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STUDY ON DEFORMATION-INDUCED MARTENSITIC TRANSFORMATION
IN AUSTENITIC IRON-BASE ALLOYS

STUDY ON DEFORMATION-INDUCED MARTENSITIC TRANSFORMATION
IN AUSTENITIC IRON-BASE ALLOYS

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CHAPTER I INTRODUCTION

It is well known that martensitic transformation can be induced by deformation in metastable austenitic steels. A remarkable increase in elongation is sometimes observed when martensitic transformation is induced during deformation, and this phenomenon is called transformation-induced plasticity (TRIP)⁽¹⁾. There are many reports^{(2)~(8)} about TRIP phenomenon. According to them, it is considered that martensite induced by deformation works to suppress necking and to increase in uniform elongation. Zackay et al⁽¹⁾ proposed a TRIP steel with high toughness and high strength through TRIP phenomenon. Usual strengthening technique (precipitation hardening or solution hardening etc) accompanies a decrease in toughness. This is why the TRIP phenomenon is considered to be very important for industrial applications.

However, deformation-induced martensitic transformation, which is the fundamental point of TRIP phenomenon, has not been made clear. The purpose of this study is to obtain a clear understanding of the role of applied stress and plastic strain during deformation for the deformation-induced martensitic transformation.

In the following section of this Chapter, a general survey of the mechanism of martensitic transformation is described. The historical current for the study on the deformation-induced martensitic transformation, and recent studies on the deformation-induced martensitic transformation are critically examined.

1.1 General Survey of the Mechanism of Martensitic Transformation

Martensitic transformation is characterized by the cooperative shear movement of atoms and by the existence of orientation relationships between the parent phase and the martensite.

As the mechanism of martensitic transformation, several models have been proposed.

Kurdjumov and Sachs⁽⁹⁾ reported K-S relationship: $(111)_\gamma // (101)_\alpha$, $[\bar{1}01]_\gamma // [\bar{1}\bar{1}1]_\alpha$ and proposed K-S shear mechanism. The mechanism is shown in Fig1.1. Martensite was considered to be generated from an austenite parent by the 1st shear on $(111)_\gamma$ plane along $[\bar{1}\bar{1}2]_\gamma$ direction and 2nd shear on $(1\bar{2}1)_\gamma$ plane along $[10\bar{1}]_\gamma$ direction. Nishiyama⁽¹⁰⁾ found N relationship: $(111)_\gamma // (011)_\alpha$, $[\bar{1}\bar{1}2]_\gamma // [0\bar{1}1]_\alpha$ and proposed N shear mechanism which is also shown in Fig.1.1. Martensite was considered to be generated from an austenite parent by the shear on $(111)_\gamma$ plane along $[\bar{1}\bar{1}2]_\gamma$ direction lying in the plane. Bogers and Burgers⁽¹¹⁾ proposed another mechanism which shown in Fig.1.2. They considered that martensite was generated from austenite by two shears, each of which has a displacement of $(1/18)a_\gamma \langle 112 \rangle_\gamma$ on the $\{111\}_\gamma$ plane and of $(1/12)a_\gamma \langle 112 \rangle_\gamma$ on the other $\{111\}_\gamma$ plane, respectively.

Although above three mechanisms explain required orientation relationships, those are not adequate when one consider the habit plane. If the shear occurs on the $(111)_\gamma$ plane, then the habit plane of the martensite should be the $(111)_\gamma$ plane. However actual habit planes which have been reported for Fe-highNi-C alloys are $\{225\}_\gamma^{(12),(13)}$, $\{259\}_\gamma^{(14),(15)}$ and $\{3\ 15\ 10\}_\gamma^{(16),(17)}$, and quite different from

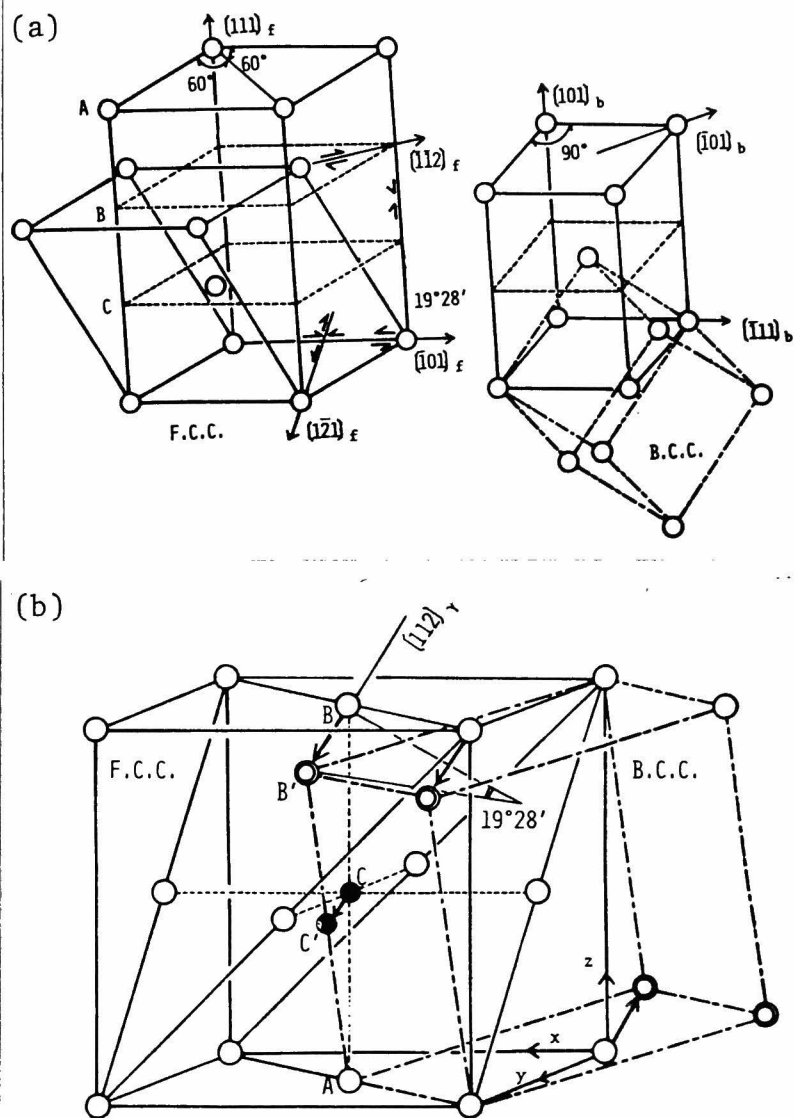


Fig.1.1 Shear mechanism of martensitic transformation.
 (a) proposed by Kurdjumov and Sachs⁽⁹⁾.
 (b) proposed by Nishiyama⁽¹⁰⁾.

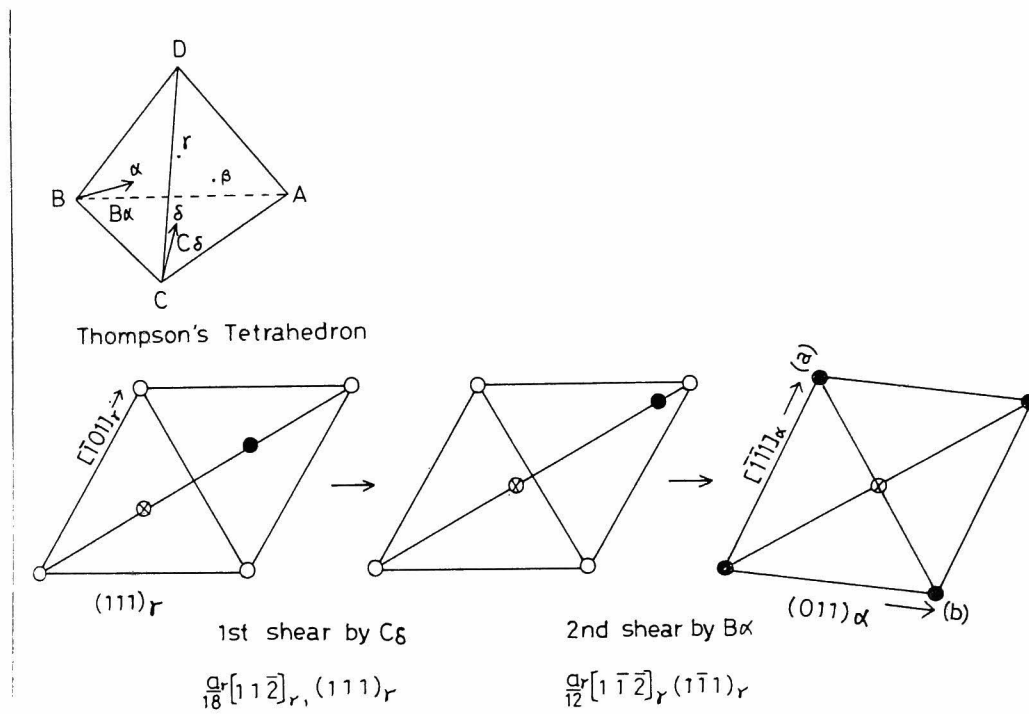


Fig.1.2 Double shear mechanism by Bogers and Burgers⁽¹¹⁾.

