
2 Formation of a Dislocation Substructure by Plastic Deformation

The plastic deformation of metals and alloys occurs by the elementary processes of nucleation, movement, interaction, and annihilation of crystal lattice imperfections (vacancies, dislocations, grains, and twin boundaries). The dislocations start moving when the applied stress exceeds the yield strength of the metal. Dislocations of opposite sign may annihilate upon meeting, whereas the rest of the dislocations interact and combine into *dislocation complexes* or *tangles*. The increasing number of moving and interacting dislocations results in barely movable complexes, which form a *dislocation substructure*.

The character of dislocation substructure depends on several factors: the deformation conditions (temperature, strain rate, amount of deformation), the crystal structure, and the SFE of the metal or alloy. The strain rate and temperature are complementary variables—the effect of a strain rate increase is equivalent to a temperature decrease. Besides the diffusion-assisted processes, the temperature influences the type of dislocation reactions because of the strong temperature dependence of the SFE value. In the case of metals and alloys, which don't undergo temperature-induced phase transformations, the value of the SFE increases with the increasing temperature.

The dislocation substructure is of special importance, not only for the metal behavior under deformation, but also for the properties of the deformed metal.

2.1 FORMATION OF A DISLOCATION SUBSTRUCTURE AT ROOM TEMPERATURE IN METALS OF HIGH STACKING FAULT ENERGY

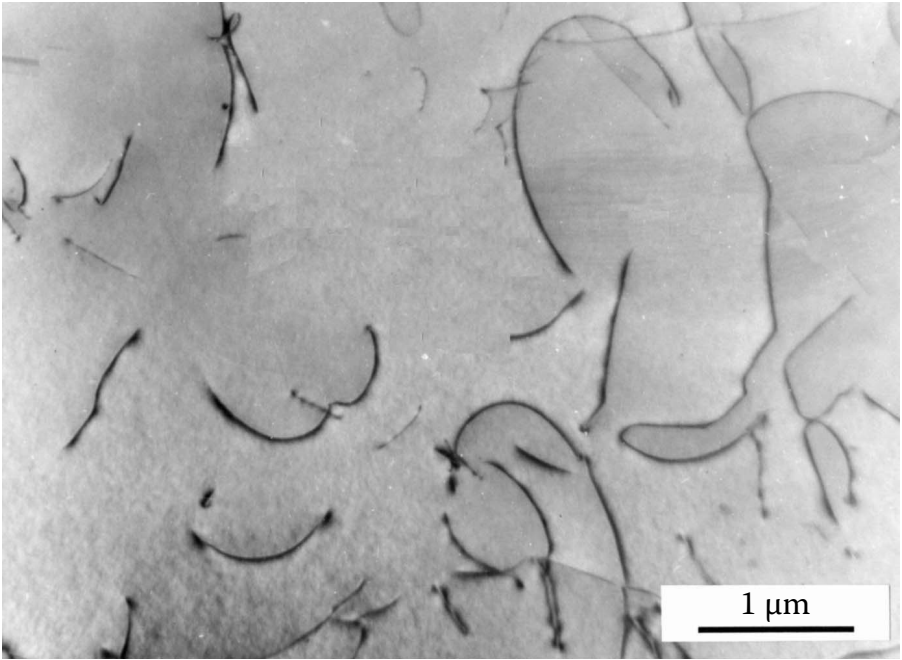


FIGURE 2.1 The density of dislocations in well-annealed metals and alloys is rather low. In metals and alloys of high SFE, such as aluminum, the unit dislocations can move on all slip planes of the crystal lattice conservatively (by slip) and nonconservatively (by cross-slip). The easy cross-slip of individual dislocations results in three-dimensional dislocation distribution.
Quenched aluminum.

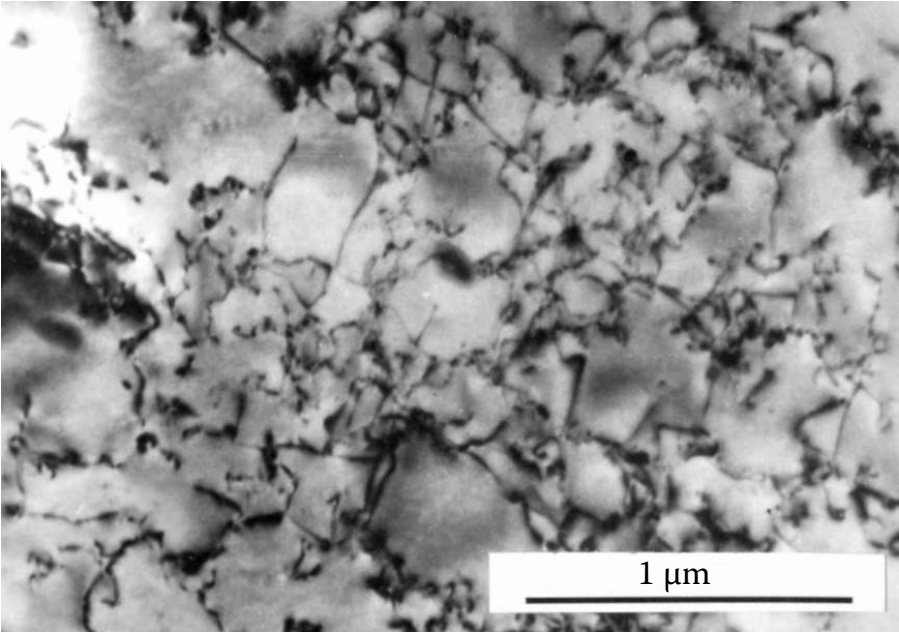


FIGURE 2.2

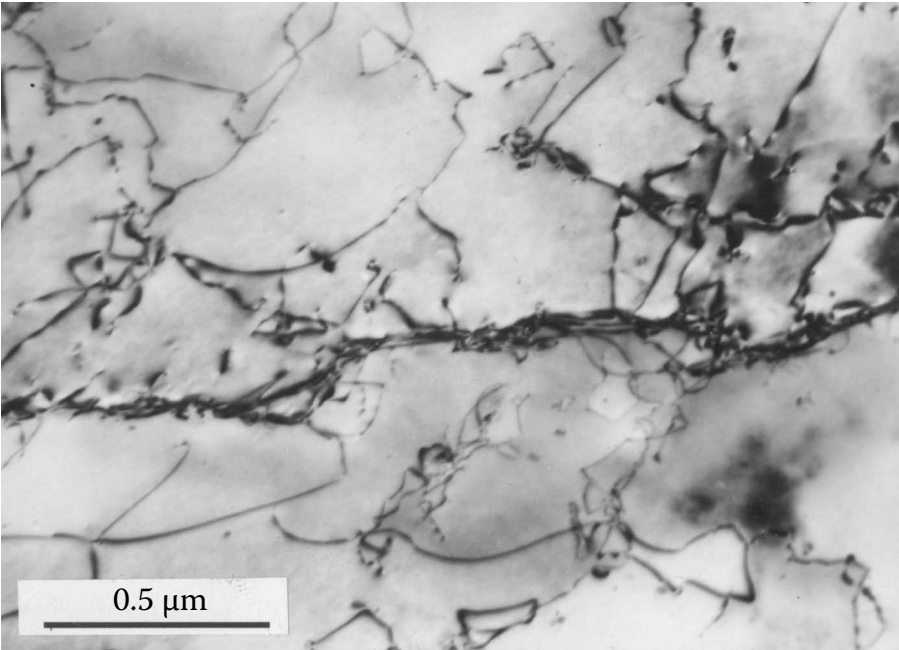


FIGURE 2.3

FIGURE 2.2 As in the high-SFE aluminium, the dislocations in high-SFE ferrite have three-dimensional distribution in spite of the different crystal lattices, fcc for aluminium and bcc for ferrite.

Plain carbon steel, mould casting. The dislocation density in this case is higher than in the quenched aluminum, shown in Figure 2.1, because of the more complicated cooling conditions occurring during the casting.

FIGURE 2.3 When an external stress is applied to the crystal, the dislocations start moving, annihilating, multiplying and interacting, and trying to form configurations of maximum stability, that is, of minimum energy. For lower amounts of cold deformation, the number of moving dislocations is small and their interaction results in the simplest of complexes—dislocation tangles.

Aluminum after 2% cold deformation.

