STRUCTURAL CHANGES IN CORROSION-RESISTANT STEELS DURING HOT DEFORMATION

A. N. Belyakov and R. O. Kaibyshev

UDC 669.14.018.8:621.77.016.2

Structural changes during the hot deformation of ferritic, austenitic, and austenitic–ferritic steels are investigated. The temperature–time conditions of deformation for which dynamic recrystallization develops are established. It is demonstrated that in the early stages of deformation, a developed structure, which is then transformed into recrystallized grains, forms in steels. The characteristic features of microstructure formation during the plastic flow of different classes of steels are addressed.

One of the effective means of raising the level of mechanical properties in steels is to produce a superfine-grain structure in them [1]. The traditional method, which consists of accelerated austenitization and a short holding time at temperatures somewhat above the critical, is inapplicable to corrosion-resistant steels of the austenitic, ferritic, and austenitic–ferritic classes. Static recrystallization does not make it possible to produce a grain with a size of less than 10 μm [2]. Hot deformation is the most universal method that ensures the formation of a fine-grain structure in steels [3, 4]. The formation of fine grains during this deformation is explained by the development of dynamic recrystallization (DR).

At the present time, it is established that dynamic recrystallization develops in simple carbon, high-strength, low-alloy, austenitic, maraging, and tool steels during hot deformation in the austenite region [5-7]. Maki et al. [8] and Chandra et al. [9] have demonstrated the development of dynamic recrystallization in ferritic and austenitic–ferritic steels. Based on analysis of experimental data, however, Glover and Sellars [10] and Sellars [11] concluded the impossibility of recrystallization in low-purity Fe_α in ferritic steels owing to the case with which dynamic recovery develops; this is associated with the high stacking energy of defects in Fe_α. It should be noted that indications of dynamic recrystallization, and also the relation between the deformation conditions and size of the recrystallized grains is well understood in different classes of steels. There are very few studies devoted to the mechanism of dynamic recrystallization. The influence exerted by the chemical and phase composition of the steels on the mechanism of dynamic recrystallization is undetermined.

In our study, we investigated structural changes in ferritic, austenitic, and ferritic–austenitic steels (Table 1) during hot deformation. The absence of polymorphic transformations under the temperature-rate conditions of deformation investigated and the use of quenching immediately after cessation of deformation made it possible to fix the high-temperature state of the steels.

Specimens 16 mm in diameter and 15-mm long were cut from bars annealed at 1200-1250°C for 4 h and deformed by upsetting on an "Instron" testing machine in the t_d = 900-1100°C interval and at strain rates $v_d = 10^{-2}$-10^{-3} sec^{-1}. After cessation of deformation, the specimens were cooled in a water jet. A metallographic analysis was conducted under an optical "Neophot-2" microscope and automated "Epiquant" structural analyzer. The fine structure was examined under a "Tesla BS-540" electron microscope. Histograms of the distribution of disorientation angles θ_{min} were constructed for a mass of 30 boundaries.

The yield-stress–strain curves for all three steels are similar to those observed earlier [3, 5, 8]. After a rapid increase in yield stress in the initial stage, further deformation of the specimens is accompanied by minor hardening. Maximum hardening is observed for the austenitic steel at 900°C.

Hot deformation of the steel 15Kh25T specimens contributes to the formation of a developed substructure in the early stages of plastic flow. After deformation with a degree ε = 50%, the substructure is exposed metallographically in the alloys (Fig. 1a). Without exerting an influence on the size of the subgrains, further deformation leads to the formation of large-angle boundaries. New grains are generated preferentially on the boundaries of the initial and in triple joints. The average size and

Institute of Problems of Metal Superplasticity, Ufa. Translated from Metallovedenie i Termicheskaya Obrabotka Metallov, No. 5, pp. 19-23, May, 1992.

 0026-0673/92/0506-0324$12.50

TABLE 1

Steel	Class of steel	Element content, %						D, μm
		C	Cr	Ni	Ti	Mo	Cu	
15Kh25T	Ferritic	0,15	26,3	—	0,9	—	—	230
12Kh18N9	Austenitic	0,12	18,4	9	—	—	—	260
ÉK-72	Austeni-tic-fer-ritic	0,15	20,5	5,5	—	2,9	1,6	32/38

Note. The size D of the austenite and ferrite grains is given in numerator and denominator, respectively. The ratio of the specific volume of austenite and ferrite is 50:50.