$\{111\}_\gamma$. Greninger and Troiano⁽¹⁴⁾ measured the shear angle from the surface relief accompanying martensitic transformation in Fe-Ni-C alloys. The shear angle was $10^\circ 45'$ and was inconsistent with the 19.5° predicted from the $(111)_\gamma$ shear mechanism. Therefore, they suggested that another shear had to be added in order to produce the bct lattice which determined experimentally. The second shear was verified by electron microscopic results which showed twinning on $\{112\}_\alpha$ plane in martensite⁽¹⁸⁾

Bowles and Mackenzie^{(19)~(22)} (B-M) and Wechsler, Lieberman and Read (W-L-R)^{(23),(24)} developed the theory which introduced the second shear in a logical manner. B-M theory and W-L-R theory are equivalent and are called phenomenological theories, and have been recognized as appropriate theories. In the phenomenological theory, a martensitic transformation is considered to consist of a lattice deformation (in the case of $\gamma \rightarrow \alpha'$ transformation, this is the Bain distortion: $B^{(25)}$), a lattice invariant shear (the second shear which mentioned above: P) and a lattice rotation (R). If the B , P and R are matrices, the total shape strain (P_1) should be described as $P_1 = R P B$. The total shape strain, the habit plane and the orientation relationship which were predicted by the phenomenological theory coincided well with experimental data^{(14),(15),(26)}.

1.2 Historical Current and Recent Studies on the Deformation-Induced Martensitic Transformation

1.2.1 Effect of stress

The effect of applied stress on martensitic transformation

has been studied for a long time. In 1932, Scheil⁽²⁷⁾ found that M_S temperature increased under tensile or compressive stress. After that, many papers have been published^{(28)~(34)}. Since martensitic transformation is a kind of shear deformation, it is clear that shear stress aids the transformation. The work done by the applied stress to induce the transformation is called the mechanical driving force. There are two different views about the effect of stress on martensitic transformation. One is that stress assists the total shape strain during the transformation. The other is that stress assists only a special component of the total shape strain accompanying the transformation.

Patel and Cohen⁽²⁸⁾ considered that the applied stress aided the total shape strain accompanying the transformation, and they proposed the following equation for the mechanical driving force(W_m).

$$W_m = \tau \gamma_0 + \sigma \epsilon_0 \quad \text{----- (1-1)}$$

where τ : the shear stress resolved along the habit plane

γ_0 : the transformation shear strain along the habit plane

σ : the normal stress resolved perpendicular to the habit plane

ϵ_0 : the normal component of the transformation strain to the habit plane.

Fig.1.3 shows chemical free energy of martensite and austenite as a function of temperature. T_0 is the temperature at which austenite and martensite are in equilibrium. When austenite is supercooled to M_S , and chemical driving force reaches to the critical value($\Delta G_{M_S}^{\gamma \rightarrow \alpha'}$) at which martensitic transformation starts. When stress is applied at T_1 (above M_S temperature) for example, martensitic transformation can be induced when the chemical($\Delta G_{T_1}^{\gamma \rightarrow \alpha'}$) and mechanical driving

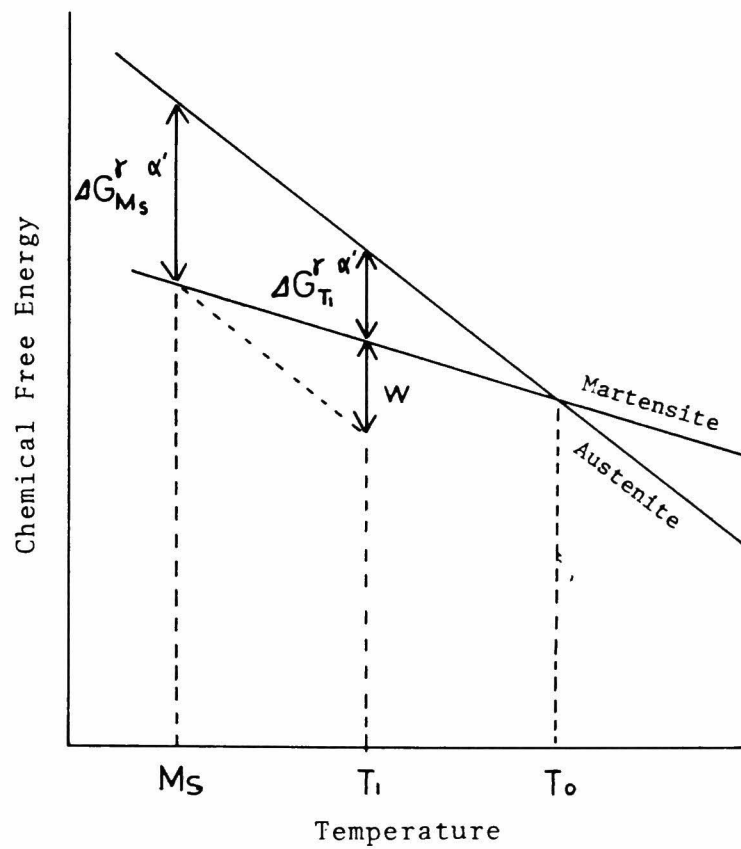


Fig.1.3 Chemical free energy of martensite and austenite as a function of temperature.

force(W_m) reaches $\Delta G_{M_S}^{\gamma \rightarrow \alpha'}$. Patel and Cohen⁽²⁸⁾ measured the increase in M_S temperature under elastic stress and found that the increase in M_S calculated from Eq.1.1 coincided well with the measured data.

Stone and Thomas⁽³⁵⁾ examined the observed martensite variant induced by the applied stress in an austenitic steel. If stress is applied during the transformation, a special variant of martensite with a certain crystallographic orientation to the parent phase will be formed preferentially. They concluded that the stress aided the total shape strain of martensitic transformation. Ahlers⁽³⁶⁾ reported almost the same results using Fe-Ni single crystals.

On the other hand, Higo et al⁽³⁷⁾ and Kato et al^{(38),(39)} suggested that martensitic transformation took place by double shear mechanism proposed by Bogers and Burgers⁽¹¹⁾ and concluded from their analysis of the martensite variants transformed under stress that the stress aided the 1st shear. Olsen and Jesser^{(40),(41)} report that the applied stress aided the shear on $\{111\}_\gamma$ plane along $\langle 111 \rangle_\gamma$ direction which corresponds to K-S or N relationship from their analysis of martensite variants transformed under stress.

As is discussed above, it is clear that the understanding of the effect of stress on the transformation is to make clear whether the applied stress aids a special component of the total shape strain of martensitic transformation or aids the total shape strain of the transformation.

If the applied stress aids only the shear on $\{111\}_\gamma$ plane along $\langle 11\bar{2} \rangle_\gamma$ direction, the difference between the effect of tensile and compressive stress on the martensitic transformation would not appear, in the case of polycrystalline specimen. However, it is known that tensile and compressive stresses make a different increase

in M_s temperature. From this, the applied stress is considered to aid the total shape strain including the volume change accompanying the transformation. However, the difference between the effect of tensile and compressive stress on the transformation has been studied only about the change in M_s temperature. If the applied stress aids the total shape strain including the volume change accompanying the transformation, the difference between the effect of tensile and compressive stress on the transformation should appear in the amount of martensite which induced by the applied stress. Then, in this study, the amount of martensite induced by the stress was studied to confirm the effect of mechanical driving force on the transformation which will be described in detail in Chapter II.