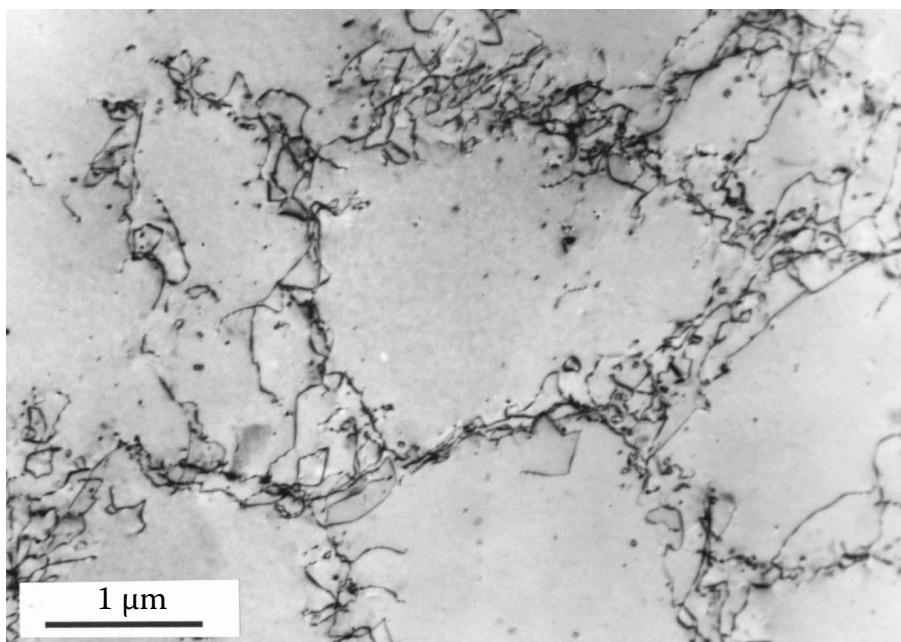


FIGURE 2.4 The number of dislocations increases with increasing amount of deformation. The greater the number of dislocations that have participated in the formation of tangles, the more perfect and stable the tangles become. They gradually transform into *dislocation walls*—interconnected dislocation complexes—which divide the volume into *dislocation cells* with an interior almost free of dislocations. The term *cell structure* has been adopted in physical metallurgy for this type of cold-deformed structure in analogy to the term *cell structure* used for plants.

Aluminum after 5% cold deformation.

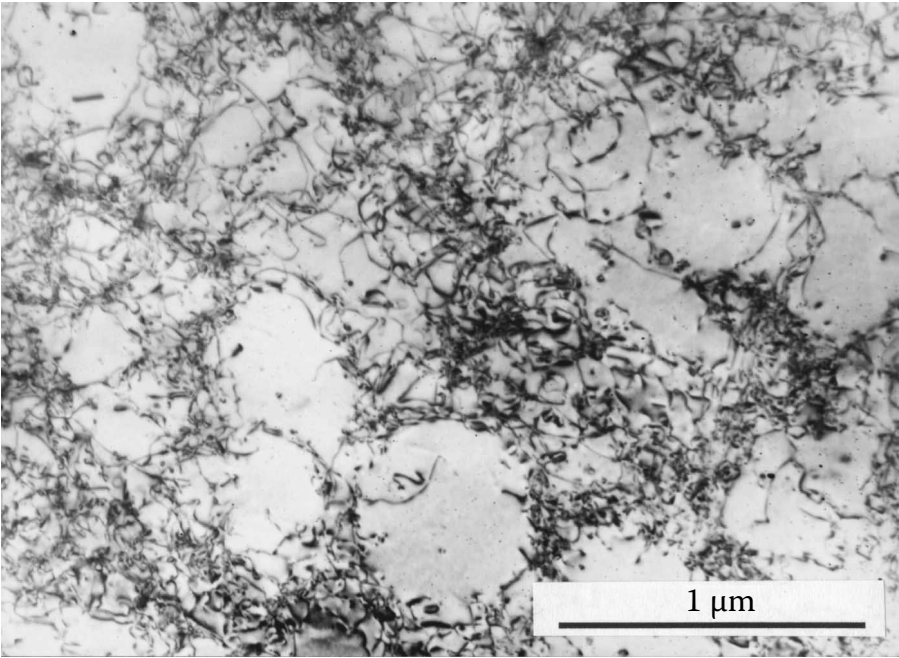


FIGURE 2.5

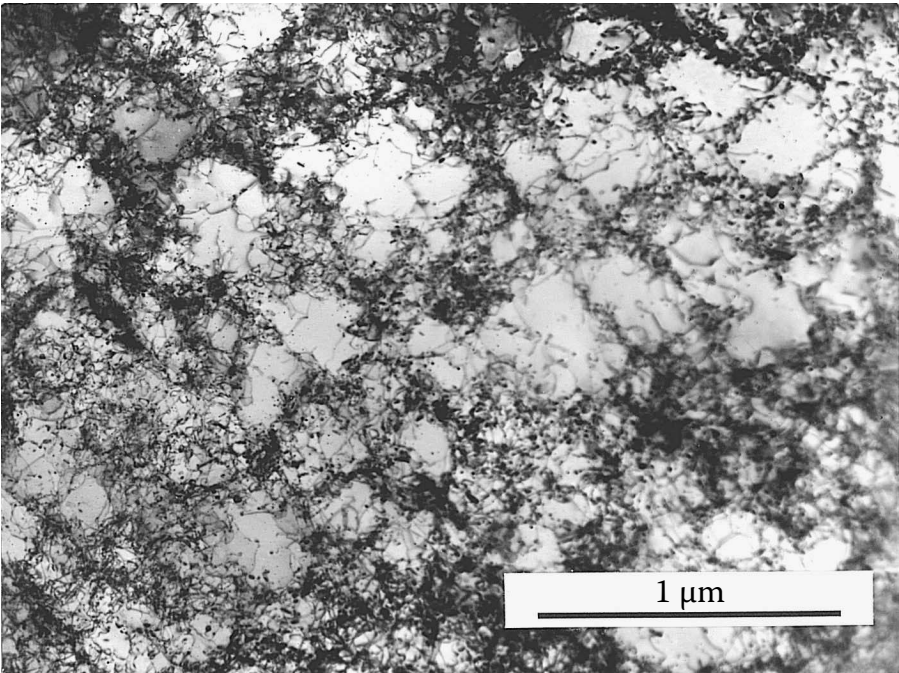


FIGURE 2.6

FIGURE 2.5 With increasing amount of deformation, the number of dislocations in cell walls increases and the size of the cells decreases, but the interior of the cells remains free of dislocations.

Aluminum after 30% cold deformation.

FIGURE 2.6 After large amounts of deformation (over 50% reduction), the refinement of cells is accompanied by their elongation parallel to the deformation direction.

Aluminum after 50% cold deformation.

2.2 FORMATION OF A DISLOCATION SUBSTRUCTURE AT HIGHER TEMPERATURE IN METALS OF HIGH STACKING FAULT ENERGY

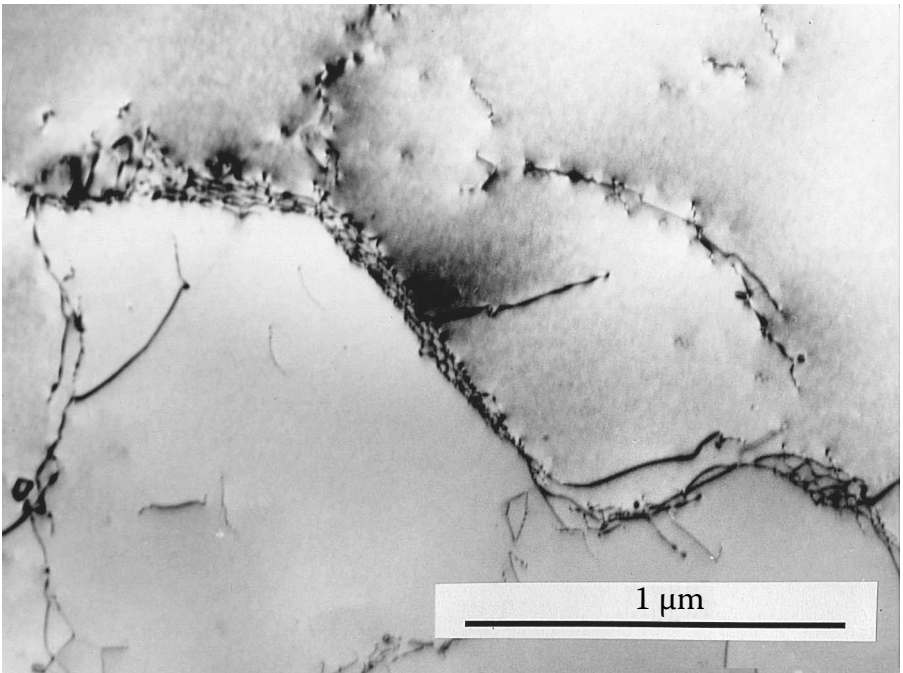


FIGURE 2.7 When the plastic deformation is carried out at higher temperatures, the cross-slip of dislocations is assisted by diffusion processes. The annihilation of dislocations of opposite sign is thus much easier, and this reduces the dislocation density. Because the number of dislocations building the dislocation tangles is smaller, the tangles are narrower but more regular than at lower temperatures.

