DISLOCATION SOURCES IN ORDERED INTERMETALLICS*

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Abstract

An overview on the current understanding of dislocation sources and multiplication mechanisms is made for ordered intermetallic alloys of the L1₂, B2, and D0₁₉ structures. In L1₂ alloys, a large disparity of edge/screw segments in their relative mobility reduces the efficiency of a Frank-Read type multiplication mechanism. In Fe-40%Al of the B2 structure, a variety of dislocation sources are available for <111> slip, including ones resulting from condensation of thermal vacancies. In NiAl with the relatively high APB energy, <100> dislocations may result from the dislocation decomposition reactions, the prismatic punching out from inclusion particles, and/or steps and coated layers of the surface. Internal interfaces often provide sites for dislocation multiplication, e.g., grain boundaries, sub-boundaries in Ni₃Ga, NiAl and Ti₃Al, and antiphase domain boundaries in Ti₃Al. As for the crack tip as a dislocation source, extended SISFs trailed by super-Shockley partials emanating from the cracks in Ni₃Al and Co₅Ti are discussed in view of a possible toughening mechanism.

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Introduction

In the analysis of dislocation mechanisms that are responsible for mechanical behavior of ordered intermetallic alloys, it is often emphasized that mobility of superdislocations in an ordered superlattice crystal structure is reduced considerably, as compared to that of dislocations in the disordered crystal structure, because of the dissociation configurations of superdislocations and/or the core structures of individual superpartials. While such mobility-limited dislocation activities may play an important role in some aspects of the mechanical behavior of intermetallic alloys, such as the physical origin of yield strength anomaly, heterogeneous sources for superdislocations and sites for their multiplication mechanisms are also important for comprehensive understanding of plastic deformation behavior and brittle fracture characteristics. The purpose of this paper is to summarize the current understanding of dislocation sources and multiplication mechanisms in ordered intermetallic alloys by extending the excellent review, "Dislocation Sources," given earlier by Li [1].

As compared with metals and disordered alloys, Frank-Read (F-R) sources are generally very difficult to operate in ordered intermetallic alloys because of the geometrical requirement for switching between the leading and trailing superpartials at each revolution [2]. Since the superpartials are, individually, more flexible than the parent superdislocation, it is easier for an alloy with low antiphase boundary (APB) energy to generate single superpartials than to generate pairs. Therefore, the variation of flow stress with respect to the degree of order in Fe-Co 2 at.% V of the B2 structure [3] and Cu3Au of the L12 structure [4] was rationalized in terms of the efficiency of F-R sources. In the case of those intermetallic alloys with the relatively high ordering energies, such as most of transition-metal aluminiides and silicides with the high ordering-transition temperatures, ordering and disordering treatments are not applicable. In this case, dislocation sources and multiplication mechanisms need be discussed by making direct comparisons with other intermetallic alloys of the same ordered crystal structures and/or those metals and alloys of the disordered crystal structures. The outline of the paper is as follows: in the next section, a brief overview is given on the dislocation sources and sinks and the multiplication and annihilation mechanisms mostly on the basis of the review articles [1,5] on metals and disordered alloys. In the subsequent sections, the available experimental results on ordered intermetallic alloys are summarized according to the crystal structure types belonging to fcc-based, bcc-based, and hcp-based superlattice structures. Discussion will include specific roles of dislocation sources and multiplication mechanisms in the mechanical response of ordered intermetallics under the stress-controlled and strain-controlled deformation experiments.