1.2.2 Effect of strain

Olson and Cohen⁽⁴²⁾ showed the critical stress for martensite formation as a function of temperature as Fig.1.4 using the experimental results of Bolling and Richman^{(43)~(46)}. They classified deformation-induced martensitic transformation into the stress-assisted and the strain-induced one from Fig.1.4. In Fig.1.4, M_s^0 is the temperature, below which yielding under applied stress is initiated by the onset of martensite formation, and above which yielding under stress is initiated by regular process(slip) in the parent phase. They called the nucleation below M_s^0 as being stress-assisted, and the nucleation above M_s^0 as being strain-induced. Tokizane⁽⁴⁷⁾ obtained the same results as Fig.1.4 using an Fe-Ni-C alloy. In Fig.1.4, the critical stress for martensite formation

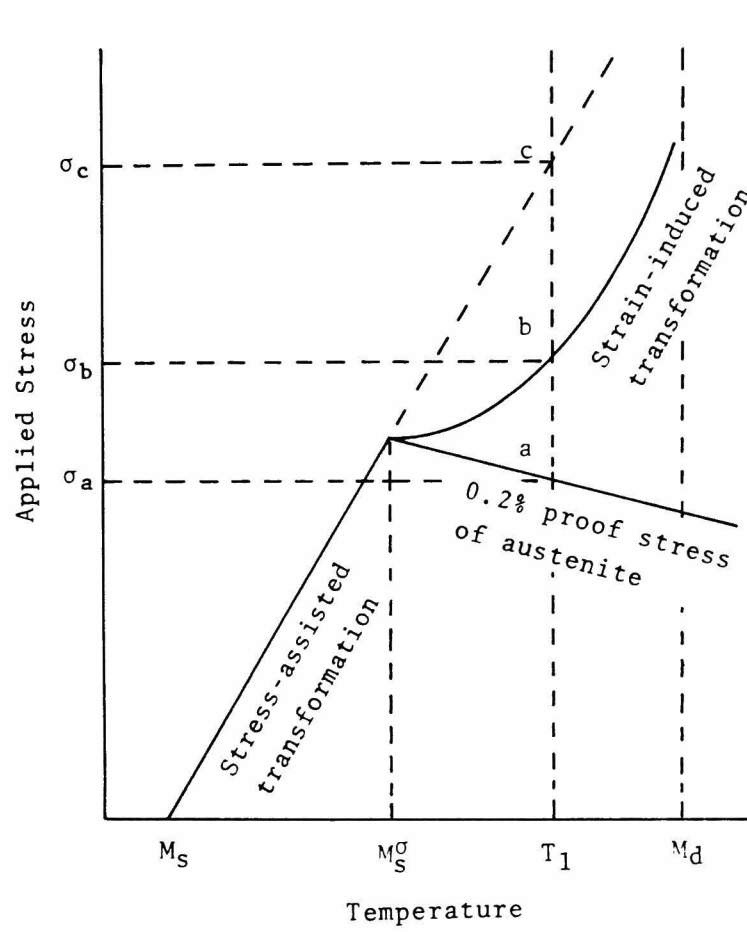


Fig.1.4 Critical stress for martensite formation as a function of temperature.

increases linearly with increase in temperature in the temperature range between M_S and M_S^G . However, at T_1 (above M_S^G), austenite begins to deform at the stress (σ_a) and is required to strain-harden up to the stress (σ_b) to form martensite. The stress (σ_b) is considerably lower than the value expected (σ_c) by extrapolating the stress-temperature line in the temperature range between M_S and M_S^G . This decrease of critical applied stress (i.e., $\sigma_c - \sigma_b$) for martensite formation is considered to be the assist of plastic strain introduced before martensite was formed. About the effect of strain on the deformation-induced martensitic transformation, there are two different views. One is the strain-induced nucleation proposed by Olson and Cohen^{(42), (48)}, and the other is that the stress is concentrated at obstacles (grain boundary, twin boundary etc.), and that concentrated stress aids the transformation.

1.2.2.1 Strain-induced nucleation

Olson and Cohen^{(42), (48)} proposed that embryo of martensite could be formed at the intersection of two shear bands. As an typical example of such a formation of embryo, they referred to Venables',⁽⁴⁹⁾ observation (Photo.1.1) in which α' -martensite was formed at the intersection of two ϵ' -bands in 18-8 stainless steel. They considered that the intersection of two shear bands, each of which has a displacement of $(1/18)a_\gamma \langle 112 \rangle_\gamma$ on a $\{111\}_\gamma$ plane and of $(1/12)a_\gamma \langle 112 \rangle_\gamma$ on another $\{111\}_\gamma$ plane, respectively, is equivalent to the Bogers-Burgers double shears and forms an embryo of martensite. The schematic picture of the Olson-Cohen model is shown in Fig.1.5. As examples of such shear bands, they considered

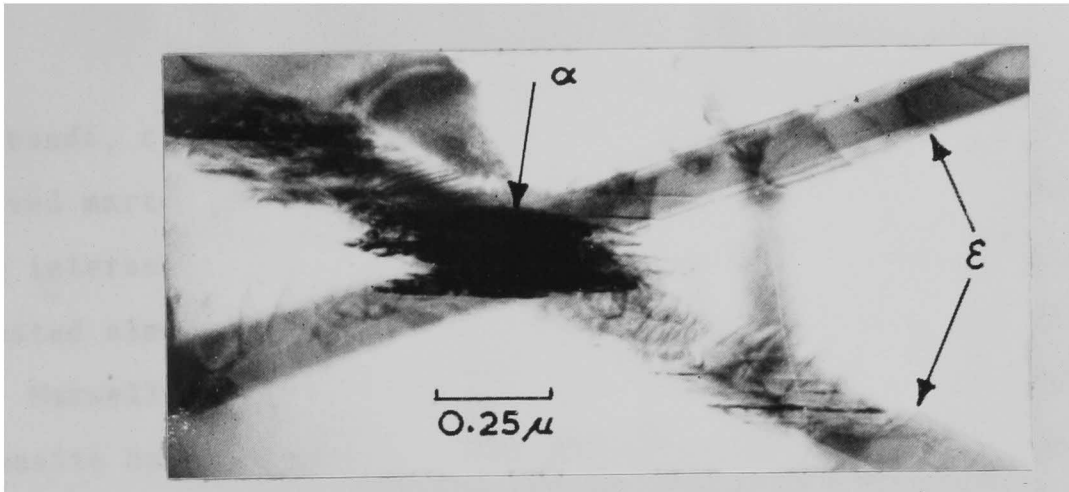


Photo.1.1 The formation of α' -martensite from ϵ' -platelets.
(by Venables⁽⁴⁹⁾)

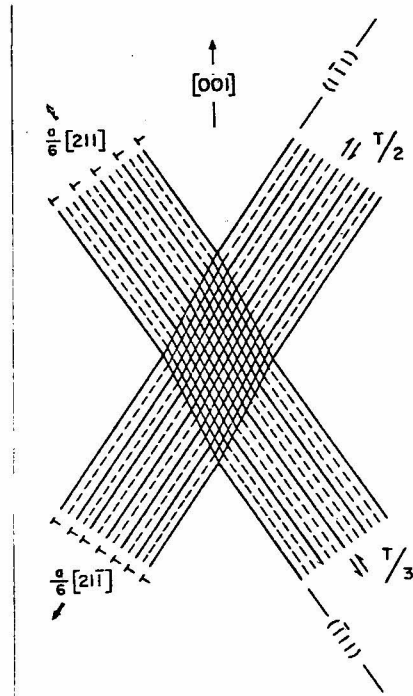


Fig.1.5 Schematic illustration of intersecting shears due to two arrays of $(a_\gamma/6)\langle 112 \rangle_\gamma$ partial dislocations in austenite. One array ($T/3$) has partial dislocations on every third $\{111\}_\gamma$ plane and averages one-third of a twinning shear, while the other array ($T/2$) has partial dislocations on every second $\{111\}_\gamma$ plane and averages one-half of a twinning shear. The resulting doubly-faulted intersection has an exact bcc structure. (by Olson and Cohen⁽⁴²⁾)

slip-bands, ϵ' -bands and deformation twins. Lacroisey and Pineau⁽⁵⁰⁾ observed martensite which nucleated at the twin boundary and grew along intersecting deformation bands in an Fe-Ni-Cr-C alloy, and suggested almost the same mechanism as that of Olson and Cohen.

Maxwell et al^{(51),(52)} reported that deformation-induced martensite had the morphology different from thermally transformed martensite. They observed a kind of banded structure (a typical example is shown in Photo.1.2) in Fe-Ni-C alloys, and reported that such a structure is a new type of strain-induced martensite and it is formed by the mechanism proposed by Olson and Cohen. They also reported that such a structure could be formed at the temperature more than 150K above M_s in an Fe-Ni-C alloy. This temperature is extremely higher than the M_s . So, it is doubtful whether or not the banded structure is martensite. Then, in this study, the banded structure was examined by optical and electron microscope.