Aluminum after 2% deformation at 400°C.

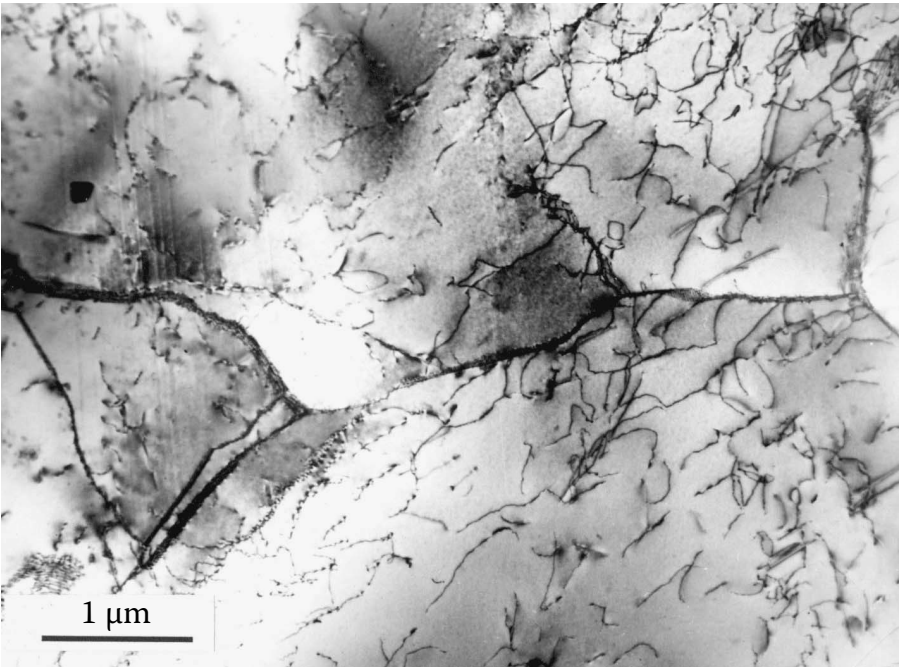


FIGURE 2.8

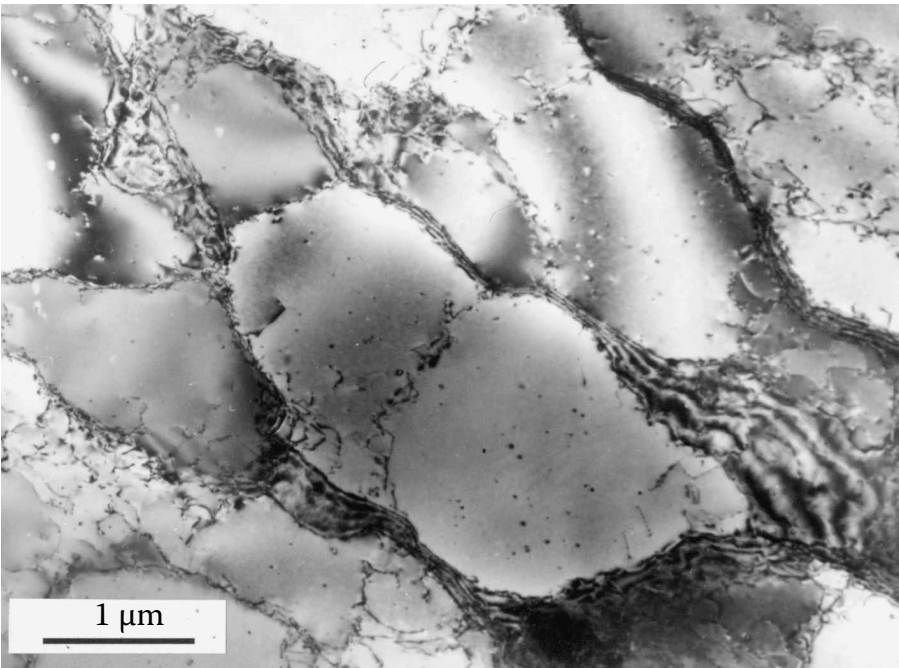


FIGURE 2.9

FIGURE 2.8 Each dislocation arriving at a cell wall contributes to the misorientation of the crystal lattice across the wall by a magnitude depending on the Burgers vector of that dislocation. At higher temperatures, the number of dislocations that succeed in passing through the free cell interior before meeting a wall is greater than at lower temperatures. When the magnitude of the misorientation between adjacent cells produced by these dislocations exceeds 1 to 2 degrees, the dislocation wall transforms into a *dislocation subboundary*.

Aluminum after 5% deformation at 400°C.

FIGURE 2.9 After a higher percentage deformation (more than 10%–15%) at a higher temperature, the three-dimensional dislocation tangles transform into two-dimensional dislocation complexes with a crystallographic misorientation of more than 10 degrees. The dislocation cell structure transforms into a *dislocation substructure* containing *subgrains* bordered by *high-angle subboundaries*. All metals and alloys of high SFE show a tendency to build a dislocation substructure when subjected to hot rolling or extrusion.

Aluminum after 15% deformation at 400°C.

2.3 FORMATION OF A DISLOCATION SUBSTRUCTURE AT ROOM TEMPERATURE IN METALS OF MEDIUM-LOW STACKING FAULT ENERGY

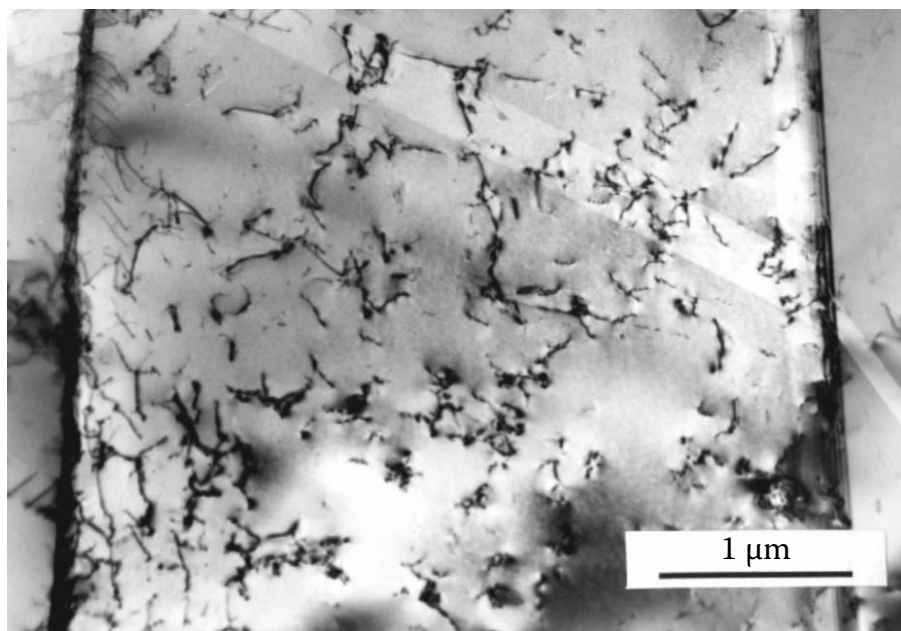


FIGURE 2.10 The splitting of dislocations in metals of low SFE inhibits cross-slip, but in metals of medium-low SFE (about 15–40 mJ/m²), the width of SF separating the partials is relatively small. So the applied external stress can easily bring the partials together and cause them to behave as unit dislocation. This is the case in slightly deformed copper and in austenitic steel Fe-18Cr-9Ni, where the major portion of the dislocations move as unit dislocations and produce three-dimensional dislocation distribution.