Background

Consider an early stage of plastic deformation (say, the total strain is less than a few percent, \(\varepsilon < 10^{-2}\)) and an ideal dislocation arrangement that causes the plastic deformation as depicted schematically in Fig. 1. It is assumed that only dislocation sources for the primary slip system are activated to give a mobile dislocation density, \(\rho\), and the grown-in dislocations are to remain immobile. Then, the shear strain contribution from an individual crystallite may be given as \(\varepsilon = bpx\), where \(b\) is the magnitude of Burgers vector and \(x\) is the slip distance. The shear strain rate may be given as

\[
\dot{\varepsilon} = b \frac{\partial}{\partial t} (px) = b \dot{\rho} x + b \rho v
\]

which becomes the Orowan's relationship when the mobile dislocation density is constant. In other words, the multiplication rate and the annihilation (or immobilization) rate of glide dislocations are balanced to give \(\dot{\rho} = 0\). As will be discussed later, such a steady state condition seldom occurs in real crystalline materials, especially in view of the plastic deformation in microscale.
Figure 1 - A schematic drawing of a source and sinks of mobile dislocations with a mean free path of $\Lambda$ under the applied resolved shear stress, $\tau$.

In "perfect" crystals with an initial dislocation density of $\rho_0 < 10^6 \text{ cm}^{-2}$ and no sub-boundaries, the multiplication starts from surface sources and the dislocations traverse paths of the order of the crystal diameter [5]. The critical resolved shear stress (CRSS) is smaller than the F-R stress calculated for dislocation segments between grown-in dislocations, since surface sources can be activated more easily [1,5]. In imperfect crystals with $\rho_0 > 10^6 \text{ cm}^{-2}$, on the other hand, dislocation multiplication starts first in sub-boundaries of minimum misorientation [5]. The CRSS of these crystals is of the order of the F-R stress calculated for these sub-boundary sources, and the glide paths are of the order of a few times of the subgrain diameters. In polycrystals each grain behaves similarly to an imperfect crystal.

Regardless of the types of F-R source, such as double or single-ended sources, double cross-slip sources, superjog sources, or grain boundary sources, the critical condition for source activation and dislocation multiplication depends sensitively on the difference in line energies between screw and edge components, anisotropic elasticity and self-interaction of the various segments of the dislocation line as well as core energy contribution. Those papers that made significant contributions in improved treatments of the original F-R model are listed in the review paper by Neuhauser [5]. Because of the higher line energy of the edge component than the screw component, the bowing-out segments resemble half ellipses with the long axis in the direction of Burgers vector. Moreover, when there exist short-range barriers specifically associated with the motion of screw dislocations, the configuration of dislocation loops emanating from a F-R source is much more strongly aligned along the Burgers vector, containing long screw segments. The two cases in point are the $<111>\{110\}$ dislocations in bcc metals and alloys and the $<101>\{111\}$ superdislocations in many L12 alloys. The schematic drawing of Fig. 2(a) illustrates the latter case [6], where either the cross-slip pinning (CSP) of the outer superpartials [7] (the top-half) or the immobilization by Kear-Wilsdorf (K-W) lock [8] (the bottom-half) causes the elongated shape of a glide loop containing a distribution of macro-kinks or the so-called super-kinks [9-13]. This will be discussed in the next section.

Since the F-R mechanism requires a dislocation segment to start with, when a "bulk source" is located within a crystallite (Fig. 1), it is really a multiplication mechanism rather than a nucleation mechanism [1]. The Bardeen-Herring mechanism, the operation of the F-R source by climb instead of glide, is also a multiplication mechanism. On the other hand, the formation of dislocation loops from supersaturated solutions of point defects, a "climb source," is a nucleation mechanism. Also considered as nucleation sites are particles and inclusions of extrinsic type as well as free surfaces and crack tips. It is ambiguous, however, to call sub-boundaries and grain boundaries as nucleation sites because these invariably contain interfacial dislocations, whose characters are closely related to the lattice dislocations of the adjacent grains.
The inhomogeneous and discontinuous nature of slip and twinning deformation by the collective motion of dislocations are well recognized, but only a few attempts have been made to analyze the dynamic operation of these sources that take into account the shapes of the source segments and the dynamic features of propagation of the loops generated around the source [5]. The difficulty here is in the rigorous treatment of bowed-out dislocation segment. Direct observation of source operation in a relatively thick foil by in-situ straining high-voltage electron microscopy (HVEM) is most desirable for an assessment of the critical condition for dislocation multiplication. For intermetallic alloys, there have been only a few reports quoted in the review papers [1,5].