1.2.2.2. Internal stress concentration

In the case of deformation-induced martensitic transformation, martensites are frequently observed at the grain boundaries or the annealing twin boundaries⁽⁴⁷⁾. During plastic deformation, many defects which have high local stress field are introduced in the parent austenite. Such stress field is considered to aid the nucleation of martensite.

Suzuki et al⁽⁵³⁾ and Ueda and Fujita⁽⁵⁴⁾ reported that martensite was observed at the intersection of two slip bands, and they suggested that martensitic transformation was induced by the stress concentration due to piled-up dislocations at such

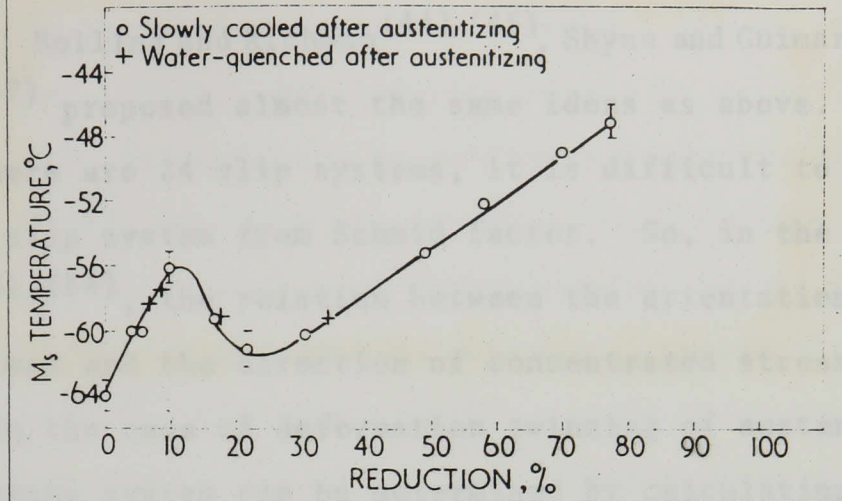


Fig.1.6 A variation in M_s with degree of reduction of austenite. (by Gooch and West⁽⁵⁸⁾)

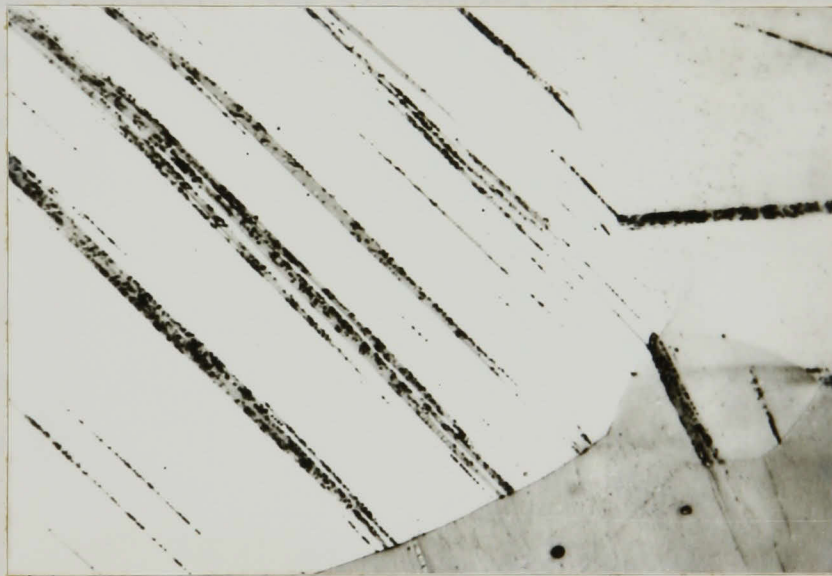


Photo.1.2 A typical microstructure of banded structure which formed by deformation of 30% reduction at 373K in an Fe-20%Ni-0.73%C alloy. (x 700)

intersections. Bolling and Richman^{(44),(45)}, Shyne and Guimaraes^{(55)~(57)} and Tokizane⁽⁴⁷⁾ proposed almost the same ideas as above.

Since there are 24 slip systems, it is difficult to determine the operating slip system from Schmid factor. So, in the above experiments^{(53),(54)}, the relation between the orientation of martensite formed and the direction of concentrated stress was not made clear. In the case of deformation twinning of austenite, the operating twinning system can be determined by calculating Schmid factor, since there are only 12 twinning systems. Then, in this study, the relation between the variant of martensite induced by deformation and the concentrated stress due to deformation twinning in austenite was examined. The crystallography of deformation twin in austenite is relatively easy to analyze and deformation twin of austenite is considered to be the most suitable to examine the strain-induced nucleation mechanism proposed by Olson and Cohen⁽⁴²⁾. In this study, the martensite which observed at the junction of two deformation twins was also examined whether or not it is formed by the intersection of two shear bands of Bogers-Burgers type suggested by Olson and Cohen.

1.3 Effect of Prior Deformation on the Martensitic Transformation

To separate effects of plastic strain from applied stress on martensitic transformation during deformation, it is necessary to examine the effect of prior deformation on martensitic transformation.

Gooch and West⁽⁵⁸⁾, using Ni-steels, measured the M_s temperature after rolling at 573K. Their results are shown in Fig.16, which reveals that the M_s first rises, then falls and rises again

monotonically with increasing degree of rolling. The M_s first rises due to the residual stress. The lowering of the M_s at 10~20% reduction of rolling is explained by hardening of matrix due to the precipitation of carbides. The rising of the M_s at heavier deformation is due to the decreased carbon content in the austenite because of extensive precipitation of carbides. Tamura et al⁽⁵⁹⁾ and Georgiyeva et al⁽⁶⁰⁾ obtained similar results using Fe-Ni-C alloys.

To prevent the effect of migration of carbon atom, Tokizane⁽⁴⁷⁾ applied deformation at lower temperature (250K) by tension and then measured the M_s temperature using an Fe-Ni-C alloy. His results showed that the M_s first fell at small strain, and then rose with increasing degree of deformation. He considered that the lowering of the M_s at small strain was due to the strain-hardening of matrix and the increase of the M_s at heavier deformation was due to the residual stress. Wallman and Guimarães⁽⁵⁷⁾ showed that in an Fe-Ni alloy, the M_s temperature rose with the degree of rolling at room temperature and reached maximum at about 50% strain. Similar results were also obtained by Guimarães and Shyne⁽⁵⁵⁾.

Generally, the M_s temperature rises by prior deformation above the M_d temperature, and the effect is called mechanical sensitization by Guimaraes^{(55)~(57)}. When austenite is plastically deformed, the regions with high local stress are introduced. Such local stress is considered to aid martensitic transformation. However, the optical microscopic observation of defects is difficult, and almost all explanations on this subject in these studies^{(55)~(60)} are suppositions from the change of the M_s by prior deformation. Only Breedis⁽⁶¹⁾ reported the effect of prior deformed microstructure of austenite on the transformation using electron microscopy.

However, the nucleation site was not made clear. The deformation twins are the most suitable structure to reveal the effect of the prior deformation on the martensitic transformation since the deformation twins can be observed by optical microscope and also the crystallographic analysis is easy. Then, in this study, the effect of deformation twins of austenite (introduced by prior deformation above the M_d temperature) on the martensitic transformation to make clear the effect of defects in austenite without applied stress.

1.4 Purpose of This Study

The purpose of this study is to reveal the effects of applied stress and plastic strain on the deformation-induced martensitic transformation by making clear the subjects mentioned above.

In Chapter II, quantitative model relating the mechanical driving force and the amount of martensite induced by the applied stress is proposed. The model was verified by comparing the amount of martensite which was calculated from the model with the one measured experimentally. It was confirmed that the effect of applied stress on the martensitic transformation was expressed by the mechanical driving force.

In Chapter III, the banded structure, which Maxwell et al called a strain-induced martensite, was examined by optical and electron microscope. It was shown that the banded structure is not martensite but deformation twins of austenite.