Austenitic steel Fe-18Cr-9Ni after 2% cold rolling. There is only slight evidence for an arrangement of dislocations in planar rows.

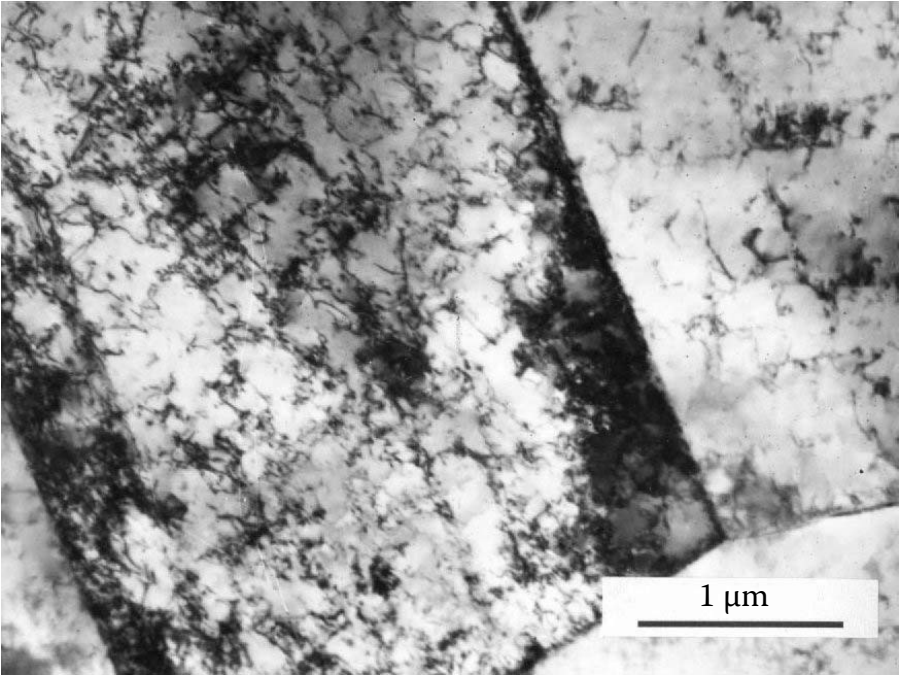


FIGURE 2.11

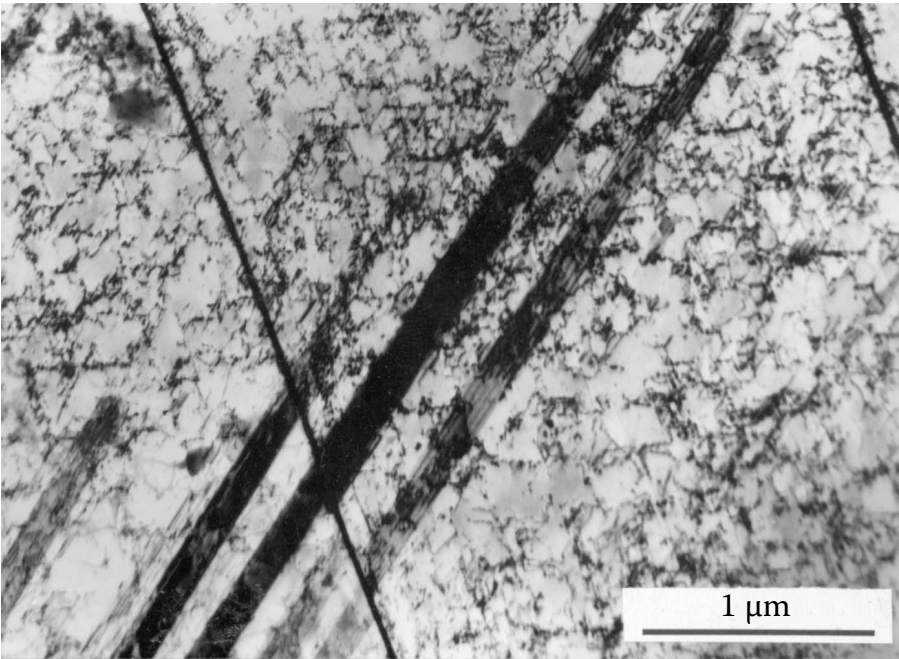


FIGURE 2.12

FIGURE 2.11 At moderate deformations, the dislocations in medium-low-SFE metals form tangles and imperfect dislocation walls but still preserve some tendency for planar arrangement.

Austenitic steel Fe-18Cr-9Ni after 10% cold rolling.

FIGURE 2.12 A frequent deformation mechanism in medium-low-SFE alloys is deformation twinning. In austenite, the twins form thin bands on the $\{111\}$ crystal planes of the fcc lattice. The contrast produced by the deformation twins (the dark bands) is similar to the contrast arising from the bands of SF.

Austenitic steel Fe-18Cr-9Ni after 10% cold rolling.

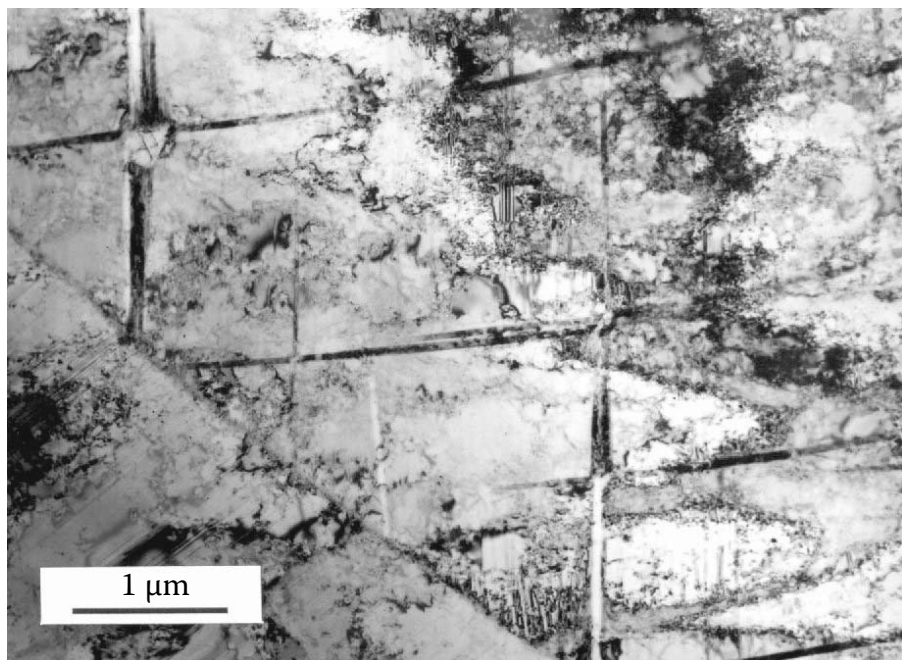


FIGURE 2.13 The substructure of a moderately deformed austenite consists of bands of deformation twins in a matrix containing refined dislocation cells.

Austenitic steel Fe-18Cr-9Ni after 20% cold rolling.

