THE DEFORMATION MICROSTRUCTURE OF THE Ni-Ni₄Mo SYSTEM

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Abstract—The deformation microstructure of stoichiometric Ni₄Mo and of an off-stoichiometric Ni–Ni₄Mo alloy are examined via transmission electron microscopy. In general, the Ni₄Mo superlattice dislocation does not form in either alloy, although one instance is documented in which the superlattice dislocation was found in the off-stoichiometric alloy. Stacking faults and deformation twins are prevalent in the stoichiometric alloy and in the off-stoichiometric alloy containing large Ni₄Mo precipitates. Although twinning is not a preferred deformation mode for many ordered alloys, a twinning mechanism is proposed which explains the prevalence of such twins in the Ni–Ni₄Mo alloy system.

Résumé—Nous avons étudié par microscopie électronique en transmission la microstructure de déformation de Ni₄Mo stoïchiométrique et de l’alliage Ni–Ni₄Mo non stoïchiométrique. D’une manière générale, on n’observe pas de superdislocations dans les deux alliages (même si nous en avons observé une fois dans l’alliage non stoïchiométrique). Dans l’alliage stoïchiométrique, on observe de nombreux défauts d’empiècement et de nombreuses maîches mécaniques, alors que l’alliage non stoïchiométrique contient de gros précipités de Ni₄Mo. Bien que le macilage ne soit pas le mode de déformation principal de nombreux alliages ordonnés, nous proposons un mécanisme de macilage pour expliquer la prépondérance de telles maîches dans le système Ni–Ni₄Mo.


INTRODUCTION

Recent investigations of the Ni–Mo system have concentrated primarily upon the mechanisms whereby the ordered Ni₄Mo phase precipitates from the disordered nickel–molybdenum solid solution. Most of this research has been directed towards the stoichiometric alloy Ni₄Mo [1–6], although more recent studies have considered certain nickel-rich off-stoichiometric alloys as well [7,8]. Despite this emphasis on the transformation mode, only a few investigators have considered the deformation microstructure and the strengthening mechanisms of either the single phase Ni₄Mo ordered alloy [9–12], or of a two-phase Ni–Ni₄Mo alloy [13].

The purpose of this paper is to document and to discuss the deformation microstructure of the fully ordered Ni₄Mo alloy and of an off-stoichiometric Ni–Ni₄Mo alloy.

Before proceeding to the main body of the paper, a brief review of the Ni₄Mo structure and of the deformation characteristics of ordered alloys is appropriate. The ordered Ni₄Mo structure (Dla, I4/m) is based upon a body-centered tetragonal Bravais lattice, and is a crystallographic derivative of the parent fcc lattice (Al, Fm3m). The Ni₄Mo structure contains ten atoms per unit cell: two Mo atoms and eight Ni atoms. The first and second nearest neighbor atoms of each Ni and Mo atom in the Ni₄Mo structure are given in Table 1. For further details concerning the crystallography of the Ni₄Mo structure, the reader is referred to other articles which review this subject [2,5,9].

Throughout this paper, the Greek letters α and β refer to the Ni terminal solid solution and to the ordered Ni₄Mo phase respectively. When these letters appear in conjunction with Miller indices and direction indices, they refer to the face-centered cubic lattice and to the ordered Ni₄Mo structure respectively.

In ordered alloys, the defect structure of major interest is the superlattice dislocation (or simply the superdislocation) and its attendant planar defects. The total Burgers vector of the superdislocation must necessarily be equal to a superlattice translation vec-
Table 1. First and second near-neighbor (N.N.) atoms of the Ni and Mo atoms in the Ni₄Mo structure.