In Chapter IV, the effect of deformation twinning on the deformation-induced martensitic transformation was examined to

reveal the effect of plastic strain on the deformation-induced martensitic transformation. From this study, it was made clear that martensite was formed by the aid of the concentrated stress created by the impingement of deformation twins on the grain boundaries or the annealing twin boundaries. The role of plastic strain was considered to create the regions where the stress was concentrated in the matrix. From the crystallographic analysis of the martensite which formed by deformation, the stress was considered to aid the total shape strain of martensitic transformation rather than a part (1st shear) of the total shape strain. Martensite which observed at the junction of two deformation twins in tension was considered not to be formed from the embryo created by the intersection of two shear bands as suggested by Olson and Cohen. In the case of compression, martensite which can be identified clearly was not seen to form preferentially at the intersection of two deformation twins that were considered to be corresponding to Burgers-Burgers double shears. From these results, it was concluded that the strain-induced nucleation mechanism which proposed by Olson and Cohen is not prevailing mechanism for deformation-induced martensitic transformation.

In Chapter V, the effect of prior deformation on martensitic transformation was examined. The purpose of this investigation is to separate the effect of strain from the effect of applied stress on the deformation-induced martensitic transformation. The nucleation site of thermally transformed martensite was studied by optical microscope. Martensites were frequently observed at the grain boundaries or the annealing twin boundaries at which deformation

twins impinged(that were introduced by prior deformation). The local residual shear stress at the boundaries was considered to aid martensitic transformation. The concentrated stress which was formed by the impingement of deformation twins was considered not to be released completely by unloading.

References for Chapter I

- 1) V. F. Zackay, E. R. Parker, D. Fahr and R. Busch: Trans. ASM, 60(1967), 252.
- 2) J. P. Bressanelli and A. Moskowitz: Trans. ASM, 59(1966), 223.
- 3) Y. Fukase, K. Ebato, N. Okubo and S. Murao: J. Japan Inst. Metals, 32(1968), 38.
- 4) F. Abrassart, F. Lecroisey and A. Pineau: Mem. Sc. Rev. Met., 66(1969), 805.
- 5) G. P. Sanderson and D. T. Llewellyn: JISI, 207(1969), 1129.
- 6) I. Tamura, T. Maki, H. Hato and K. Aburai: J. Japan Inst. Metals, 33(1969), 1383.
- 7) I. Tamura, T. Maki and H. Hato: Trans. ISIJ, 10(1970), 163.
- 8) I. Tamura, T. Maki, H. Hato, Y. Tomota and M. Okada: Proc. of 2nd Int. Conf. on the Strength of Metals and Alloys. at Asilomar (1970), 901.
- 9) G. V. Kurdjumov and G. Sachs: Z. Phys., 64(1930), 325.
- 10) Z. Nishiyama: Sci. Rep. Tohoku Univ., 23(1934), 637; 25(1936),
- 11) A. J. Bogers and W. G. Burgers: Acta Met., 12(1964), 255.
- 12) H. M. Otte: Acta Met., 8(1960), 892.
- 13) A. B. Greninger and A. R. Troiano: Trans. AIME, 140(1940), 307
- 14) A. B. Greninger and A. R. Troiano: Trans. AIME, 145(1941), 289
Trans. AIME, 185(1949), 590.
- 15) P. M. Kelly and J. Nutting: JISI, 197(1961), 199.
- 16) E. J. Efsic and C. M. Wayman: Trans. AIME, 239(1967), 873.
- 17) T. Tadaki and K. Shimizu: Trans. JIM, 11(1970), 44.
- 18) I. Tamura: Trans. ISIJ, 6(1966), 249.
- 19) J. S. Bowles and J. K. Mackenzie: Acta Met., 2(1954), 129.
- 20) J. S. Bowles and J. K. Mackenzie: Acta Met., 2(1954), 224.
- 21) J. K. Mackenzie and J. S. Bowles: Acta Met., 2(1954), 138.
- 22) J. K. Mackenzie and J. S. Bowles: Acta Met., 5(1957), 137.
- 23) M. S. Wechsler, D. S. Lieberman and T. A. Read: Trans. AIME, 197(1953), 1503.
- 24) M. S. Wechsler and H. M. Otte: Acta Met., 7(1959), 793.
- 25) E. C. Bain: Trans. AIME, 70(1924), 25.
- 26) J. F. Breedis and C. M. Wayman: Trans. AIME, 224(1962), 1128.

- 27) E. Scheil: Z. anorg. alleg. chem., 207(1932), 21.
- 28) J. R. Patel and M. Cohen: Acta Met., 1(1953), 531.
- 29) A. W. McReynolds: J. Appl. Phys., 20(1949), 896.
- 30) L. F. Potter and R. C. Rosenthal: Acta Met., 7(1959), 504.
- 31) T. Hirayama and M. Kogirima: J. Japan Inst. Metals, 34(1970), 507.
- 32) M. G. Wells and D. R. F. West: JISI, 200(1962), 710.
- 33) I. Hagiwara and S. Kanazawa: J. Japan Inst. Metals, 23(1959), 735;
25(1961), 213.
- 34) Y. Hosoi and Y. Kawakami: J. ISIJ, 49(1963), 1780.
- 35) G. A. Stone and G. Thomas: Met. Trans., 5(1974), 2095.
- 36) M. Ahlers: Scripta Met., 10(1976), 989.
- 37) Y. Higo, F. Lecroisey and T. Mori: Acta Met., 22(1974), 313.
- 38) M. Kato and T. Mori: Acta Met., 24(1976), 853.
- 39) M. Kato, R. Monzen and T. Mori: Acta Met., 26(1978), 605.
- 40) G. H. Olsen and W. A. Jesser: Acta Met., 19(1971), 1009.
- 41) G. H. Olsen and W. A. Jesser: Acta Met., 19(1971), 1299.
- 42) G. B. Olson and M. Cohen: J. Less-Common Metals, 28(1972), 107.
- 43) G. F. Bolling and R. H. Richman: Scripta Met., 4(1970), 539.
- 44) G. F. Bolling and R. H. Richman: Phil. Mag., 19(1969), 247.
- 45) G. F. Bolling and R. H. Richman: Acta Met., 18(1970), 673.
- 46) R. H. Richman and G. F. Bolling: Met. Trans., 2(1971), 2451.
- 47) M. Tokizane: Scripta Met., 10(1976), 459.
- 48) G. B. Olson and M. Cohen: Met. Trans., 6(1975), 791.
- 49) J. A. Venables: Phil. Mag., 7(1962), 35.
- 50) F. Lecroisey and A. Pineau: Met. Trans., 3(1973), 387.
- 51) P. C. Maxwell, A. Goldberg and J. C. Shyne: Met. Trans., 5(1974),
1305.
- 52) P. C. Maxwell, A. Goldberg and J. C. Shyne: Met. Trans., 5(1974),
1319.
- 53) T. Suzuki, H. Kojima, K. Suzuki, T. Hashimoto and M. Ichihara:
Acta Met., 25(1977), 1162.
- 54) S. Ueda and H. Fijita: J. Japan Inst. Metals, 41(1977), 1007.
- 55) J. R. C. Guimarães and J. C. Shyne: Scripta Met., 4(1970), 1019.
- 56) J. R. C. Guimarães and J. C. Shyne: Met. Trans., 2(1971), 2063.
- 57) D. R. Wallman and J. R. C. Guimâraes: Scripta Met., 7(1973), 355.
- 58) T. G. Gooch and D. R. F. West: J. Iron Steel Inst., 205(1967), 555.

- 59) I. Tamura, T. Maki, M. Nakanishi and H. Oka: Suiyokaishi, 17(1970), 75.
- 60) I. Ya. Georgiyeva, G. V. Kurdjumov, O. P. Maksimova and V. V. Nemirovsky: Fiz. Met. Metalloved, 23(1967), 23.
- 61) J. F. Breedis: Acta Met., 13(1965), 239.

CHAPTER II THE EFFECT OF STRESS ON THE DEFORMATION-INDUCED MARTENSITIC TRANSFORMATION

2.1 Introduction

The effect of applied stress on the martensitic transformation is studied by revealing the relation between the mechanical driving force given by the applied stress and the amount of martensite induced by deformation. As described in Chapter I, an applied stress is considered to aid the total shape strain accompanying the transformation. It can also be assumed that the mechanical driving force given by applied stress has the same contribution on the martensitic transformation as the chemical driving force. If an applied stress aids the total shape strain involving the volume change during the transformation, the mechanical driving force is given by Eq.1.1(in Chapter I).

$$W_m = \tau\gamma_0 + \sigma\epsilon_0 \quad \text{-----(1.1)}$$

The mechanical driving force depends on both the sign of applied stress(σ is positive for tension and negative for compression) and the volume change(ϵ_0 is positive for dilatation and negative for contraction). In present study, the difference between the tensile and compressive stress in the contribution on the transformation was studied using two kinds of polycrystalline metastable austenitic alloys with different volume changes during the transformation. One was an Fe-29%Ni-0.2%C alloy which formed α' -martensite(bcc) with dilatation of about 4%, and the other was an Fe-32%Mn-0.2%C alloy which formed ϵ' -martensite(hcp) with contraction of about 2%.