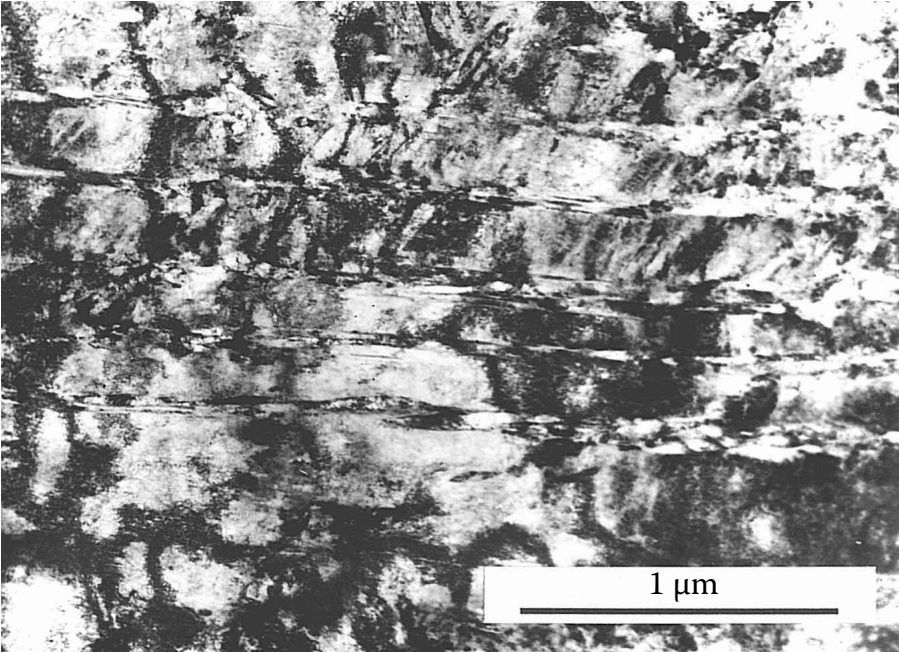


FIGURE 2.14

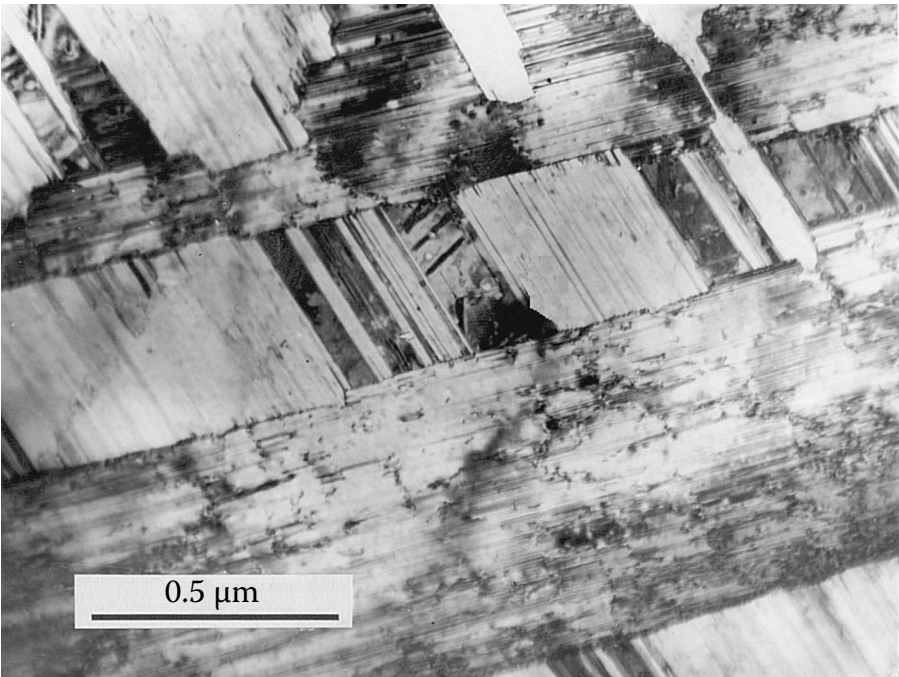


FIGURE 2.15

FIGURE 2.14 Deformation twinning develops only at moderate deformation amounts. The microstructure of severely deformed medium-low-SFE alloys contains extremely fine dislocation cells, in which individual dislocations are barely resolvable, and long straight microbands, which are due to the highly concentrated slip along the traces of the active slip planes. The microbands give rise to the well-known *slip lines*, which can easily be observed in a metallographic microscope.

Austenitic steel Fe-18Cr-9Ni after 50% cold rolling.

FIGURE 2.15 In some low-SFE steels—for example, the Hadfield steel—deformation twinning is the main mechanism of plastic deformation over the whole deformation range.

Fe-0.2C-14Mn steel, tensile loaded to 20% elongation. The structure consists of twins of various widths. Note that the twins stop at other twins boundaries. This behavior demonstrates the main difference between twins and plate (twinned) martensite, the latter always terminating at a boundary (see [Chapter 5.5](#)).

2.4 FORMATION OF A DISLOCATION SUBSTRUCTURE AT ROOM TEMPERATURE IN METALS OF VERY LOW STACKING FAULT ENERGY

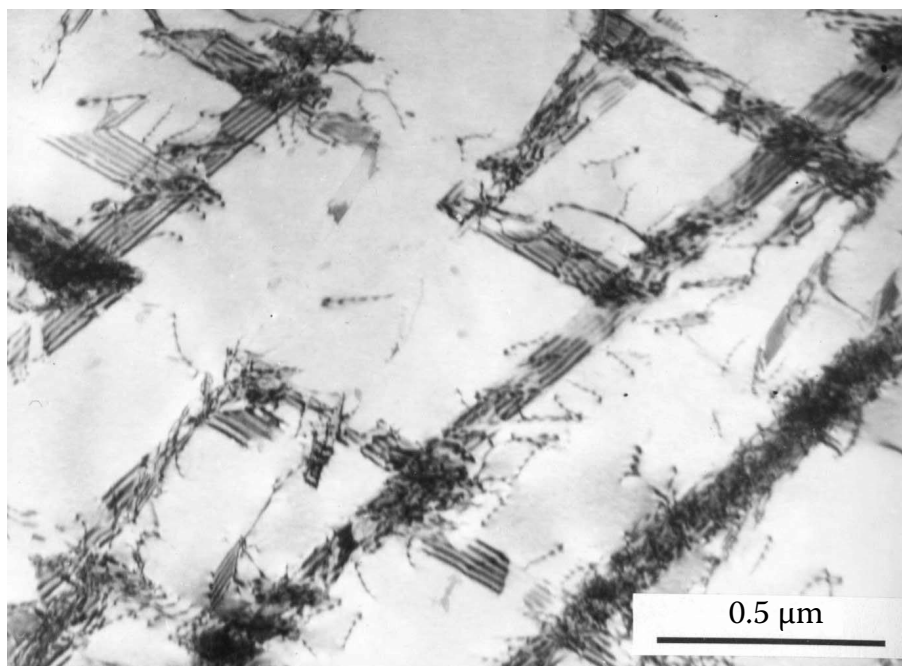


FIGURE 2.16 When the SFE value is very low (less than 10 mJ/m^2), all the dislocations are split into partials separated by wide SF ribbons. The cross-slip of such a complex structure is impossible. It can move only by gliding in its own crystal plane, thus producing a well-defined planar distribution of dislocations. Structures of this type are observed in silver, brass, aluminum bronze, and austenitic steels alloyed with nitrogen.

Austenitic nitrogen steel Fe-18Cr-14Mn-0.6N after 2% cold deformation.