<table>
<thead>
<tr>
<th>Central Atom</th>
<th>1st N.N.</th>
<th>2nd N.N.</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Ni</td>
<td>Mo</td>
</tr>
<tr>
<td>Mo</td>
<td>12 0 4 2</td>
<td></td>
</tr>
<tr>
<td>Ni</td>
<td>9 3 5 1</td>
<td></td>
</tr>
</tbody>
</table>

tor. For the Dla structure, the equivalent atomic positions are located at the (0, 0, 0)a, and at the \((\frac{1}{4}, \frac{1}{4}, \frac{3}{4})\)b or \((0, \frac{1}{4}, \frac{3}{4})\)c positions. Since the ordered Ni₄Mo phase has a \(\langle c/a \rangle\) ratio only slightly less than one (0.986), the slip planes are expected to be the \(\{111\}_a\) or equivalently the \(\{121\}_b\) planes, and the close-packed directions to be the \(\langle 110 \rangle_a\) or \(\langle 120 \rangle_b\) directions. Fig. 1 shows the \(\{111\}_a\) plane of the fully ordered Ni₄Mo structure. Notice that in each of the close-packed \(\langle 110 \rangle_a\) directions, that five \(\frac{1}{4}\langle 110 \rangle_a\) displacements are required to bring the ordered lattice back into coincidence with itself, as previously pointed out by Ruedl et al. [9]. Although this is a rather large superdislocation, it has the advantage that all of the component unit dislocations have identical Burgers vectors. Hence five unit dislocations of \(\mathbf{b} = \frac{1}{4}\langle 110 \rangle_a\) could conceivably emanate from the same source. Other superdislocations composed of fewer unit dislocations are geometrically possible, as illustrated by Ruedl et al. [9]. Such superdislocations would have a lower self-energy but the Burgers vectors of the unit dislocations would not be identical.

Although twinning is not a common deformation mode of ordered alloys, deformation twins are found in both the stoichiometric and off-stoichiometric alloys of the Ni–Ni₄Mo system. Laves [14] first recognized that deformation twins generally fail to form in ordered alloys since the mechanism which produces a twin in a disordered alloy does not necessarily produce a twin in the ordered structure, in the strict crystallographic sense of the word ‘twin.’ In particular, he demonstrated that the reported ‘twins’ in Fe₃Si (DO₃, Fm 3m) were not true twins, since they were of a different crystal structure (orthorhombic) than the DO₃ (cubic) structure. Nevertheless, twins can form in many ordered alloys if the twinning mechanism is different from that of the disordered alloy. For example, in CuₓAu a true twin can be formed by displacements of \(\mathbf{b} = (a/3)\langle 112 \rangle\) on each successive close-packed plane. Such displacements are equivalent to forming extrinsic stacking faults on every \(\{111\}\) plane or intrinsic stacking faults on every other \(\{111\}\) plane.

In the Ni₄Mo structure, (Dla) a true twin can be produced by the normal f.c.c. deformation twinning mode. That is, the movement of an \((a/6)\langle 112 \rangle\) partial dislocation on each successive close-packed plane may yield a twinned region which retains the Dla crystal structure. Details of this deformation mode are elucidated below.

**EXPERIMENTAL PROCEDURE**

The stoichiometric Ni₄Mo alloy was prepared from powders of each of the component metals; each metal powder having an initial purity of 99.9%. After the powders were mixed and arc-melted, the ensuing alloy was homogenized at 1425 K in an inert atmosphere. The alloy was then cold-rolled into sheet form approximately 0.25 mm thick, and then annealed at 1273 K. At the termination of this second anneal, the stoichiometric alloy was furnace cooled to 1108 K (30 K below the critical ordering temperature) and aged for 3 × 10⁴ min. A final deformation of 1% was induced into the alloy by cold-rolling it at room temperature.