First, a model of transformation kinetics (the variation of amount of martensite as a function of deformation parameter) as a function of stress was proposed, and then the amount of martensite calculated from the model was compared with that measured experimentally.

The variation of amount of martensite with the degree of deformation is important for practical application of TRIP phenomenon, and many studies on this subject have been performed. However, in former studies, the transformation kinetics was treated as a function of strain. Angel⁽¹⁾ studied the deformation-induced martensitic transformation using 18-8 stainless steel, and showed that the amount of martensite as a function of strain at constant temperature is sigmoidal in shape and proposed the following equation.

$$V_m = A \epsilon^B V_\gamma \quad \text{-----} (2.1)$$

where $B \approx 3$

V_m : amount of martensite

V_γ : amount of austenite $V_m + V_\gamma = 1$.

Using Eq.2.1, Ludwigson et al⁽²⁾ analyzed the stress-strain curves of metastable austenitic steels in which martensitic transformation took place during deformation. Gerberich et al⁽³⁾ derived the following parabolic relation from their data obtained on TRIP steels.

$$V_m = A' \epsilon^{1/2} \quad \text{-----} (2.2)$$

where A' is constant. Olson and Cohen⁽⁴⁾ proposed a model for the mechanism of strain-induced martensitic transformation and the following relation as for the transformation kinetics.

$$V_m = 1 - \exp[-\beta\{1 - \exp(-\alpha\epsilon)\}^n] \quad \text{-----} (2.3)$$

where α , β and n are constants. Guimarães⁽⁵⁾ simplified the Eq.2.3 as

$$V_m = 1 - \exp(-k\epsilon^z) \text{ -----(2.4)}$$

where k and z are constants. They reported that Eq.2.4 fitted well with the data obtained in Fe-Ni-C alloys when $k=28$ and $z=3.7$.

In all the former studies, the transformation kinetics has been treated as a function of strain. However, the transformation is promoted by the applied stress. The theoretical treatment of the transformation kinetics should be based based on the applied stress instead of strain. In this study, the transformation kinetics are treated theoretically as a function of applied stress.

2.2 Experimental Procedure

2.2.1 Specimen preparations

Alloys used in this study were prepared by Vacuum induction melting. The chemical compositions and M_s temperatures are listed in Table 2.1. The ingot of alloy 1 and alloy 3 were hot-forged to bars of 10 mm diameter and 150 mm length, and homogenized in a vacuum furnace at 1473K for 24h and were drawn to bars of 6 mm diameter.

Specimens as shown in Fig.2.1(a) were machined from the bars. Specimens of the same configuration(Fig.2.1(a)) were used with tensile and compressive tests. The ingot of alloy 2 was hot-forged at 1473K and homogenized in a vacuum furnace at 1473K for 24h, and cold-rolled to make a sheet of 0.8 mm thickness. Specimens for tensile tests as shown in Fig.2.1(b) were machined from the sheet.

Specimens of alloy 1 and alloy 2 were austenitized in a vacuum furnace at 1423K for 3h and for 1h, respectively, and

Table 2.1 Chemical compositions(wt%), M_S and M_d temperatures (K) of alloys.

Alloy	C	Si	Mn	P	S	Ni	Fe	M_S	M_d
1. Fe-29%Ni-0.2%C	0.22	0.009	<0.002	0.002	0.004	28.94	bal.	~ 243	~303
2. Fe-25%Ni-0.7%C	0.65	0.067	0.17	0.010	0.009	25.31	bal.	~ 163	~253
3. Fe-32%Mn-0.2%C	0.18	0.008	31.78	-	0.018	-	bal.	<77	>303

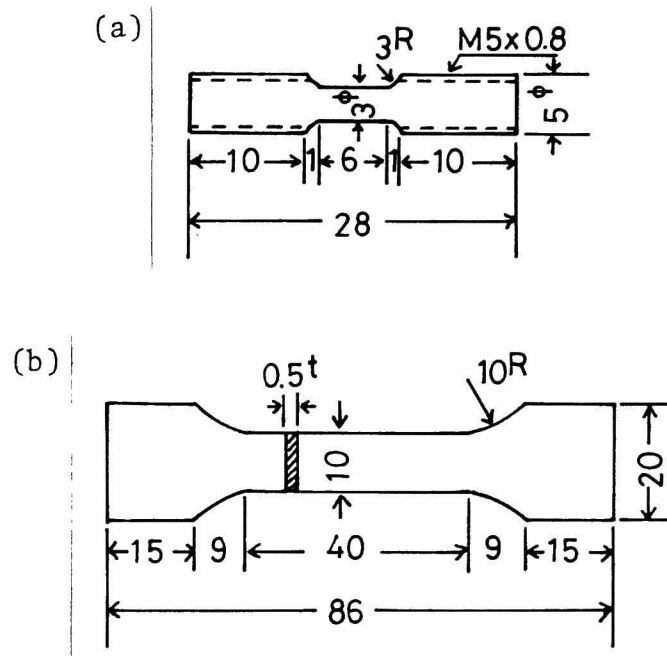


Fig.2.1 Shape and size of specimen(mm).

quenched in iced brine. Specimens of alloy 3 were austenitized in an Ar furnace at 1323K for 2h and quenched in oil.

Crystallographic features of martensite which forms in each alloy used in this alloy are shown in Table 2.2.

Tensile and compressive tests were performed with an Instron type testing machine at various temperatures from 303K to 77K. Temperatures below room temperature were obtained in a cold chamber which was cooled by evaporation of liquid nitrogen. The strain rate was $4 \times 10^{-3} \text{ s}^{-1}$ for the specimen of Fig.2.1(a) and $3 \times 10^{-4} \text{ s}^{-1}$ for the specimen of Fig.2.1(b), respectively.

2.2.2 Measurement of amount of martensite

Measurement of amount of α' -martensite(Fe-Ni-C) was performed by lineal analysis, and that of ϵ' -martensite(Fe-Mn-C) was performed by X-ray diffraction⁽¹³⁾ using following equation.

$$V_m(\text{amount of martensite}) = 1/(1 + GS) \quad \text{----- (2.5)}$$

where $S = I_{(111)_\gamma} / I_{(10.1)_\epsilon}$

$I_{(111)_\gamma}$: Integrated intensity of $(111)_\gamma$ diffraction line

$I_{(10.1)_\epsilon}$: Integrated intensity of $(10.1)_\epsilon$ diffraction line

$G=0.24$. G was calculated from the following equation.

$$G = \frac{[V_\gamma^2 F_\epsilon^2 P_\epsilon (1 + \cos^2 2\theta_\epsilon) / \sin^2 \theta_\epsilon \cos \theta_\epsilon]}{[V_\epsilon^2 F_\gamma^2 P_\gamma (1 + \cos^2 2\theta_\gamma) / \sin^2 \theta_\gamma \cos \theta_\gamma]} \quad \text{----- (2.6)}$$

where temperature dependence of intensity is neglected, where V is volume of unit cell ($V_\gamma/V_\epsilon=1.02$), and F is structure factor ($|F|_\gamma^2=16f^2$ and $|F|_\epsilon^2=3f^2$, f is atomic scattering factor), and P is multiplicity factor ($P_\gamma=8$ and $P_\epsilon=12$), and θ is Bragg angle ($\theta_{(111)_\gamma}=29.9^\circ$, $\theta_{(10.1)_\epsilon}=27.5^\circ$).

Table 2.2 Crystallographic features of martensite.

Alloy	Crystal structure of martensite	Orientation relationship	Habit plane	Volume change
Fe-29%Ni-0.2%C Fe-25%Ni-0.7%C	α' : bcc	G-T relationship ⁽⁶⁾ $(111)_\gamma \sim 1^\circ (011)_\alpha$ $[101]_\gamma \sim 2.5^\circ [\bar{1}\bar{1}1]_\alpha$	$\{3\ 15\ 10\}_\gamma^{(7),(8)}$	expansion of 4% ⁽⁹⁾
Fe-32%Mn-0.2%C	ϵ' : hcp	S-N relationship ⁽¹⁰⁾ , ⁽¹¹⁾ $(111)_\gamma // (0001)_\epsilon$, $[11\bar{2}]_\gamma // [\bar{1}\bar{1}00]_\epsilon$ or $[\bar{1}\bar{1}0]_\gamma // [\bar{1}\bar{1}\bar{2}0]_\epsilon$	$\{111\}_\gamma^{(10),(11)}$	contraction of 2% ⁽¹²⁾

In present study, radiation was supplied with Fe target(Mn filter) powered at 30 kV and 10 mA.