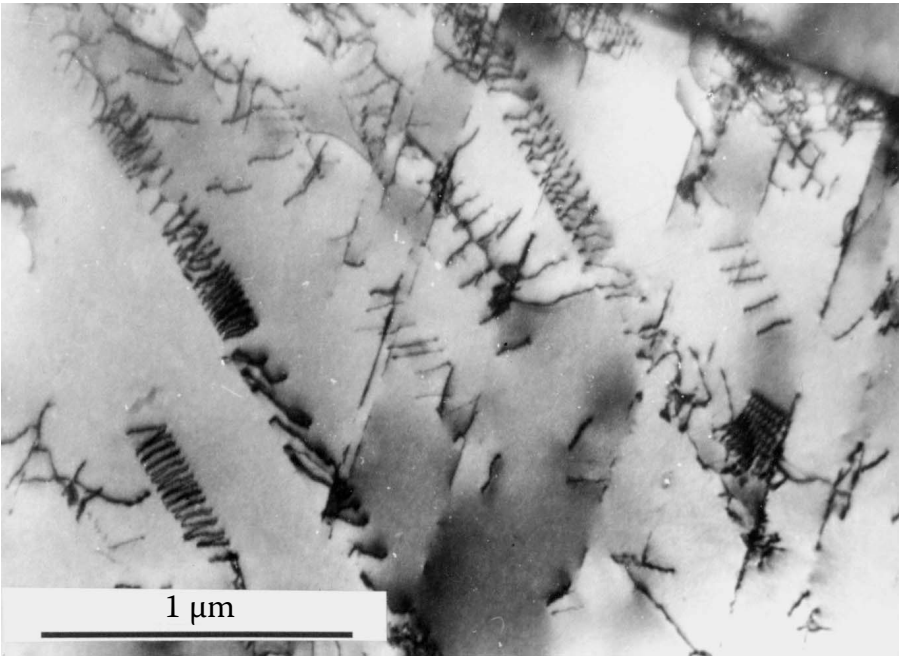


FIGURE 2.17

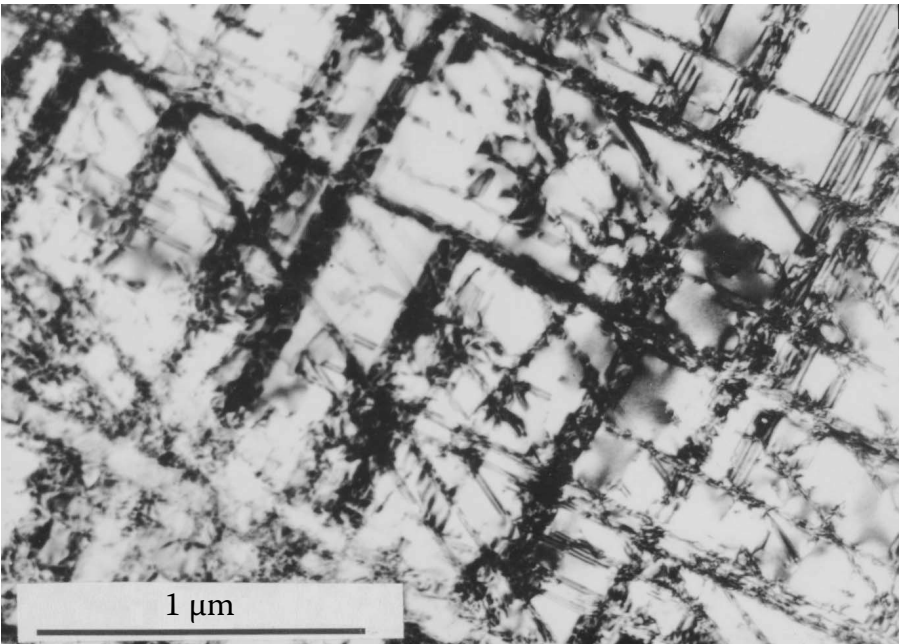


FIGURE 2.18

FIGURE 2.17 Split dislocations in metals of very low SFE usually move in rows, each dislocation following in the track of the preceding one.

Austenitic nitrogen steel Fe-18Cr-14Mn-0.6N after 2% cold deformation. The imaging conditions have allowed only the partial dislocations to be visible, while the SF ribbons are invisible.

FIGURE 2.18 The deformation twinning and the formation of hcp ϵ -martensite are frequent deformation mechanisms in the fcc crystals of very low SFE. The morphology and the crystallography of both these deformation products are very similar to those of stacking faults, and it is very hard to distinguish them even by TEM.

Austenitic nitrogen steel Fe-18Cr-14Mn-0.6N after 10% cold deformation. The dark bands are probably sheets of ϵ -martensite because they do not stop when meeting another band but propagate further, suffering only a slight displacement. The ribbons containing black and white fringes are stacking faults.

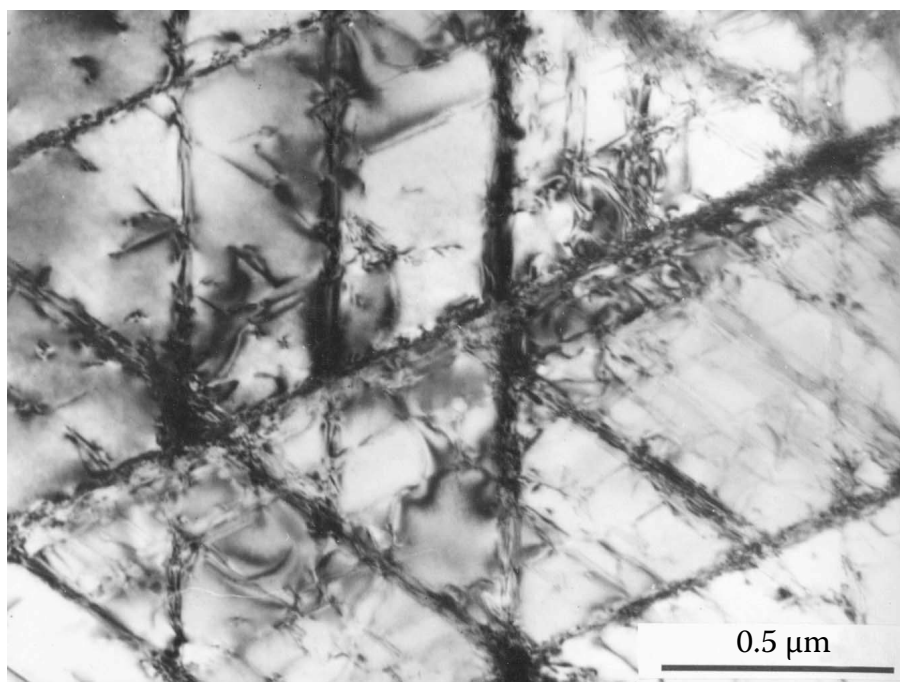


FIGURE 2.19 The probability of the activation of dislocation movement on a particular close-packed plane depends on the orientation of applied stress to that plane. The movement starts initially only on planes that are “active” with respect to the applied stress. Due to the higher dislocation density, the active planes are visualized in the TEM as dark slip lines or slip bands.

Austenitic nitrogen steel Fe-18Cr-14Mn-0.6N after 20% cold deformation. Note the 60° angle between the separate bands on the micrograph; this is the angle between the three active $\{111\}$ planes of the fcc crystal.

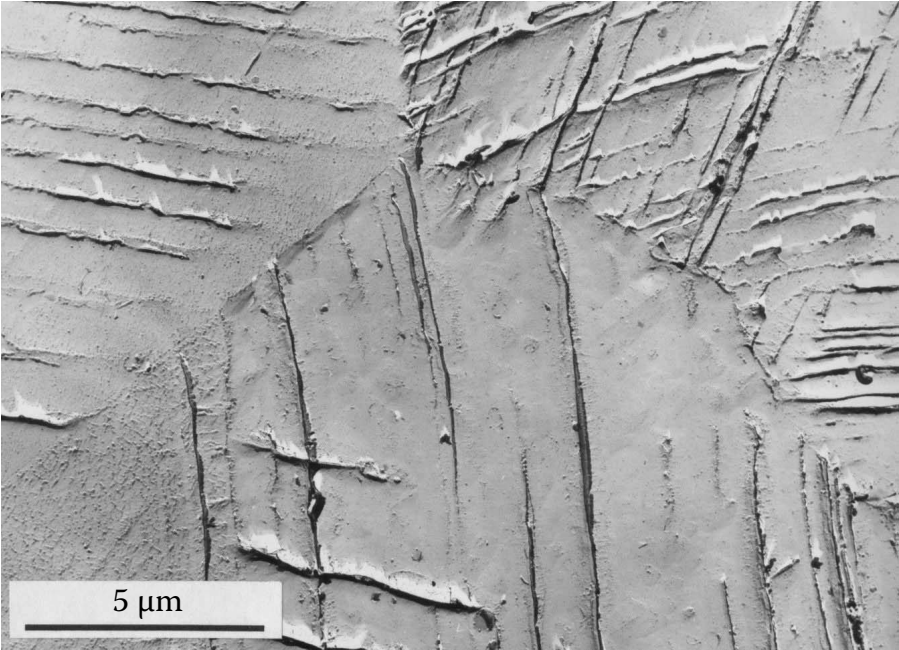


FIGURE 2.20

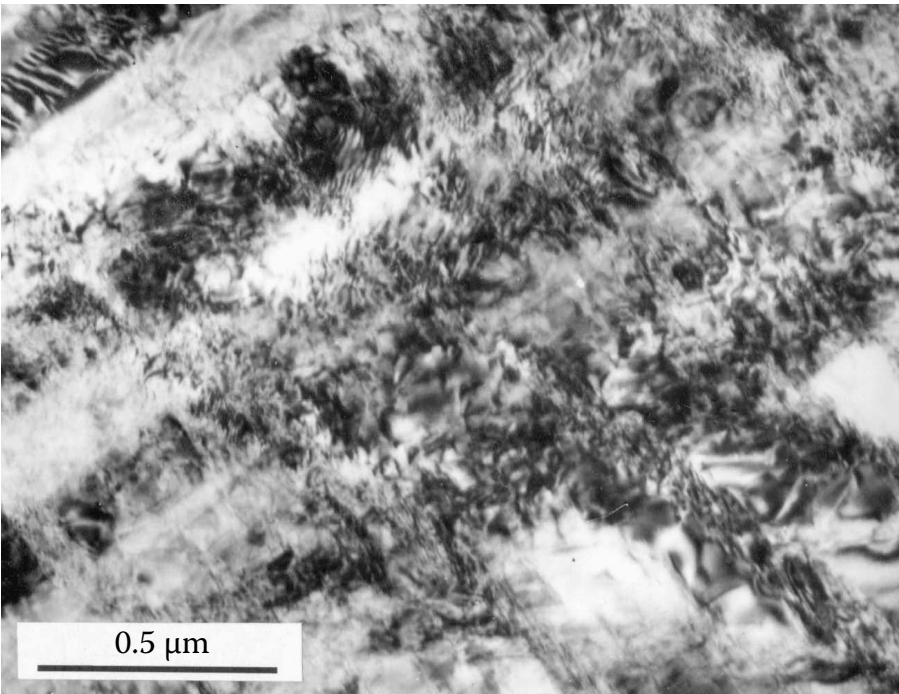


FIGURE 2.21

FIGURE 2.20 The preferred movement of dislocations in separate slip planes results in the formation of more stable *persistent slip bands*. The development of these bands depends on the value of SFE: the lower the SFE, the faster the bands form. The deformation bands are easily detected by light microscopy and by TEM replicas due to the accelerated etching of the bands' intersections with the polished and etched surface of the specimen.