**L1_2 Alloys**

**Frank-Read Sources and Sub-boundaries**

Inhomogeneous and discontinuous \{111\} slip deformation in \(\text{Ni}_3\text{Ga}\) single crystals was well characterized by Takeuchi, Suzuki, Ichihara [14] using in-situ electron and optical microscopic observations. At high temperatures (>600 K), two types of multiplication process of superdislocations (dislocations, hereon) for \{100\} slip were observed: one is that from isolated single-ended F-R sources and the other is that from sub-boundaries. Dislocation movement on a \{100\} plane was steady and continuous, indicating viscous glide controlled by the Peierls mechanism. In contrast to the \{100\} slip, dislocation motion on a \{111\} plane at lower temperatures (≤500 K) was so quick and jerky that a direct recording of the dislocation motion was not possible. It was estimated that each \{111\} slip band was formed within a time interval of 0.03 s. Once dislocations were multiplied on the \{111\} plane, they seldom moved again on the same plane when they were oriented along the \langle 110 \rangle screw direction. These features are essentially the same as those observed in \(\text{Ni}_3\text{Al}\) single crystals by Nemoto et al. [15]. Based on the electron microscope observation of foils parallel to the primary \{111\} slip plane of \(\text{Ni}_3\text{Ga}\) compressed at 373 K, Takeuchi et al. [14] identified sub-boundaries and precipitates as dislocation multiplication sites. Their observation of sub-boundaries during in-situ deformation is schematically illustrated in Fig. 3.
On the basis of the *in-situ* HVEM observations [14,15], it may be concluded that the macroscopic yield stress in Ni$_3$Ga and Ni$_3$Al is determined by the activation of slip band sources rather than the mobility of multiplied dislocations. In order for the multiplication of dislocations necessary for the slip band formation to occur, both edge and screw dislocations must glide a considerable distance, except for the special case of surface sources. Once those dislocations of near-edge orientation overcome their obstacles, such as forest dislocations or jogs, they travel a relatively large distance until they encounter the next effective barrier (see Fig. 3). On the other hand, those of near-screw orientation gliding on the $\{111\}$ plane tend to be self-trapped due to the CSP and K-W mechanisms (see Fig. 2(a)). Since the cross-slip is a thermally activated process, the degree of self-trapping increases as temperature increases. Between the two adjacent long screw segments, secondary dislocation sources may be activated from super-kinks [9,16], as illustrated by Fig. 2(b). Each type of these obstacles to the dislocation motion requires an incremental local stress, which contributes to the back stress exerted on the dislocation source. When the back stress reaches a certain level sufficiently high enough to stop effectively the dislocation source, this condition is in accord with the exhaustion hardening mechanism proposed by Thornton, Davies, and Johnson [17].

The relationships between the yield stress at a macro-strain range (e.g., 0.2% offset method) and the dislocation microstructure described above are generally consistent with other *in-situ* deformation and post-mortem transmission electron microscopy (TEM) data in the literature (e.g., see the recent review papers [18-21]). A dynamic simulation study of the glide dislocation loops originating from a F-R source in L1$_2$ alloys was performed by Mills and Chrzan [22], who implemented in their model the non-random distribution of pinning points along the screw dislocations and the expansion of the screw dislocations by means of the lateral motion of super-kinks. They concluded that the principal effect of CSP is not the reduction in the dislocation velocity, but rather the exhaustion of the density of mobile dislocations. In other words, with reference to Eq. (1), the strain rate is controlled by how many dislocations are mobile, rather than how fast they may be moving. They also noted from the simulation analyses that the collective behavior of dislocations undergo a pinning-depinning transition at a critical stress [23]. More recently, Ezz and Hirsch [24] proposed a model for the temperature and strain-rate dependences of the flow stress in L1$_2$ alloys, including the yield stress reversibility, in which the critical step is the operation of F-R sources. In the dynamics of dislocation organization, Shi, Saada, and Veyssière [25] considered the multiplication rate of the kinks and the depletion rate of the mobile dislocations, where the former is controlled by a mean kink height and the latter is related to a mean free path determined by the neighboring microstructure.
Grain Boundaries