The off-stoichiometric alloy was prepared by the International Nickel Company (INCO) at a composition of Ni–15.6 at.% Mo. This alloy was machined into standard quarter-inch diameter tensile specimens, in accordance with ASTM standard E8-69 [15], and then solution treated at 1273 K, prior to down-quenching the alloy to 948 K or 1025 K for various aging times up to 10,000 min. The tensile specimens were pulled to fracture on an Instron Universal Testing Instrument at a strain rate of 3.3 × 10⁻⁴ s⁻¹. In those tensile specimens which proved to be extremely brittle after the final aging anneal, a plastic strain of 3–4% was induced by deforming the specimens in compression.

The alloy microstructures were examined with a JEM 100B transmission electron microscope,
equipped with a double-tilt stage. The TEM specimens were electropolished via the twin-jet technique in a solution of 1/3 nitric acid and 2/3 methanol at 248 K.

**RESULTS**

The deformation microstructure of Ni₄Mo

The deformation microstructure of fully ordered Ni₄Mo, deformed one percent by rolling, contains numerous stacking faults with widely separated partial dislocations.

As shown in Fig. 2, the stacking faults are often several hundred nanometers wide, which is indicative of a low stacking fault energy. When individual stacking faults were found, such as those designated by the letters A and B in Fig. 2, two techniques of analysis were employed in order to confirm the nature of the faults. The first method, as developed by Gevers et al. [16] makes use of the fringe contrast in a dark-field micrograph of the stacking fault, along with the relative direction of the operating g-vector. The second technique of fault analysis, as described by Art et al. [17] is based upon a comparison of the fringe contrast in both the bright and dark-field micrographs of the stacking faults. The advantage of this latter technique is that the direction of the tilt of the stacking fault in the foil is inherently determined, as well as the nature of the stacking fault. In Fig. 2, if the origin of the arrow marked C is placed at the center of a given stacking fault, it will point towards the side of the stacking fault that intersects the top of the foil. The results from applying both of the above techniques confirm that the stacking faults in Fig. 2 are intrinsic in nature.

In some grains several stacking faults overlap one another, making analysis of these stacking faults extremely difficult. See Fig. 3. When individual or non-overlapping faults were found they also were determined to be intrinsic in nature. The stacking faults in Fig. 3 are viewed under the diffracting conditions of \( \vec{n} = [101]_g \) and \( \vec{g} = [111]_s \). Notice the numerous discontinuities in the fringe contrast along these stacking faults. When this same grain is examined with an operating \( \vec{g} \) of \( \vec{g} = [202]_s \), dislocations are observed which correspond to the discontinuities in the fringe contrast. See Fig. 4 and compare it with Fig. 3. These dislocations are most likely partial dislocations with Burgers vectors of \( 1/6[211]_s \) or \( 1/6[112]_s \). The \( 1/6[121]_s \) partial dislocation is not observed under the given diffracting conditions.

When this same grain is rotated to \( \vec{n} = [110]_g \), the overlapping stacking faults are observed on edge in direct space, and they are found to produce extra reflections which are twin related to the fundamental reflections in reciprocal space. See Figs 5a and 5c. In Fig. 5c the twin axis is the [111]ₗ direction. Only one set of extra twin reflections is observed on the diffracting pattern, which implies that twinning has taken place on only one of two possible [111]ₗ planes that diffract in this reciprocal lattice plane. This implication is confirmed by imaging a twin reflection in dark-field (Fig. 5b) and comparing it with the corresponding bright-field image (Fig. 5a). Since the twinned regions are relatively thin, the twin reflections are elongated in the [111]ₗ direction, and in fact a

Fig. 2. Stacking faults in fully ordered Ni₄Mo, deformed 1%, as observed along the [101]ₗ zone axis with \( \vec{g} = [111]_s \).

Fig. 3. Overlapping stacking faults in fully ordered Ni₄Mo, deformed 1%, as observed along the [101]ₗ zone axis with \( \vec{g} = [111]_s \).

Fig. 4. Fully ordered Ni₄Mo deformed 1% as observed along the [101]ₗ zone axis for \( \vec{g} = [202]_s \). The area of this micrograph corresponds to approximately the same area as observed in Fig. 3.
faint streak of diffracted intensity lies in this same direction.