Fig. 1. Microstructure of corrosion-resistant steels after deformation at 1100°C at rate $\dot{\varepsilon} = 8.3 \cdot 10^{-4}$ sec^{-1}: a) 15Kh25T, $\varepsilon = 50\%$, ×100; b) 12Kh18N9, $\varepsilon = 75\%$, ×150; c) ÉK-72, $\varepsilon = 75\%$, ×250.

specific volume of recrystallized steel 15Kh25T grains after hot deformation are presented in Table 2. It should be noted that the specific volume of recrystallized grains after increasing the degree of deformation to $\varepsilon = 75\%$ approaches that of the specific volume of the substructure exposed metallographically when $\varepsilon = 50\%$. In this case, structural inhomogeneity is observed in the steel. Subgrains of approximately the same size are detected in addition to the recrystallized grains. An increase in the deformation temperature of steel 15Kh25T effects an increase in the size of the recrystallized grains. A distinct correlation is established between the average size of the recrystallized grains and the Zener–Hollomon parameter "Z."

The data derived from the electron-microscopic examinations (Fig. 2) confirm the results of the metallographic analysis. In the initial stages of deformation, a network of subgrain boundaries (Fig. 2a) in the form of dislocation walls forms within the initial grains; this suggests very small disorientation angles of these boundaries. This confirms the results of the measurement of disorientation angles (Fig. 3a-c). It should be noted that in steel 15Kh25T, the subboundaries have high mobility. Buckling of the segment of the low-angle boundaries is observed in the vicinity of carbide particles. Subboundaries, which are not ruptured in the wall of individual dislocations, appear when the degree of deformation is increased. This is associated with a shift in the spectrum of angles of boundary disorientations of the deformed origins in the direction of large angles as the degree of deformation is increased. Thus, the maximum corresponding to the boundaries with θ_{min} = 4-8° after upsetting with $\varepsilon = 15\%$ is shifted toward boundaries with $\theta_{min} > 16°$ after deformation with $\varepsilon = 50\%$. Hence, it is possible to conclude that the subgrain boundaries exposed by the metallographic method have disorientation angles θ_{min} = 8-16°. For large degrees of deformation ($\varepsilon = 75\%$), interlacings of subboundaries and high-angle grain boundaries are formed in the structure of the ferritic steel. This suggests both a form of fine structure in the material, and a spectrum of disorientation angles of the boundaries that have been reformed. A low dislocation density ($\rho = 10^9$ cm^{-2}), which is slightly dependent on the degree of deformation, is a characteristic feature of the ferritic steel. It should be noted that the size of the subgrains depends on the deformation temperature, just as the size of the recrystallized grains. The size of the subgrains in steel 15Kh25T increases from 4-6 to 25-30 μm as the deformation temperature is raised from 900 to 1100°C.

Under similar temperature-rate conditions, a different pattern of the formation of recrystallized grains is observed in the austenitic steel 12Kh18N9 than in the ferritic steels. The specific volume and average size of the new grains in the austenitic steel, as in the ferritic steel, increase with increasing temperature or decreasing strain rate (Table 2). At the same time, two temperature regions that are distinguished one from the other by sites of recrystallized-grain formation can be isolated for steel 12Kh18N9. At t_d = 900°C, the appearance of new grains is observed both in the center of the initial grains, and on their boundaries.

TABLE 2

Steel	t_1 °C	ε, %	Phase	d_{rec} μm	V_{rec} %
15X25T	1000	75	Ferrite	18/33	32
	1100			30/50	41
12X18H9	1000	50	Austenite	6/8	11
	1100			13/15	15
EK-72	1000	50	Ferrite Austenite	4 8	25 40
	1100		Ferrite	12/14	75
			Austenite	16/23	86

Notations: d_{rec} and V_{rec} are the average size and specific volume of recrystallized grains.
Note. d_{rec} in the longitudinal and transverse directions to the axis of upsetting are given in the numerator and denominator, respectively.