Grain boundaries in L1₂ alloys are often observed by TEM not only as barriers to (or annihilation sites for) dislocations, but also as sites for dislocation multiplication. Some typical examples were given in the in-situ TEM study of slip transfer at grain boundaries in Ni₃Al [26] and the post-mortem TEM observation of dislocation-grain boundary interactions in Al₆₆Ti₂₅Mn₀ [27]. In the latter case of Al₅Ti-based ternary alloy of the L1₂ structure, the activity of dislocation transfer at grain boundaries was observed to increase with increasing temperature. Using the coincident-site-lattice (CSL) theory, King and Yoo [28] showed that many of the possible dislocation reactions at grain boundaries become energetically unfavorable if chemical coordination must be maintained in the structure of the grain boundary. As for the dislocation absorption or initiation at grain boundaries of Σ=3N type (Σ is the volume ratio of the coincident unit to the crystal unit, and N is an integer), the number of allowed dislocation reactions is increased four-fold from 3 to 12, out of the total of 30 possible reactions, when the requirement of L₁₂ ordering is removed (see Table 1). The total of 30 Burgers vectors consist of <110>/2 type superpartials, <112>/6 type Shockley partials, and <112>/3 type super-Shockley partial dislocations. At a grain boundary of Σ≠3N type (see Table 1), no absorption (or initiation) reaction is permitted when the grain boundary region is to be fully ordered. This result suggests that localized compositional disordering (hypostoichiometric, or Ni-rich Ni₃Al) at the vicinity of grain boundaries may facilitate dislocation emission from high-angle grain boundaries, thus enhancing the generalized plasticity of a polycrystalline alloy. Although the concept of localized disordering at grain boundaries has been supported by various theoretical modeling analyses [29], there appears to be no firm experimental evidence for the localized disordering [29,30]. There have been a number of TEM observations for the emission of partial dislocations from grain boundaries, however. For instance, in Ni₃Al, emission of single super-Shockley <112>/3 dislocations from a grain boundary trailing strips of superlattice intrinsic stacking fault (SISF) is often observed [31], and nucleation of a microtwin from a grain boundary results from the emission of super-Shockley partials on the successive {111} layers [32].

Table I. Numbers of Allowed Dislocation Reactions in L₁₂ Alloys

<table>
<thead>
<tr>
<th>Absorption</th>
<th>Transmission</th>
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<tr>
<td>Σ = 3N</td>
<td>Σ ≠ 3N</td>
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<tr>
<td>Σ = 3N</td>
<td>Σ ≠ 3N</td>
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| Disordered (fcc) | 12 | 6 | 288 | 216 |
| Ordered (pc)    | 3  | 0 | 72  | 72  |

[Table 1 continued]

Crack Tips

Pertaining to crack tip as a dislocation source, Rice [33] made a hypothesis that the energy release rate for dislocation emission scales with the so-called “unstable stacking energy,” γᵤₛ, which is the saddle-point maximum associated with a rigid body shear displacement of one part of crystal over another. For the case of an edge dislocation (φ = 0°) on the slip plane that intersects the Mode I crack tip at an angle, θ, as shown in Fig. 4(a), a criterion for ductile vs. brittle behavior is given by [33,34]

$$q = \frac{\gamma_{us}}{2\gamma_s f_1(\theta)}$$  \hspace{1cm} (2)
where \( \gamma \) is the cleavage surface energy and \( f_1(\theta) = \cos^2(\theta/2) \sin(\theta/2) \) is for the case of elastic isotropy. The lower the value of \( q \) is, the more likely the emission of dislocations from the crack tip, instead of crack extension, would be. Using the calculated values of cleavage energies, elastic anisotropies, and \( \gamma_w \) estimated based on the complex stacking fault (CSF) energies [35], it is predicted from Eq. (2) that Ni$_3$Si is intrinsically more brittle than Ni$_3$Al. For instance, for a (001)[110] crack, \( q \) for Ni$_3$Si is estimated to be larger than that for Ni$_3$Al by a factor of two.