A comparison of Figs 5a and 3 indicates that a one-to-one correspondence exists between the twins in Fig. 5a and the overlapping stacking faults in Fig. 3. Thus the deformation twins are composed of a series of overlapping intrinsic stacking faults.

When the microstructure of deformed Ni$_3$Mo is observed along either the $\langle 100 \rangle$, or the $\langle 112 \rangle$, zone axes, it is possible to view the antiphase boundaries between unit dislocations by imaging the microstructure with a g-vector of $\mathbf{g} = 1/5 \langle 420 \rangle$. The micrographs in Fig. 6 are all of approximately the same area of the microstructure, but each micrograph is taken under different diffracting conditions. Figure 6a shows stacking faults between many, but not all of the dislocations. For instance, the dislocation just above the letter A in this figure is not associated with a stacking fault; whereas a stacking fault is observed just below the letter B in this same micrograph. Figures 6b and 6c display the antiphase boundaries as observed in bright-field ($\mathbf{g} = 1/5 \langle 042 \rangle$) and in the
The deformation microstructure of an off-stoichiometric alloy

The deformation microstructure of the off-stoichiometric alloy in which no ordered Ni$_3$Mo precipitates are present, but only short range order, displays neither stacking faults nor pairs or groups of dislocations, as might be expected of alloys with short-range order (SRO). The dislocations that are present are found in broad slip bands, with the slip planes along the traces of the \{111\}$_{\alpha}$ planes. In reciprocal space, the short-range order reflections are still present after deformation, albeit they are fainter than the same reflections from an undeformed region of the same specimen.

Specimens aged at 948 K (675°C), with a corresponding precipitate size of 7.5 nm are void of stacking faults and any identifiable superdislocations. Most of the unit dislocations are congregated into a relatively few narrow slip bands as shown in Fig. 7a. The superlattice dark-field micrograph in Fig. 7b reveals comparatively few Ni$_3$Mo precipitates in the slip bands. In reciprocal space, the superlattice reflections are more diffuse than those from an undeformed region of the same specimen.

The decrease in the intensity of the SRO reflections in specimens in which only short-range order is present, and the increase in the diffuseness of the superlattice reflections in specimens with small precipitates, results from dislocations destroying the respective types of order. In the latter instance, the increase in the diffuseness of the superlattice reflections indicates that the dislocations are cutting the precipitates and effectively reducing their size, rather than looping around the precipitates. Some of the ordered precipitates in the slip bands are cut to the extent of being destroyed, as evidenced by their diminished number in the slip bands relative to the undeformed matrix.

Certain heat treatments result in extremely brittle specimens, so as to preclude plastic deformation in tension and to necessitate deformation in compression. An example of such a heat treatment is an aging time of 10,000 min at a temperature of 1023 K (750°C), which produces precipitates 80-100 nm in size as measured perpendicular to the Dia c-axis. A specimen which had been aged in this manner was deformed in compression by 3.5%. The resulting microstructure contains stacking faults which are unimpeded and undeviated by the presence of the precipitates. See Fig. 8. The stacking faults seem to nucleate at grain boundaries, since the stacking faults often extend from a grain boundary into the grain and terminate either with a partial dislocation within the grain or at the intersection with another stacking fault. Analysis of the stacking faults in order to determine their specific nature (i.e. intrinsic or extrinsic) was hindered by the fact that the faults pass through both the matrix and the precipitates, which consequently disrupts the corresponding fringe contrast.
Fig. 8. The deformation microstructure of the off-stoichiometric alloy aged at 750°C (1023 K) for 10,000 min and subsequently deformed in compression by 3.5%. Stacking faults are observed along the traces of the [111], planes as viewed along the [001], zone axis.