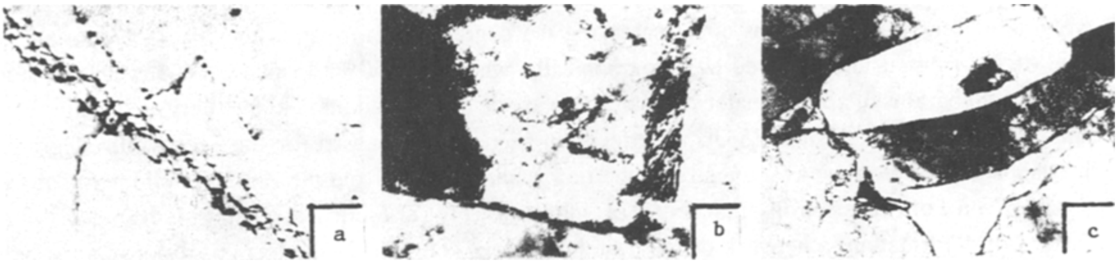

Fig. 2. Fine structure of corrosion-resistant steels after deformation ($\dot{\varepsilon} = 8.3 \cdot 10^{-4}$ sec^{-1}): a) 15Kh25T, $t_d = 1100$°C, $\varepsilon = 15\%$, ×30000; b) 12Kh18N9, $t_d = 900$°C, $\varepsilon = 50\%$, ×20000; c) 12Kh18N9, $t_d = 1100$°C, $\varepsilon = 15\%$, ×5000.

During deformation in a region of higher temperatures, recrystallized grains are formed only on the boundaries of the initial grains or on the site of former annealing twins (see Fig. 1b). Recrystallization grains are observed at lower degrees of deformation as the temperature increases.

The results of investigation of the fine structure confirmed the presence of two temperature regions of deformations for the austenitic class steel 12Kh18N9. The formation of a different type of dislocation structure in the early stages of plastic flow is characteristic for these regions.

At a relatively low temperature $t_d = 900$°C, plastic deformation leads to the formation of three different types of substructures. The most characteristic of these is the fragmented structure (Fig. 2b), which is described by Rybin [12]. "Knife-edge" boundaries, the length of which are the same as the size of the initial grain, are an indication of this structure. The width of the interval between these boundaries, within which finer fragments with a size of 1-2 μm are disposed, ranges from 4 to 15 μm. These boundaries provide "dislocation" contrast on electron-microscope photographs, and their disorientation angles amount to several tens of degrees; this agrees with Rybin's data [12], which were derived from alloys subjected to high degrees of cold deformation. Another type of substructure is the cellular structure, which consists of flat balls of twisted dislocations. It is formed both in the unrecrystallized portion of the initial grains of material, and also within the fragments, as well as in the body of the subgrains and recrystallized grains. Moreover, low-angle boundaries are observed in the substructure. For small