Extending the original analysis of Rice and Thomson [36], Bartholomeusz and Wert [37] investigated numerically the effect of APB or SISF energy on the so-called fractional angular range \( \zeta \) for spontaneous or thermally activated emission of a dissociated \{111\}<110> dislocation from a crack tip. When a transgranular microcrack impinges on a grain boundary, in the case of Ni$_3$Al with \{111\} APB energy of 175 mJ/m$^2$ [35], it is more likely to cause dislocation emission into the neighboring grain (i.e., a high \( \zeta \)) rather than propagating transgranularly or intergranularly. This is supported by the massive localized deformation at and near a grain boundary observed during an in-situ TEM experiment using stoichiometric Ni$_3$Al [35].

A group of extended SISF strips are often observed to form in the wake of an intermittently advancing crack, e.g., in Ni$_3$Al [35,38] and Co$_3$Ti [39]. When the trailing super-Shockley partial has a mixed character (\( \phi = \pm 60^\circ \)), as depicted in Fig. 4(b), the driving force on it is influenced strongly by Mode III loading [35]

\[
F = \frac{1}{2} \left( 0.41 K_I \pm 0.87 K_{III} \sqrt{3} \right) - \frac{b}{\sqrt{2 \pi r}},
\]

where \( b \) is the magnitude of \( [112]/3 \), \( r \) is the radial distance from the crack tip, and \( K_I \) and \( K_{III} \) are the stress intensity factors. For the (111)[101] slip ahead of the (001)[110] crack tip, the force on the trailing super-Shockley partial, \( b_z = (211)/3 \), is directed toward the crack when \( K_{III} > 0.27 K_I \). If the SISF energy is low, e.g., 40 mJ/m$^2$ in Ni$_3$Al [35], only the leading super-Shockley partial, \( b_1 = [112]/3 \), which has edge character (\( \phi = 0^\circ \)), can be emitted from the crack tip. This explains the extended SISF formation directly observed during in-situ TEM of the Ni$_3$Al foil, Fig. 7 of Ref. [35].
B2 Alloys

In the main, B2-type intermetallic alloys may be divided into two groups according to the active slip vectors observed at ambient temperatures: (a) $<111>$ slip (e.g., in $\beta$-CuZn, FeAl, and AgMg) and (b) $<100>$ slip (e.g., NiAl and CoTi). Since dislocation sources and multiplication mechanisms in B2 alloys were reviewed earlier [40], only some pertinent papers that were not included in the review and more recent results are discussed in this section.

Dislocation Reactions

The decomposition process of a superdislocation in the B2 structure [41],

$$[111] \rightarrow [010] + [101]$$

is considered more likely to occur in an alloy with a high (101) APB energy (e.g., 810 mJ/m$^2$ of NiAl [42]) because the extent of a spontaneous dissociation of $[111]$ into two $[111]/2$ superpartials in this case is irresolvably small [43]. Conversely, an alloy with a relatively low APB energy should exhibit a dissociated configuration rather than the decomposition. The TEM analysis by Yoshimi et al. [44] showed, however, ample evidence of the decomposition configuration in deformed Fe-39%Al whose APB energy may be less than 300 mJ/m$^2$ of FeAl [42]. Moreover, even in $\beta$-CuZn with the (101) APB energy of only 50 mJ/m$^2$ [45], the decomposition of a $[111]$ edge dislocation was observed after deformation at 190°C [46]. When a NiAl single crystal is oriented along $<100>$ direction, it deforms by $<111>$ slip at low temperatures, but non-$<111>$ slip occurs at high temperatures [47,48]. On the basis of TEM analyses of both deformed and annealed samples, the slip transition in a Ni-rich NiAl crystal at 773 K was attributed to the thermal instability of $<111>$ dislocations against the decomposition reaction [49]. In the Ni-49.75Al-0.25Fe single crystals compressed along the hard orientation, a large number of decomposition reactions were observed in deformation bands that contain $<111>$ screw dislocations [50].