When the above specimen is tilted to the [110], zone axis, streaks are observed in the [111], direction, with extra spots or reflections along these streaks. These extra reflections are twin related to the fundamental reflections across the (111), reciprocal space plane, and they are identical to the twin reflections from deformed Ni$_5$Mo (see Fig. 5c). The bright- and dark-field micrographs of the twinned regions corresponding to the twin reflections are shown in Figs 9a and 9b, respectively.

Based upon the previous analysis of the deformation twins in ordered and deformed Ni$_5$Mo, and upon the fact that the stacking faults pass through both the matrix and the precipitates in the two-phase alloy, it is reasonable to conclude that the deformation twins in Figs 9a and 9b are composed of a series of stacking faults.

The deformation microstructure of an off-stoichiometric Ni$_5$Mo alloy therefore has the following characteristics as a function of aging time. At short aging times when only short-range order is present or when the ordered precipitates are sufficiently small, f.c.c. unit dislocations pass through the matrix phase, cutting the precipitates if the latter are present. When the precipitates are larger, f.c.c. partial dislocations and their attendant stacking faults pass through both the matrix and the precipitate phases. These stacking faults may in turn constitute deformation twins.

The Ni$_5$Mo superdislocation

In general the superdislocation of the ordered phase does not form in either the fully ordered stoichiometric alloy or in the off-stoichiometric alloy. However, occasionally groups of five dislocations per group were observed in the off-stoichiometric alloy, as shown in Fig. 10. These three groups of dislocations are followed by individual unit dislocations. All of the dislocations have identical Burgers vectors in the [101], direction as observed along the [001], zone axis.

Fig. 9. (a) Bright-field micrograph of the deformation microstructure as observed along the [001], zone axis under multiple-beam diffracting conditions. (b) Dark-field micrograph of deformation twins produced by imaging a twin reflection.
It was not possible to image the APB between the dislocations as the superlattice reflections were too weak and diffuse to permit the operation of a strong \( g = 1/5(420) \). The distribution of the unit dislocations indicates that three superdislocations are required to completely sever the \( \text{Ni}_4\text{Mo} \) precipitates so that the slip plane no longer lies within the precipitates, but only along the precipitate-matrix boundary. For a Burgers vector of \( b = 1/2[101] \), and using the lattice parameter values of Guthrie and Stansbury [19], the length of the Burgers vector for a unit dislocation is 0.254 nm. Thus the average precipitate diameter, \( d \), can be approximated as

\[
d = n|b| = 15(0.254) \approx 3.8 \text{ nm}
\]

where \( n \) is the total number of unit dislocations in the identifiable superdislocations. Dark-field electron microscopy of the superlattice reflections reveals that the precipitates are 3–5 nm in size. This latter result is in good agreement with the precipitate size calculated from the above equation, and is consistent with the suggestion that the dislocations in Fig. 10 are f.c.c. unit dislocations, and that each group of five dislocations constitutes a superdislocation of the \( \text{Ni}_4\text{Mo} \) structure.

The separation between the unit dislocations within a superdislocation is the result of an energy balance among the self-energies of the dislocations, the interaction energy between the dislocations, and the antiphase boundary energy. For two identical, parallel unit dislocations, the first two energy contributions cause the unit dislocations to repel each other, while the third term (the APB energy) tends to hold the unit dislocations together at a fixed separation.

The change in the spacing between the unit dislocations within a superdislocation in Fig. 10 results from the dislocations shearing the ordered precipitates which traverse the slip plane. Each unit dislocation as it passes along the slip plane, not only cuts these precipitates, but also displaces the cut portions with respect to each other by one Burgers vector. This process reduces the total antiphase boundary area along the slip plane, while creating additional precipitate-matrix interfacial area. As a result the total antiphase boundary energy between unit dislocations is reduced. Thus as a superdislocation composed of more than
two unit dislocations moves along a slip plane the separation should be the least between the leading two dislocations and the greatest between the final two dislocations. From this analysis, the direction of motion of the dislocation array in Fig. 10 is in the direction of the accompanying arrow.