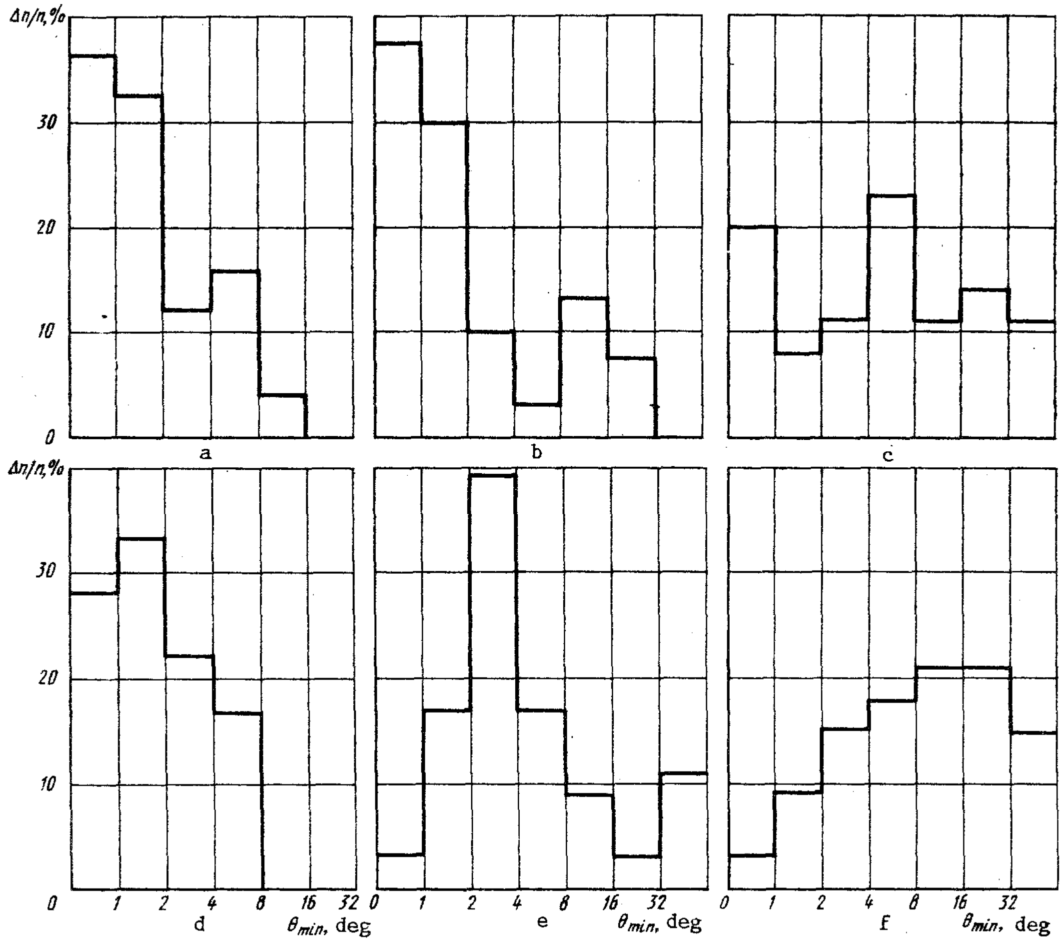

Fig. 3. Distribution of grain boundaries of steels 15Kh25T (a-c) and 12Kh18N9 (d-f) (n is number of cases) with respect to their disorientation angles θ_{min} (t_d = 1100°C, $\dot{\varepsilon} = 8.3 \cdot 10^{-4}$ sec^{-1}): a, d) ε = 15%; b, e) ε = 50%; c, f) ε = 70%.

degrees of deformation, they are dislocation walls. When ε = 50% and higher, these boundaries provide extinction contrast; this suggests an increase in their disorientation angles during deformation.

An increase in the deformation temperature to 1000°C sharply reduces the number of fragments, cells, and balls of dislocations, while at 1100°C, they are not detected. The formation of a subgrain structure is the basic stage in the structural changes preventing the formation of recrystallized grains in the region of high temperatures. Investigation of the fine structure indicated that subgrains with an average size of 6 μm are formed after deformation at 1100°C with a degree ε = 15% (Fig. 2c). Their disorientation angles are relatively small (Fig. 3d-f). At 1100°C, the curve of the distribution of disorientation angles shifts in the direction of higher values with increasing degree of deformation, and forms, as in steel 15Kh25T, an interlacing of low- and high-angle grain boundaries. In contrast to the ferritic class of steel, however, the size of the subgrains that are formed is 2-3 times smaller than the recrystallized ones. The presence of a subgrain structure in both the unrecrystallized segments of the structure, and also the newly formed grains is characteristic in this case.

Structural changes in the austenitic–ferritic steel ÉK-72 during plastic flow possess certain characteristic features. Nonuniform deformation, which is localized in the α-phase, is observed. As a result, the elongation factor of the γ-phase does not exceed 1.5 under all temperature-rate conditions investigated when a degree of deformation ε = 75% is attained (see Fig. 1c). The metallographic and electron-microscopic investigations indicated that the process of recrystallized-grain formation in the austenitic–ferritic steel occurs more rapidly than in single-phase steels. Even after upsetting with ε = 15% at 1100°C, a subgrain structure is formed in the α- and γ-phases. The subgrains in the ferrite are equiaxial with an average size of 6-8 μm, and long subboundaries, located 2-3 μm apart, are observed in the austenite. The presence of "knife-blade" boundaries is characteristic preferentially in the γ-phase. The subgrains in the austenite become equiaxial under further deformation. The vigorous formation of new grains occurs (Table 2). They develop over the entire volume in the α-phase, and preferentially in the vicinity of boundaries and along twins in the γ-phase. When ε = 75%, nearly 80% of the ferritic phase consists of recrystallized grains. An increase in the strain rate or a lowering of the temperature will effect a reduction in the size of the grains and subgrains and an increase in the dislocation density. A characteristic feature of the austenitic–ferritic steel is the formation of round disperse particles of austenite with a size of 2-5 μm in the ferrite component during plastic flow.