Frank-Read Sources

The HVEM in-situ deformation experiments on $\beta$-CuZn single crystals were performed by Nohara and Imura [51]. Although active sources of dislocations were not revealed, it was shown by these experiments that at 300 K expanding $<111>$ superdislocations take up nearly circular shape on the (110) slip plane, and at 425 K the typical dislocation shape is an elliptical form on the slip plane, the mobility of the screw component being slightly lower than that of the edge character. The lack of disparity in dislocation mobility between the edge and screw components of $<111>$ superdislocations, unlike the cases of bcc metals and L1$_2$ alloys, has been confirmed also in those B2 alloys with the relatively low APB energies by post-mortem TEM studies (e.g., in Fe-39%Al at 673 K [44]). Another feature that is common among $\beta$-CuZn and Fe-39%Al single crystals is the $<111>$-$<100>$ slip transition at their respective peak temperatures of the observed yield strength anomaly. In these B2 alloys, whether dislocation sources for the $<111>$ slip vectors are exhausted [17] as temperature approaches the peak temperatures, or the decomposition process of $<111>$ is enhanced with increasing temperature is not clear at present.

The $<100>$ dislocations in $\beta$-CuZn are multiplied by the super-jog type single-ended F-R mechanism, which was confirmed by the in-situ deformation HVEM experiment at 500 K [52]. The (011)[010] glide loops expanded very smoothly under the applied stress near the peak stress, always keeping the square shape that is consistent with the line tension instability predicted by Head [53].

Vacancy Condensation and Prismatic Punching

The formation of {001}${<001}$ prismatic loops is often observed in NiAl [48], which results from either a supersaturation of vacancies retained after cooling from high temperatures, or prismatic punching from an inclusion particle due to thermoelastical misfit. There are two distinct forms of these loops, as shown in Fig. 5, (a) square loops having the line orientation along $<100>$ with rounded corners, and (b) square loops having the $<110>$ orientation with
extremely sharp corners. The first type, more commonly observed, corresponds to the equilibrium shape calculated on the basis of elastic anisotropy [54,55]. The second type has segments aligned along elastically unstable directions. It has been proposed [54] that the second type is stabilized by dissociation of the total dislocation into three partials

\[
<001> \rightarrow \frac{1}{4} <1\bar{1}1> + \frac{1}{2} <001> + \frac{1}{4} <1\bar{1}1>
\]  

(5)

on the intersecting \{110\} planes coznal to the \langle001\rangle direction, thus suggesting a source of pencil glide type. Though this dissociation has not been confirmed either by high resolution electron microscopy or atomistic simulation of dislocation core, it is a possible source mechanism for \langle001\rangle\{110\} slip which is experimentally observed as the primary deformation mode in NiAl.

Figure 5 - Types of prismatic dislocation loops commonly observed in NiAl (Ref. [48]).

Dislocation microstructures resulting from quenched-in vacancies in FeAl are much more complex than in NiAl. The principal reason for this difference seems to be the relatively high density of grown-in dislocations in the case of FeAl. In Fe-40\%Al alloys, the resulting dislocation microstructure varies with the quenching and aging temperatures, ranging from perfect loops and square helices to more frequently observed "superstructure dislocation loops" and paired helices connected to an APB [56]. The in-situ annealing HVEM of Fe-42\%Al single crystals [57] revealed that the heterogeneous nucleation and growth of dislocation "bundles" were attributable to a sequence of the dislocation dissociation (\langle100\rangle \rightarrow \langle111\rangle/2) and combination of these partials (\langle111\rangle/2 \rightarrow \langle100\rangle). The bundles spread throughout the sample to produce, at the end of annealing, a high density of dislocations of \langle100\rangle Burgers vector.