Austenitic nitrogen steel Fe-18Cr-14Mn-0.6N after 20% cold deformation. Replica. The direction of the bands is related to the crystallographic orientation of the active *slip systems* in each specific grain. The bands in adjacent grains have a different direction due to the different crystallographic orientation of the grains.

FIGURE 2.21 The planar dislocation distribution in low-SFE alloys remains unchanged up to a reduction of several tens of percents. With the increase of deformation, the newly introduced dislocations not only join the existing rows but also form new planar rows.

Austenitic nitrogen steel Fe-18Cr-14Mn-0.6N after 20% cold deformation. The number of dislocations in the shown planar rows is so high that it is difficult to resolve the individual dislocations.

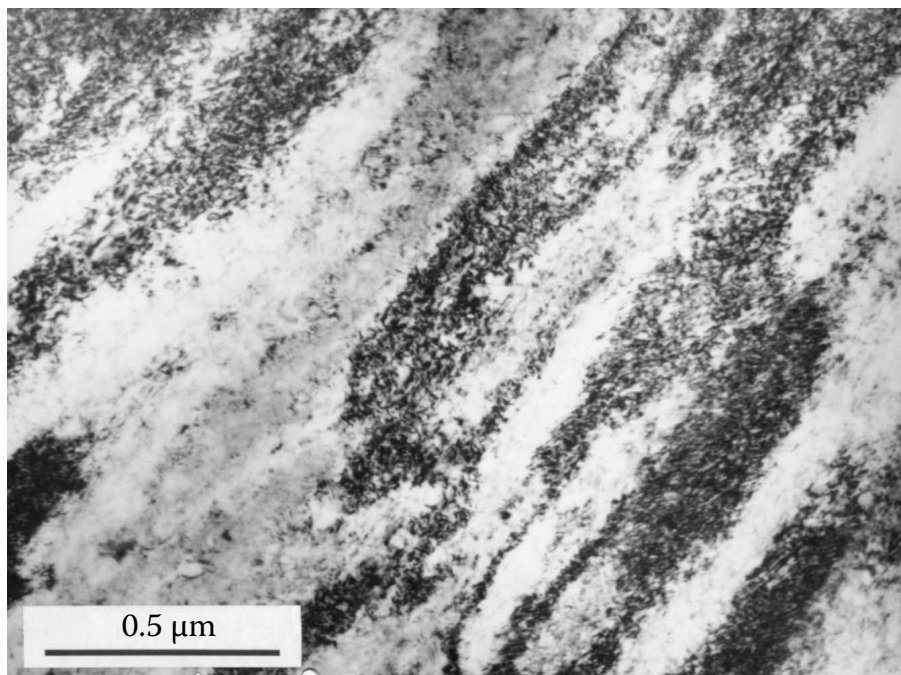


FIGURE 2.22 Severe cold deformation brings the partial dislocations closer together and forces them to move on all close-packed planes. However, if the SFE value is extremely low, the wide SF ribbons cannot be fully constricted even at deformations above 50%. Thus, severely deformed alloys of extremely low SFE develop a special grid structure—a fine three-dimensional grid of partial dislocations separated by extremely narrow stacking faults.

Austenitic nitrogen steel Fe-18Cr-14Mn-0.6N after 50% cold deformation.

2.5 FORMATION OF A DISLOCATION SUBSTRUCTURE AT HIGHER TEMPERATURE IN METALS OF LOW STACKING FAULT ENERGY

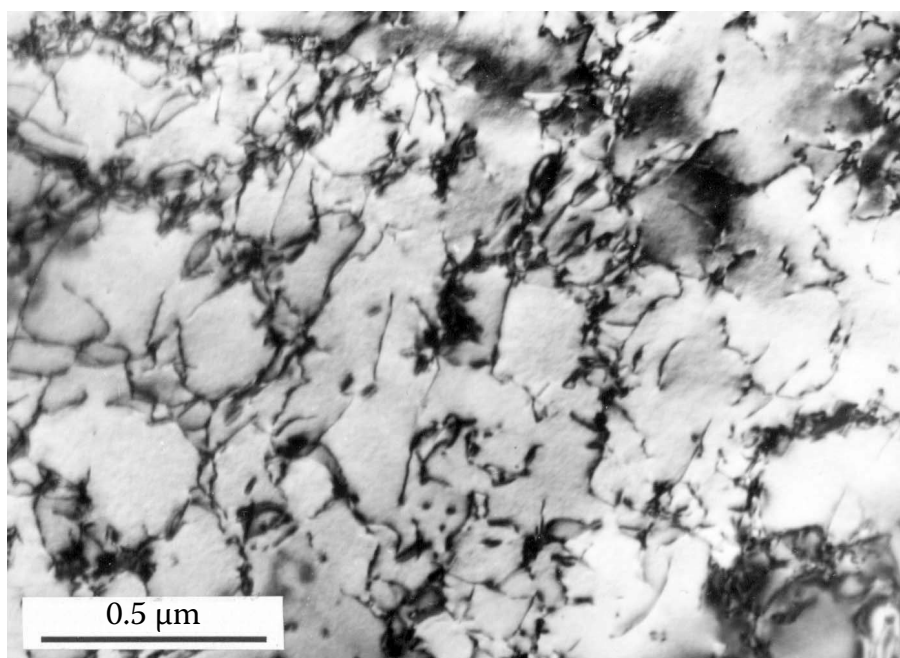


FIGURE 2.23 The SFE of a metal increases with temperature. The plastic deformation at elevated temperatures ($T < 0.5T_{\text{melt}}$ —not higher than the temperature of recrystallization) is known in industrial practice as *warm plastic deformation*. The “warm”-deformed medium-low-SFE alloys (copper, austenitic steel Fe-18Cr-9Ni) develop a cell structure similar to that of cold-deformed high-SFE alloys but form a larger number of imperfect cells.

Austenitic steel Fe-18Cr-9Ni after 5% deformation at 400°C.

