initial grains. Under the effect of mutual attraction forces and due to the high mobility they are easily superimposed one over the other and form networks with a stable configuration. It is known [18,21] that such grains have long effective stress fields, as the angle between the Burgers vectors of screw dislocations in ferritic steels may have only two values, 48° and 71°, and, hence, the network, consisting of two dislocation sets, will always have a noncompensated stress field [18]. The mobility of such subboundaries is determined by the conservative motion of dislocations. Transformation of a rhomboid network into a hexagonal one by splitting one junction of screw dislocations into two junctions is observed, the process is followed by the formation of a low-mobility edge dislocation according to the reaction [18,21]:

$$a/2[111] + a/2[\overline{1}\overline{1}\overline{1}] = a[100].$$

But for the given material, characterized by a SFE high value, it is the nonconservative dislocation motion that plays an important role in the migration process of low-angle boundaries. The latter is especially to be taken into account at high temperatures.

Another important peculiarity of the dislocation structure is that only up to 25-30% of dislocations have interoverlapping Burgers vectors and annihilate when coming across each other. This follows from the measurements of misorientation change within one grain. A uniform distribution of dislocations smoothly changes the crystallite misorientation to a small but finite value, the algebraic sum of misorientation angles during transition from one section to another being more than the misorientation angle between the initial and last points. Hence it appears that in the general case all components of the dislocation density tensor, especially the diagonal ones, which are responsible for screw dislocations, have values different from zero, though some overlapping of Burgers vectors takes place. The latter makes possible the formation of networks and subboundaries from screw dislocations as a result of the effect of mutual attraction forces [18]. The preferred formation of dislocation networks, and then subgrains and recrystallized grains in triple junctions and on the original boundaries, is associated with the local accumulation of dislocations there [10,21].

In the process of deformation the individual subboundaries lose their unique ability to migrate. Precipitation of titanium carbides plays here a definite role, they serve as stoppers to boundary migration according the Zener mechanism [22]. Such subboundaries attract both the individual dislocations and other twist subboundaries, which have mobility and corresponding Burgers vectors. Despite the partial overlapping of Burgers vectors it provides the stable growth of dislocation density in subboundaries and of misorientation angle with increasing deformation degree, which in its turn decreases still more the mobility of the given boundary [18].

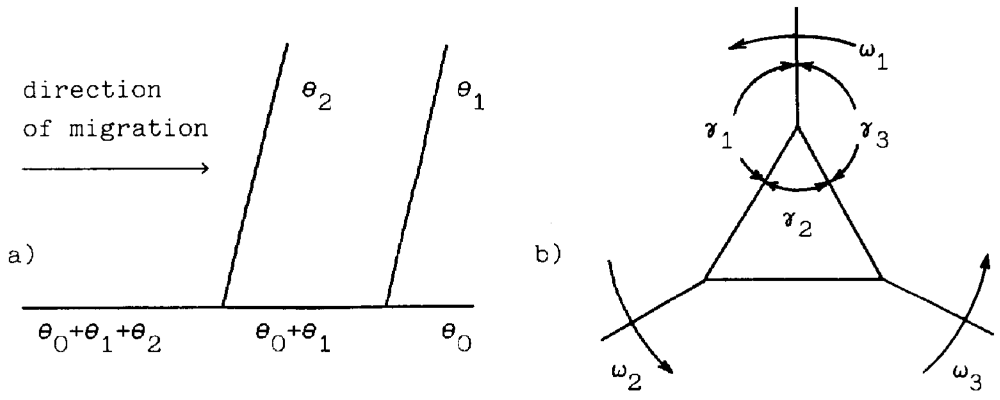


Figure 3. a) misorientation change of the joining boundaries during their migration;
b) formation of stable boundary configuration.

Let us consider this process in detail (Fig.3a,b). The observed migration of low-angle boundaries under the effect of both the mutual attraction forces, and the external stress field leads to their merging with a change of misorientation. The migration of a subboundary with a misorientation angle equal to $\vartheta_1$ results in a change of misorientation of the boundary, in junction with it, from $\vartheta=\vartheta_0$ to $\vartheta=(\vartheta_0+\vartheta_1)$, the next subboundary with $\vartheta=\vartheta_2$ changes the misorientation from $\vartheta=(\vartheta_0+\vartheta_1)$, to $\vartheta=(\vartheta_0+\vartheta_1+\vartheta_2)$, etc. (Fig.3a). The concentration of total misorientation in different parts of the initial grain in low-mobility boundaries takes place. Migration and merging of low-angle boundaries eliminate "minor instabilities" at the cost of local motion of the boundaries and provide the equilibrium angles of the junction $\gamma_i \rightarrow 120°$ (Fig. 3b). "The instability at large" is eliminated by the migration of boundaries and individual dislocations in the long-range stress fields of low-angle boundaries, which can be interpreted as disclinations for long distances, leading to the minimization of energy at the expense of $\Sigma\omega_i \rightarrow 0$ [18]. An equilibrium network of middle and high-angle grain boundaries is formed.

It is interesting to note that one can experimentally observe the merging of migrating boundaries having misorientation angles of about 1° with subboundaries which have far greater angle $\theta$ and which do not demonstrate any signs of motion. That is, during plastic deformation in ferritic steels there occurs the formation of networks out of one or two sets of screw dislocations having the misorientation angle of 0,5-1,5°, which then, merging with subboundaries with large angle $\theta$, provide the transformation of the latter into middle and high-angle boundaries. The investigations performed allow us to believe unambiguously that in the given case this process is the main mechanism of formation of the high-angle grain boundaries of general type during deformation.

In favour of the present model of DR speaks the fact that the recrystallized grains have high values of elongation ratio in the upsetting direction. It means that boundaries of such grains are formed continuously in the course of deformation as networks of low-angle boundaries of deformation origin. As deformation develops the subgrains formed are deformed together with the initial grains and, under the effect of the above described mechanism of DR, are transformed into grains with high-angle boundaries of the general type. This is associated with the simultaneous formation of a great number of recrystallized grains after large deformation degrees (which is not typical of most materials) and with the shape of $\sigma - \epsilon$ curves characteristic of dynamic recovery.

## CONCLUSIONS

The formation of dislocation networks and their subsequent merging, resulting in the gradual transformation of individual low-angle boundaries into high-angle ones during deformation, is the main mechanism of DR in the present materials at high temperature. The phenomenology and kinetics of DR in ferritic steel (not typical of other materials) are explained by the effects of this mechanism.

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