Crack Tips

An in-situ straining TEM investigation was conducted on three different Fe-Al alloys (34, 40, and 50 \% Al) [58] in order to understand the effect of composition on fracture behavior. In all three cases, dislocation generation was profuse at the foil edges and at grain boundaries. Dislocation emission from a crack was a common observation, in all compositions, whether it was a pre-existing crack, a growing crack, or a crack advancing across a grain boundary. Though not determined in this experiment [58], Burgers vector of these dislocations is most likely of the \langle111\rangle type, perhaps APB-dissociated by two superpartials.
According to the model of crack-tip dislocation emission in B2 alloys [59], an extension of the earlier work on L12 alloys [37], the dissociated \( \langle 111 \rangle \) dislocations are more readily emitted than undissociated dislocations. Another conclusion of this work [59] is that emission of \( \langle 111 \rangle \) superdislocations from crack tips is more difficult than that of \( \langle 100 \rangle \) dislocations. This implies that changing the slip vector of NiAl from \( \langle 100 \rangle \) to \( \langle 111 \rangle \) for satisfying the compatibility condition has a deleterious effect on material's ability to undergo crack-tip blunting and/or shielding. A sequence of dislocation emission from a crack was observed during in-situ tensile testing of the Ni-Al-Fe alloy (Fig. 12 of Ref. [50]). The dislocations were determined to be of the \( (100)(010) \) slip system, the propagating segments having edge character. This case is entirely consistent with the prediction of the modeling study [59].

**Interfaces and Other Effects**

Recently, Levit, Bul, Hu, and Kaufman [60] have reported tensile elongation up to 28% at room temperature in NiAl single crystals of a soft orientation along the [213] direction. Prestraining at 1273 K also led to an increase in the tensile elongation from an average value of 16 to 26%. While this effect is not completely understood, it was noted that the high temperature prestraining resulted in the development of dislocation-free subgrains separated by low energy sub-boundaries. These boundaries might have acted as sources for mobile dislocations and contributed to more uniform strain throughout the gage section.

In two-phase intermetallic alloys, such as \( \beta \) grains in \( \gamma' \) matrix, dislocations could be nucleated into both phases from the heterophase interfaces. One example is the TEM investigation of dislocation microstructure in two-phase alloys comprising \( \gamma'-\text{Ni}_3(\text{Al},\text{Ti}) \) and \( \beta-\text{Ni}(\text{Al},\text{Ti}) \) after inhomogeneous plastic deformation [61]. Another is the in-situ straining TEM analysis of the multiphase Ni-20Al-30Fe alloy [62], which consists of the \( \beta \) phase and the f.c.c./L12 phase. The latter is involved with localized disordering of the L12 particles by slip band intersection. Using thin thermally deposited oxides, Noebe and Gibala [63] observed a surface film softening effect in NiAl, as in several bcc metals. The proposed mechanism for this effect involves the continuous generation and motion of dislocations from surface film-substrate “interfacial dislocation sources” due to elastic and plastic constraint [63]. An important difference of this case from aforementioned heterophase interfaces is the marked disparity in elastic moduli between NiAl and the oxide film.

Effects of pressurization on the generation of dislocations were investigated in nominally stoichiometric cast and extruded polycrystalline NiAl [64]. While the pressure-induced generation of mobile dislocations at second-phase particles was detected, another potential means of dislocation generation was identified in order to account for the increased density of \( \langle 100 \rangle \) dislocations by two orders of magnitude as compared to the annealed NiAl. Slight compositional grain-to-grain variations were revealed to be sufficient to cause the pressure-induced generation of dislocations.