Analysis of the results of mechanical tests and microstructural investigations indicates that during hot deformation, dynamic recrystallization develops in ferritic, austenitic, and austenitic–ferritic corrosion–resistant steels. The size and specific volume of the recrystallized grains are determined by the Zener–Hollomon parameter "Z." The value of this parameter also determines the active mechanism of dynamic recrystallization.

When the austenitic and austenitic–ferritic steels are deformed in the region of high temperatures and the ferritic steel over the entire temperature investigated, the mechanisms of dynamic recrystallization are similar and consist in the formation of a subgrain structure in the early stages of plastic flow. Multiple slip occurs in the fcc, and, especially the bcc lattice of the steels. It develops through a large number of slip systems and occurs uniformly throughout the entire volume of the initial grain. This ensures the easy formation of bulk networks of subboundaries and leads to the appearance of macroregions surrounded by low-angle boundaries. In the early stages of plastic flow, the disorientation angle of the reformed subboundaries is low and does not exceed several degrees. The dislocation density in the subboundaries increases with increasing degree of deformation due to lattice dislocations, and, accordingly, the disorientation angle of the subboundaries increases. The rate of this process increases appreciably as a result of the plural character of dislocation slip. In this case, the low-angle boundaries are transformed into high-angle ones [13]. Dislocations accumulate locally in the vicinity of the initial grain boundaries and in the triple joints, since plane defects in the crystal lattice play the roll of retainers. The most favorable conditions are created for the formation of a network of low-angle boundaries with a gradual increase in their disorientation angles; the first recrystallized grains therefore appear in the vicinity of the boundaries of the initial grains.

It should be noted that the mechanism of dynamic recrystallization has its own characteristic features in each steel. The higher value of the defect-packing energy in steel 15Kh25T contributes to greater mobility of both the individual dislocations, and also the entire subboundaries; this is the cause of the formation of very course subgrains. The linking of subboundaries into one boundary evidently plays a major role in this steel with a high defect-packing energy. The realization of this mechanism, which is based on the vigorous migration of subboundaries during hot deformation, determines the development of dynamic recrystallization at high degrees of deformation and the simultaneous formation of an appreciable amount of recrystallized grains. The low density of linear defects in this steel is associated with their "convergence" at low- and high-angle boundaries as a result of high dislocation mobility.

As in the previous case, networks of subboundaries are formed in the austenitic steel 12Kh18N9 at high deformation temperature (1000-1100°C). The defect-packing energy in this steel is considerably lower, however, than that in steel 15Kh25T, and the dislocation density is therefore appreciably higher. The increase in the disorientation angles of the low-angle boundaries during high-temperature deformation is most likely ensured by lattice dislocations of different systems that enter the dislocation walls near the initial boundaries. This mechanism for disorientation-angle growth in steel 12Kh18N9 ensures a much higher rate of formation of recrystallized grains than in steel 15Kh25T. In this case, the appearance of the first new grains in steel 12Kh18N9 is observed at lower degrees of deformation. The reduced dislocation mobility in this steel causes dynamic recovery to occur in the recrystallized grains. Another characteristic feature of the austenitic steel is the formation of two different types of substructure in the initial stages of plastic flow in regions of high and moderate deformation temperatures. When t_d = 900°C, a fragmented structure, which Rybin describes in detail [12], form is in steel 12Kh18N9. It is formed as a result of the passage of disclination dipoles and fragmentation of the volume located between the "knife-edge" boundaries. Crystallographic slip leads to the formation of a cellular structure and a certain amount of subboundaries; these processes cannot, however, exert a significant influence on dynamic recrystallization. In the low-temperature region, new grains in the austenitic steel appear at points of intersection of disclination dipoles [12], or in the vicinity of initial boundaries in accordance with the above-discussed mechanism. In the first case, transformation of the "knife-edge" boundaries occurs at a common type of boundary, and new grains form in the volume of the initial boundaries. The mechanism responsible for the transformation of fragmentary to high-angle boundaries remains unclear.