**DISCUSSION**

Probably the most significant observation concerning the deformation microstructure of the Ni–Ni$_5$Mo system is the general absence of an identifiable superdislocation both in the fully ordered stoichiometric alloy, as well as in the partially ordered off-stoichiometric alloy. (The three groups of 5 dislocations in Fig. 10 were a rarity.) The most probable explanation for this absence is the relative complexity of the superdislocation, since five unit dislocations are necessarily required for its formation. This complexity inhibits the nucleation of the full superdislocation as a discernible entity when deformation takes place at room temperature. Lakso and Marcinkowski [20] investigated a similar situation in ordered Fe$_3$Si (DO$_3$ structure) in which the superdislocation is composed of four unit dislocations. The superdislocations do not form at 243 K, but are present at 298 K. In the present case of ordered Ni$_5$Mo, a deformation temperature of 298 K is apparently too low for the nucleation of the superdislocation, which may form at higher deformation temperatures.

Recent work by Greggi and Soffa [21] on a Cu–Ti alloy, in which the Da structure (Cu$_6$Ti) [22] also forms; indicates that deformation of that alloy at room temperature results in numerous superdislocations. These results and the general absence of superdislocations in our Ni–Mo alloy are consistent with the hypothesis proposed by Lakso and Marcinkowski [20], since the Cu–Ti alloy at room temperature is at a higher deformation temperature relative to its absolute melting point than is the Ni–Mo alloy.

Another significant observation concerning the deformation microstructure is the presence of large (wide) stacking faults in the ordered stoichiometric alloy (see Fig. 2) and also in the partially ordered off-stoichiometric alloy containing precipitates 80–100 nm in size (see Fig. 8). These stacking faults are not observed in the supersaturated or terminal solid solution and therefore the stacking fault energy of the ordered phase must be lower than that of the disordered solid solution.

In order to ascertain the reason for the prevalence of such stacking faults in the ordered or partially ordered structure, it is necessary to analyze the crystallographic effect of the passage of a partial dislocation on a slip plane compared to the movement of a unit dislocation. The movement of a unit dislocation with a Burgers vector of $\mathbf{b} = 1/2[110]_a$ on the $(111)_a$ plane through the ordered structure ($\text{Dla c-axis parallel to the [001]}_a$, direction) results in a net change of four nearest neighbor Ni–Mo bonds to 2 Ni–Ni bonds and 2 Mo–Mo bonds over an area of approximately 0.56 nm$^2$. On the other hand, the passage of a partial dislocation with a Burgers vector of either $\mathbf{b} = 1/6[12\bar{1}]_a$, on the same $(111)_a$ plane, does not result in any net change in the number of Ni–Mo bonds. However, a partial dislocation with $\mathbf{b} = 1/6[112]_a$ would result in a net change in the number of such bonds. Since the movement of two of the three partial dislocations on a given close-packed plane does not create a net change in the number of each kind of near-neighbor bonds, the associated antiphase boundary energy should be relatively low. This particular characteristic of the Dla crystallography accounts, in part, for the numerous stacking faults observed in the fully ordered stoichiometric alloy and in the partially ordered off-stoichiometric alloy containing large Ni$_5$Mo precipitates.

An additional feature of the deformation microstructure with regard to the stacking faults in the ordered Ni$_5$Mo alloy is that these faults tend to form in overlapping groups on parallel close-packed planes, so as to constitute twins, or more specifically deformation twins. Venable [23] has demonstrated that a twinned crystal may be built up by a series of intrinsic stacking faults on successive close-packed planes in the f.c.c. lattice. This particular model inherently requires that these stacking faults be produced by partial dislocations of the same Burgers vector on each close-packed plane.