**D0\textsubscript{19} Alloys**

\( \text{T}_3\text{Al} \)

Three important Burgers vectors in the D0\textsubscript{19} crystal structure are involved in the following decomposition reaction of a superpartial dislocation:

\[
< \mathbf{c} + \mathbf{a}/2 > \leftrightarrow [ \mathbf{c} ] + < \mathbf{a}/2 > .
\]  

(6)

In the polycrystalline samples compressed to a low strain (\( \approx 1\% \)), localized distributions of rectilinear \( \mathbf{c} \) dislocations were observed in pure \( \text{T}_3\text{Al} \) [65] and the Nb-containing \( \text{T}_3\text{Al} \) [66]. Since the TEM analysis of undeformed samples revealed a similar dislocation density as in the deformed samples, Court, Löfvander, Loretto, and Fraser [66] conjectured that the observed \( \mathbf{c} \) dislocations were produced most probably during fabrication of the alloy. It is possible also that some of these \( \mathbf{c} \) dislocations may be the reaction products of Eq. (6) resulting from \( \mathbf{c} + \mathbf{a}/2 \) dislocations of an active pyramidal slip, in an analogous manner as in hcp metals [67]. The post-
mortem TEM analysis of Ti₃Al single crystals deformed at 1173 K by Minonishi [68] gives supporting evidence for the decomposition process.

When a single crystal of the DO₁₉ structure is deformed along the [0001] direction, the Schmid factor for a/2 or c slip is zero, hence activating a pyramidal slip system of the relatively high CRSS. This has been demonstrated in both stoichiometric and off-stoichiometric Ti₃Al single crystals [69,70]. Though source and multiplication mechanism for 2c+a superdislocations are not understood, grown-in sub-boundaries composed of a dislocations of the near edge character and a small amount of 2c+a dislocations could have acted as the dislocation sources in the case of Ti-36.5at.%Al single crystals [70]. Alternatively, Court et al. [66] suggested that c+a/2 dislocations, constituting part of the dislocation microstructure in α-Ti phase, will transform from being unit translation vectors to superpartial dislocations upon ordering to α₂-Ti₃Al phase. This implies that a 2c+a superdislocation is formed in Ti₃Al by a pair of superpartials sequentially emanating from a disordered region.

Legros, Court, and Caillard [71] performed both in situ straining experiment and post-mortem weak-beam TEM analysis in Ti₃Al. The uniformly dissociated a dislocations on a prism plane showed the width of ≈9 nm along curved screw segments of a pair of a/2 superpartials. This is in good agreement with ≈10 nm of the recent weak-beam TEM analysis by Wiezorek, Court, and Humphreys [72] and 7 nm of the theoretical determination based on the calculated elastic constants and APB energies [73]. Two different types of sources for the prism slip were observed at 300K. Single-ended dislocation source for b = [1120]/3 dislocations on the (1100) plane was observed in four sequences within the elapsed time of 55 s [71], and the head of a slip band in the basal plane, consisting of ⟨a⟩ screw dislocations originating from a pure-twist sub-boundary, acted as a source of prism slip dislocations [74].

Summary

Dislocations generated by one of the multiplication mechanisms, discussed in the previous Sections, propagate through a crystal over a distance that depends on the substructure of the crystal. The sink (or source) boundaries, schematically shown in one-dimensional view by Fig. 1, may represent dislocation multipoles, sub-boundaries, APD boundaries, grain boundaries, or two-phase interfaces. Unless interfacial dislocations are directly involved in creating a multiplication site at an interface for dislocations into the matrix crystal, these may be classified as the immobile dislocations that are integral parts of the interface.

In macro-crystals of high perfection or thin foils used for TEM, single and/or a group of mobile dislocation(s) appear to move always in a discontinuous manner, indicating that a certain fraction of mobile dislocations become immobilized over some elapsed time of a deformation period. Because of the long-range elastic interaction among dislocations involved, the dislocation source operation is strongly influenced by the energetics and kinetics of dislocation annihilation at sinks (see Fig. 1). In addition, with respect to a group of dislocation loops, immobilization of certain segments of the loops (e.g., the pinning and unpinning process of screw dislocations in the L1₂ structure) is another reason why the first term of Eq. (1) is not zero, thus requiring the prescription of both spatial and temporal discretizations for dislocation configuration in the simulation study of dislocation dynamics [75]. Accordingly, the efficiency of dislocation source and multiplication process is closely influenced by all types of resistance against dislocation motion.
References