2.3 Results and Discussion

2.3.1 Model of transformation kinetics

Fig.2.2 shows schematically the free energies of martensite and austenite as a function of temperature. T_0 is the temperature at which austenite and martensite are in equilibrium. When austenite is supercooled to M_s temperature at which the chemical driving force is W_C^0 , martensitic transformation starts.

Variation of amount of martensite as a function of applied stress is now considered when austenite is deformed at T_1 (above the M_s) in Fig.2.2.

When martensitic transformation occurs under applied stress, the mechanical work (W_m) is done on or by the transformation due to the action of applied stress. The mechanical work was expressed by Patel and Cohen⁽⁹⁾ as

$$W_m = \tau \gamma_0 + \sigma \epsilon_0 \quad \text{-----(1.1)}$$

where τ is the shear stress resolved along a habit plane, and γ_0 is the transformation shear strain along the habit plane, and σ is the normal stress resolved perpendicular to the habit plane, and ϵ_0 is the normal component of the transformation strain. For any given orientation of martensite plate, Eq.1.1 is expressed as

$$W_m(\sigma', \theta, \alpha) = (1/2)\sigma' [\gamma_0 \sin 2\theta \cos \alpha \pm \epsilon_0 (1 + \cos 2\theta)] \quad \text{-----(2.7)}$$

where σ' is absolute value of the applied stress and θ is an angle between the applied stress axis and the normal direction to the habit plane and α is an angle between the shear direction of the

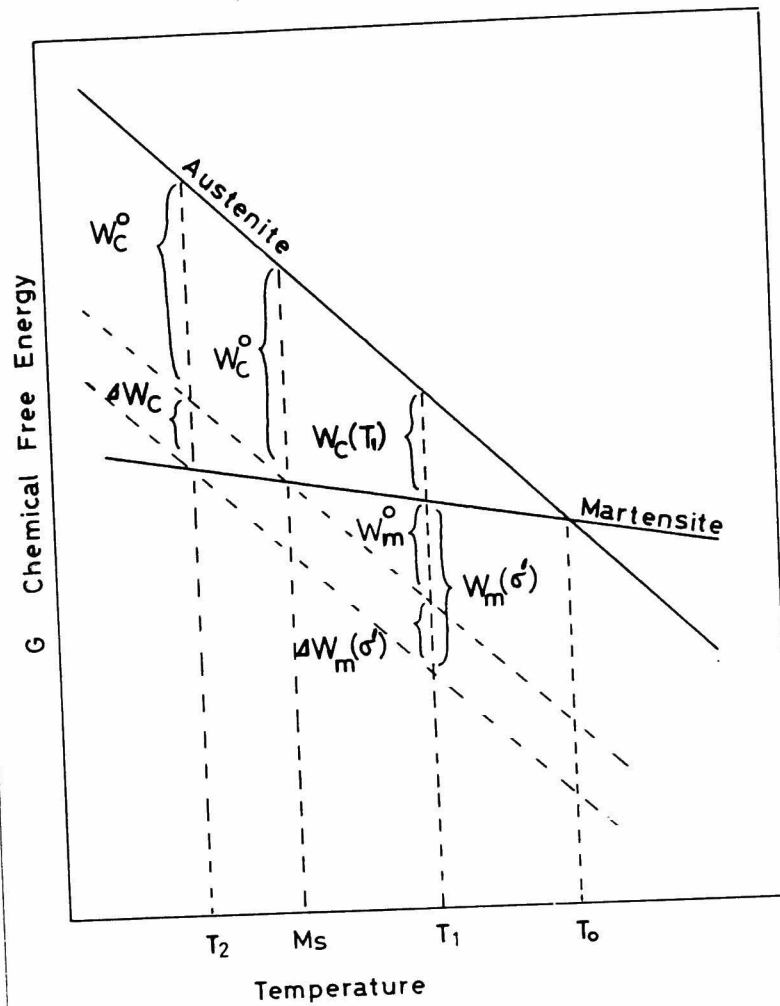


Fig.2.2 A schematic relationship between free energy of martensite and austenite as a function of temperature.

the transformation and the maximum shear direction of the applied stress on the habit plane, and + and - correspond to tension and compression, respectively. Fig.2.3 shows the schematic illustration of the relation among the applied stress axis, θ and α . It is assumed that the mechanical driving force contributes to the transformation equivalently with the chemical driving force.

When stress is applied to austenite at $T_1(>M_s)$ in Fig.2.2, the mechanical driving force $W_m(\sigma')$ as expressed in Eq.2.7 is added to the chemical driving force at $T_1(W_c(T_1))$, and martensitic transformation starts at the critical stress(σ'_0) where the total driving force $\{W_c(T_1) + W_m^0(\sigma'_0)\}$ is equal to W_c^0 . As the mechanical driving force increases from W_m^0 by increase in σ' , the amount of martensite increases. It is then assumed that the amount of martensite(V_m) will be given as a function of the increment of mechanical driving force from W_m^0 (i.e., $\Delta W_m(\sigma') = W_m(\sigma') - W_m^0$) as

$$V_m = f(\Delta W_m(\sigma')) \text{ ----- (2.8)}$$

It is assumed that martensites with various orientations are formed in a polycrystalline austenite since orientation of each grain is distributed at random. The value of mechanical driving force depends on the orientation of martensite. Then the total amount of martensite in the polycrystalline specimen will be obtained by averaging Eq.2.8 for θ and α . It is assumed that θ varies from 0 to $\pi/2$, and Eq.2.8 is averaged about θ from 0 to $\pi/2$.

$$V_m(\sigma', \bar{\theta}, \alpha) = (2/\pi) \int_0^{\pi/2} f(\Delta W_m(\sigma', \theta, \alpha)) d\theta \text{ --- (2.9)}$$

The total $\bar{V}_m$ will be obtained by averaging Eq.2.9 for α in the range where $V_m(\sigma', \bar{\theta}, \alpha)$ is larger than 0. $V_m(\sigma', \bar{\theta}, \alpha)$ can be

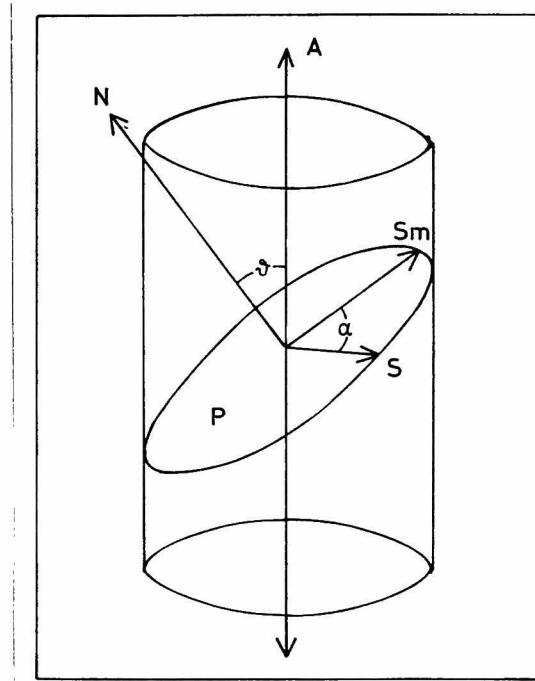


Fig.2.3 Relation among the applied stress axis A , θ and α .
 N is normal direction to the habit plane P , and
 S is shear direction of the transformation and
 Sm is maximum shear direction on P .

shown qualitatively as in Fig.2.4 as a function of α . $V_m(\sigma', \bar{\theta}, \alpha)$ is positive in the range $0 < \alpha < \alpha_0$, and the total $\bar{V}_m$ will be obtained as

$$\bar{V}_m = (1/\alpha_0) \int_0^{\alpha_0} (2/\pi) \int_0^{\pi/2} f(\Delta W_m(\sigma', \theta, \alpha)) d\theta d\alpha \quad \text{-----} (2.10)$$

where $V_m(\sigma', \bar{\theta}, \alpha_0) = 0$.