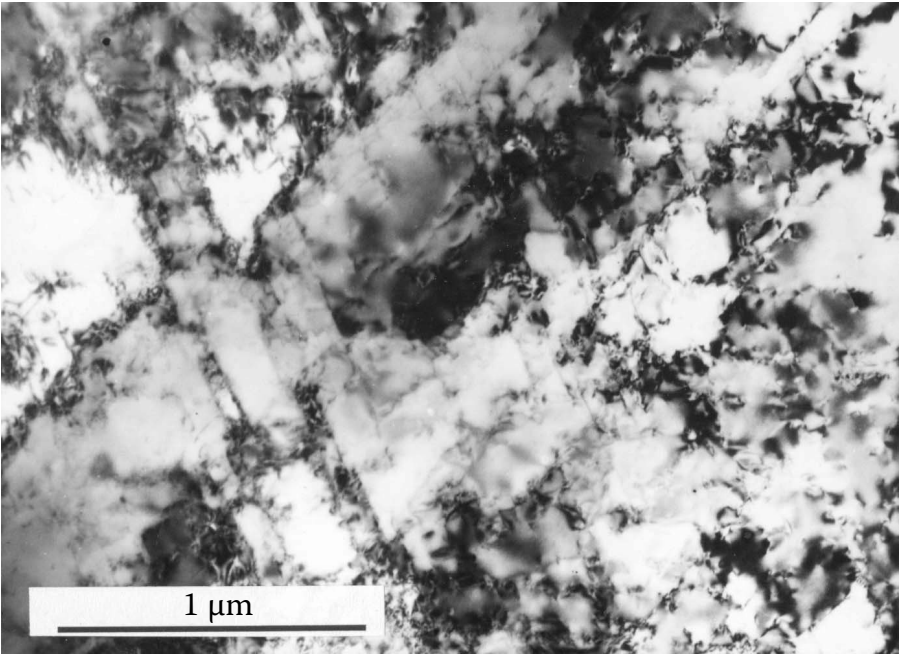


FIGURE 2.24

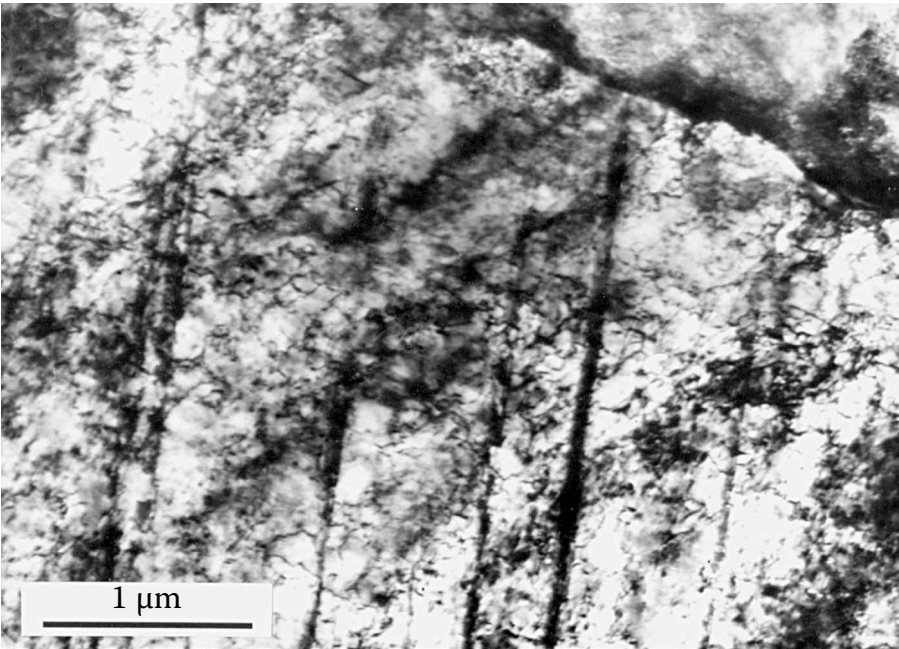


FIGURE 2.25

FIGURE 2.24 The increase of SFE in alloys of very low initial SFE (e.g., silver, brass, aluminum bronze, nitrogen-alloyed austenitic steels) at the temperature of warm deformation is not sufficient for the recombination of partial dislocations into unit dislocations. That is why some elements of planar dislocation structure are preserved in these alloys after a small amount of warm deformation.

Austenitic nitrogen steel Fe-18Cr-14Mn-0,6N after 10% deformation at 400°C.

FIGURE 2.25 After a larger amount of warm deformation, the majority of the partial dislocations in low-SFE alloys are brought together to form unit dislocations, which then develop a three-dimensional dislocation distribution. However, the presence of planar bands in the structure indicates that some dislocations still have preserved their separation and have thus continued to move as partials.

Austenitic nitrogen steel Fe-18Cr-14Mn-0,6N after 30% deformation at 400°C.

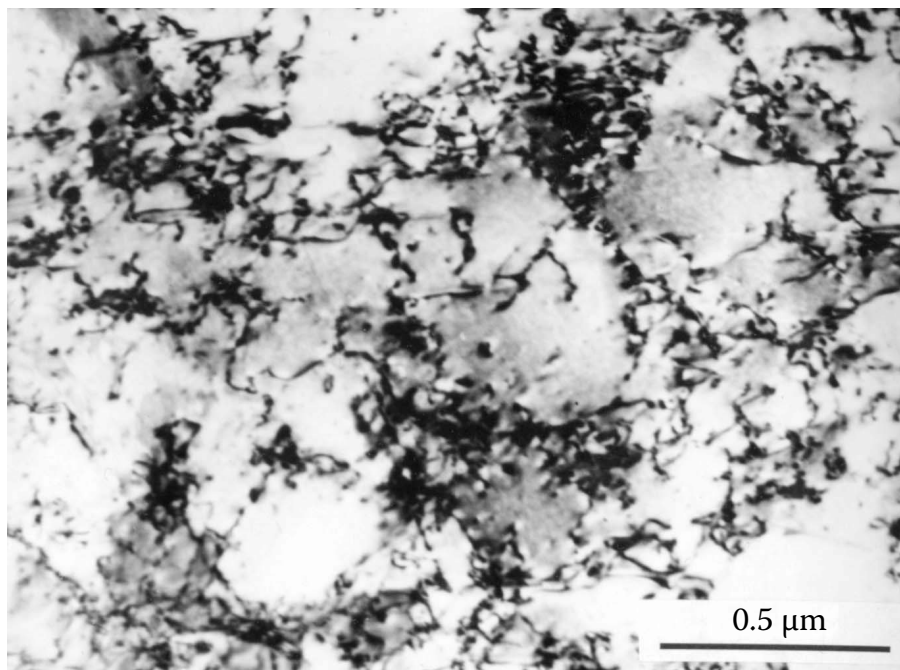


FIGURE 2.26 The increase of SFE at the temperatures of *hot deformation* is sufficient to make all dislocations become unit ones. These dislocations can move by climb and cross-slip in all the slip planes to form tangles and cells. The cells are similar to the cells of cold-deformed high-SFE alloys except for wider cell walls, the smaller misorientation between cells, and the larger number of dislocations in the cell interior in the hot-deformed low-SFE material.

Austenitic nitrogen steel Fe-18Cr-14Mn-0,6N after 5% deformation at 900°C.

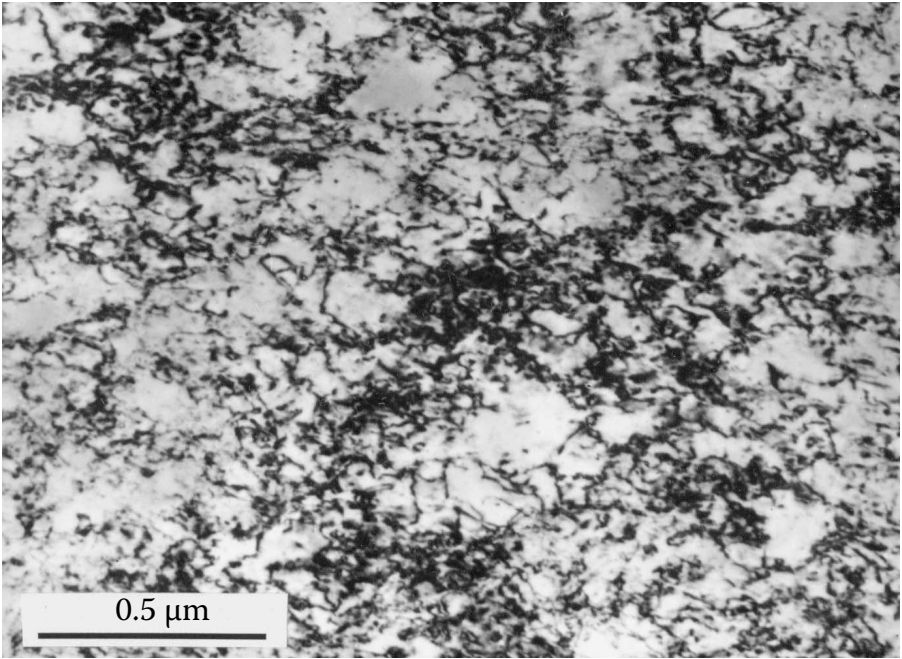


FIGURE 2.27 Even after a large amount of hot deformation, the width of cell walls in low-SFE alloys remains large and the misorientation between cells is small. In this group of alloys, the dislocations never build the substructure typical of the hot-deformed high-SFE alloys (see Figures 2.7–2.9).

Austenitic nitrogen steel Fe-18Cr-14Mn-0,6N after 30% deformation at 900°C.