An unusual characteristic of such deformation twins in ordered Ni$_5$Mo is that if they are composed of the same partial dislocations which yield low energy stacking faults, then not only will these regions be twins of the f.c.c. lattice, but they will also be true crystallographic twins of the ordered Dla structure. That is, the twinned regions will retain the Dla structure although it will be of a different orientation variant than that of the untwinned region. In order to understand how such a twin might form, compare Fig. 11a with Fig. 1. Figure 11a shows the positions of the Mo atoms on three successive $(111)_a$ planes. The triangles represent the Mo atoms in the lowest (bottom) plane, the circles indicate the positions of the Mo atoms in the middle plane, and the stars mark the Mo atom positions in the third or top plane. The projection of the Dla c-axis lies in the direction delineated by a triangle, a circle, and a star in succession. In Fig. 11a, the Dla c-axis is in the [001]$_a$ direction and its projection is along the [112]$_a$ direction. Using the bottom $(111)_a$ plane (the triangles) as a reference plane, let a partial dislocation of $\mathbf{b} = 1/6[12\bar{1}]_a$ pass between the bottom and middle planes, thus shifting the Mo atoms on the middle and top planes (the circles and stars, respectively) in the [12\bar{1}]$_a$ direction by $(a/\sqrt{6})$. Now if an identical partial dislocation passes between the middle and the top planes, the Mo atoms on the top plane (the stars) will slip in the [12\bar{1}]$_a$ direction again by $(a/\sqrt{6})$. The net result of these slips on successive $(111)_a$ planes is shown in Fig. 11b. Again the Dla c-axis projects along
plane in this case. The atoms are reflected across the twinning plane and the (121) plane.

The fact that it is possible for the twinned region to retain the Dla structure and for the habit plane of the twin to be a low energy combination of stacking fault and perpendicular twin boundary, is consistent with the prevalence of deformation twins in the deformed microstructure of Ni$_3$Mo and of the off-stoichiometric alloy containing large precipitates.

Various models of dislocation mechanisms have been proposed which attempt to explain how deformation twins in general nucleate and grow within an alloy. Two basic approaches have been pursued in attempting to explain their formation. The first approach, as proposed by Orowan [24] envisages a twin embryo nucleating homogeneously in a region of high stress concentration. The second approach as developed by Venables [25], Cohen and Weertman [26], Hirth [27], and Mahajan and Chin [28] hypothesizes that the twin nucleates from a structural heterogeneity, such as a particular dislocation arrangement within the crystal.

The existing evidence from the deformation of both the stoichiometric and off-stoichiometric Ni–Mo alloys is that the stacking faults, which constitute the deformation twins, nucleate at grain boundaries and then move into the grains. Thus the deformation twins in the Ni–Ni$_3$Mo system nucleate at a structural heterogeneity.

CONCLUSION

The deformation microstructures of the stoichiometric Ni$_3$Mo alloy and of an off-stoichiometric Ni–Ni$_3$Mo alloy have several unique features which are due in part to the crystallography of the Dla (ordered Ni$_3$Mo) structure. In general the superdislocation does not form as an identifiable entity, even in the fully ordered stoichiometric alloy, when deformation takes place at room temperature. In these instances in which the superdislocation was observed, it consisted of five identical f.c.c. unit dislocations ($\{011\}_h^1$). Wide, intrinsic stacking faults are also observed in the deformed microstructures of both the fully ordered Ni$_3$Mo alloy and of the off-stoichiometric alloy containing large precipitates, indicating that the stacking fault energy is considerably lower in the Ni$_3$Mo structure than in the disordered solid solution. This low stacking fault energy in the ordered phase is due to the fact that the passage of two of the possible three partial dislocations on a $\{111\}_h$ close-packed plane does not change the number of each kind of first nearest-neighbor atom pairs across the slip plane. A third feature of the deformation microstructure is the formation of deformation twins composed of and formed by the aforementioned low energy stacking faults. Consequently the twinned regions retain the Dla structure of the untwinned region, albeit the twinned region is of a different orientation variant.
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