The characteristic features of dynamic recrystallization in the dual-phase steel ÉK-72 are associated with the influence of the structural components. Localization of deformation in the α-phase occurs as a result of a lower level of yield stresses in the ferrite than in the austenite. In this connection, the start of crystallized-grain formation is observed in steel ÉK-72 at appreciably lower degrees of deformation than in steel 15Kh25T. The austenite particles in the α-phase, which are segregated during plastic flow, prevent the growth of recrystallized grains, and the small initial size of the ferritic component dictates a more complete recrystallization due to an increase in the number of points of origin of new grains. In the early stages of deformation, a substructure is formed in the austenitic phase over the entire temperature interval investigated primarily as a result of dislocation walls. The percentage of the fragmented volume increases with increasing degree of deformation. After achieving $\varepsilon = 75\%$, the vigorous formation of a knife-edge" boundaries in the unrecrystallized volume occurs in the γ-phase even when $t_d = 1100°C$ due to the movement of dislocation dipoles. Structure formation will begin to be determined by collective processes.

CONCLUSIONS

1. The hot deformation of corrosion-resistant steels of the ferritic, austenitic, and austenitic–ferritic classes in the 900-1100°C interval at rates $\dot{\varepsilon} = 10^{-2}$-$10^{-3}$ sec^{-1} results in dynamic recrystallization.

2. Various types of substructures are formed in the initial stages of plastic flow, depending on the deformation conditions and the type of crystal lattice of the steels, and, consequently, different mechanisms of dynamic recrystallization are realized.

3. Substructure formation in the initial grains and gradual transformation of low- to high-angle boundaries are the mechanisms responsible for the formation of recrystallized grains in the ferritic steels over the entire temperature region investigated, and in the austenitic-class steel at high temperatures.

LITERATURE CITED

1. M. I. Gol'dshtein, S. V. Grachev, and Yu. G. Veksler, Special Steels [in Russian], Metallurgiya, Moscow (1985).
2. S. S. Gorelik, Recrystallization of Metals and Alloys [in Russian], Metallurgiya, Moscow (1978).
3. M. L. Bernshtein, V. A. Zaimovskii, and L. M. Kaputkina, Thermomechanical Treatment of Steel [in Russian], Metallurgiya, Moscow (1983).
4. O. A. Kaibyshev, Superplasticity of Commercial Alloys [in Russian], Metallurgiya, Moscow (1984).
5. T. Maki, K. Akasaka, K. Okuno, and I. Tamura, "Dynamic recrystallization of austenite in 18-8 stainless steel and 18 Ni maraging steel," Trans. Iron Steel Inst. Jpn., **22**, 253-261 (1982).
6. S. Sakui, T. Sakai, and K. Takeishi, "deformation of austenite in a plain carbon steel," Trans. Iron Steel Inst. Jpn., **17**, 718-725 (1977).
7. C. Ouchi and T. Okita, "Dynamic recrystallization behavior of austenite in Nb-bearing high-strength low-alloy steels and stainless steel," Trans. Iron Steel Inst. Jpn., **22**, 543-551 (1982).
8. T. Maki, S. Okaguchi, and I. Tamura, "Dynamic recrystallization in ferritic stainless steel," ICSMA6, **1**, 529-534 (1982).
9. T. Chandra, D. Bendeich, and B. P. Dunne, "Dynamic recovery and recrystallization in a duplex stainless," ICSMA6, **1**, 505-510 (1982).
10. G. Glover and C. M. Sellars, "Recovery and recrystallization during high temperature deformation of iron," Metall. Trans., **4**, No. 3, 765-775 (1973).
11. C. M. Sellars, "Recrystallization of metals during hot deformation," Philos. Trans. R. Soc. London, Ser. A, **A288**, 147-158 (1978).
12. V. V. Rybin, Large Plastic Deformations and Failure of Metals [in Russian], Metallurgiya, Moscow (1986).
13. N. G. Zaripov, A. R. Vaganov, and R. O. Kaibyshev, "Dynamic recrystallization of a magnesium alloy," Fiz. Met. Metalloved., **63**, No. 4, 774-781 (1987).