In this study, the function (f) was obtained from the following cooling experiment. When austenite is cooled below the M_s , the amount of martensite increases with increasing chemical driving force. It is assumed that the amount of martensite V_m at $T_2 (< M_s)$ in Fig.2.2 will be given by the equation(2.11) in which V_m is expressed as a function of the increment of chemical driving force from W_C^0 (i.e., $\Delta W_C(T_2) = W_C(T_2) - W_C^0$).

$$V_m = g(\Delta W_C(T_2)) \quad \text{-----} (2.11)$$

The Eq.2.11 for each alloy was obtained by measuring the amount of martensite as a function of temperature difference ($\Delta T = M_s - T$) and the function (f) was assumed to be nearly equal to the function (g).

2.3.2 Effect of volume change on martensitic transformation

The difference between tensile and compressive stress in the contribution to the transformation was studied using two kinds of polycrystalline metastable austenitic alloys with different volume changes during the transformation. One was an Fe-29%Ni-0.2%C alloy (which forms bcc α' -martensite), and the other was an Fe-32%Mn-0.2% alloy (which forms hcp ϵ' -martensite).

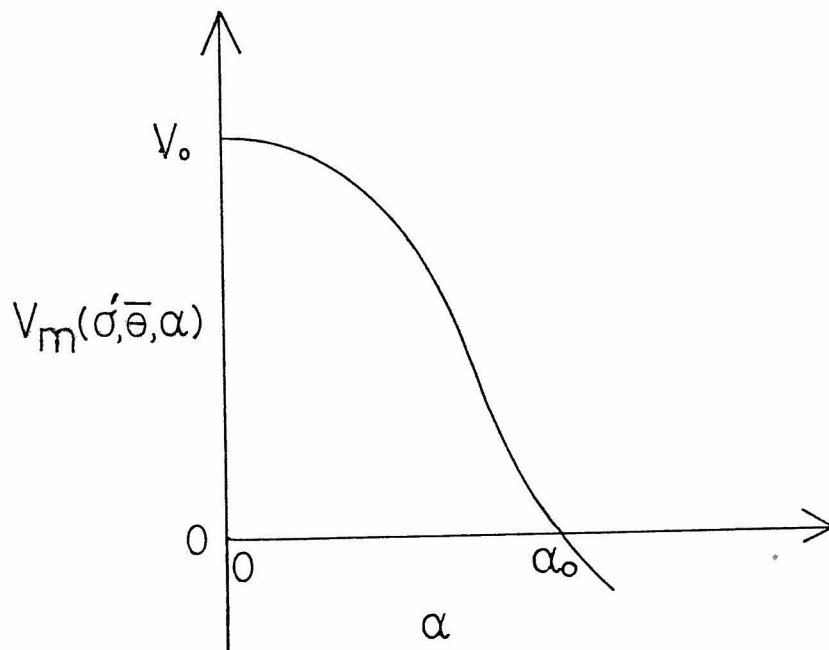


Fig.2.4 A schematic illustration of $V_m(\sigma', \bar{\theta}, \alpha)$ as a function of α . $V_0 = V_m(\sigma', \bar{\theta}, 0)$.

2.3.2.1 Fe-29%Ni-0.2%C (expansion with $\gamma \rightarrow \alpha'$)

2.3.2.1.a Stress-strain curve

Fig.2.5 shows stress-strain curves of the Fe-29%Ni-0.2%C alloy in tension and compression at three different temperatures. Martensite was not induced up to 15% strain in both tension and compression at 303K. A remarkable difference between tension and compression is not seen at 303K. At 247K, just above the M_s temperature, the flow stress is much lower in tension than in compression over the initial portion of stress-strain curve, and reversed after 10% strain. The difference in initial portion arose since martensitic transformation occurred in tension before austenite yielded. The latter behavior can be attributed to the larger amount of martensite formed under tension than compression(as will be described in section 2.3.2.1.c), therefore resulting in a higher strain-hardening in tension than in compression. At 193K, the same tendency is observed due to the transformation of retained austenite.

2.3.2.1.b Microstructure

Photo.2.1 shows a microstructure at 23% strain in tension at 247K. Almost all martensites are lenticular martensites. Photo.2.2 shows a microstructure at 43% strain in compression at 247K. The morphology of martensite in compression is quite different from that in tension. Very fine martensites and banded structure are observed. It seems that the defects introduced by deformation inhibit martensite to grow. It is considered that martensite formed in

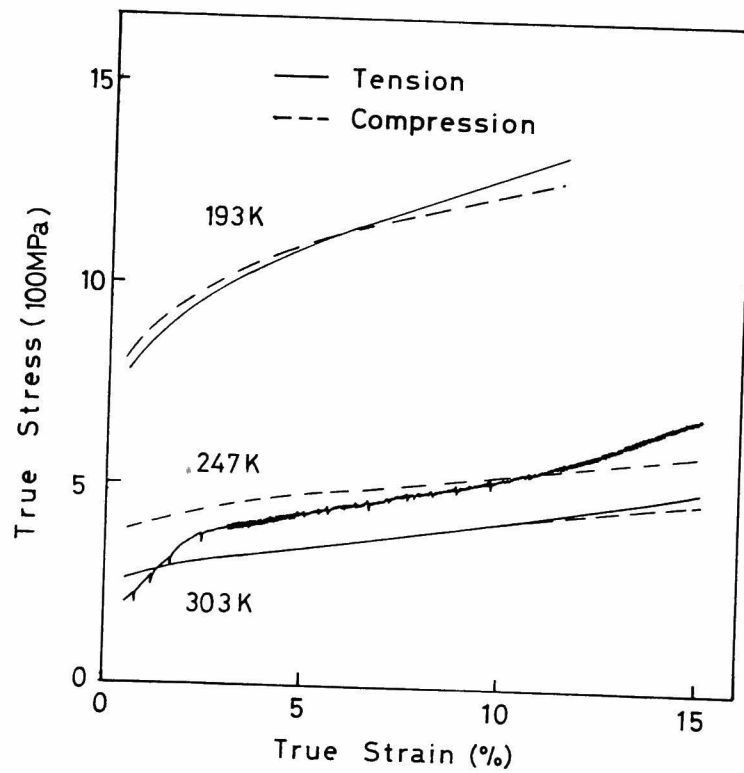


Fig.2.5 Stress-strain curves of the Fe-29%Ni-0.2%C alloy at three different temperatures in tension and compression.



Photo.2.1 Optical microstructure of the Fe-29%Ni-0.2%C alloy strained 23% in tension at 247K.
Etched by 35% $\text{Na}_2\text{S}_2\text{O}_5$. (x 700)

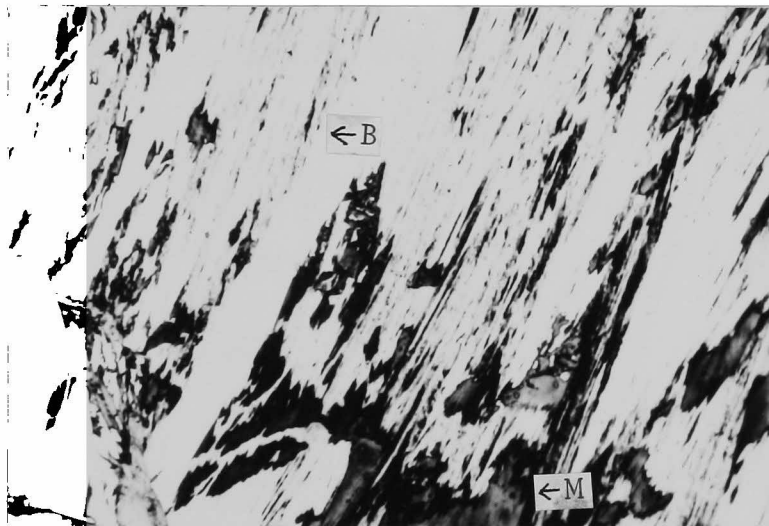


Photo.2.2 Optical microstructure of the Fe-29%Ni-0.2%C alloy strained 43% in compression at 247K. Martensite(M) and banded structure(B) are observed. Etched by 35% $\text{Na}_2\text{S}_2\text{O}_5$. (x 700)

compression is finer than in tension because martensitic transformation starts at higher strain (higher density of defects) in compression than in tension. The banded structure which is observed in Photo.2.2 has been called strain-induced martensite by Maxwell et al⁽¹⁴⁾. However, in present study, it is made clear that the structure is not martensite but deformation twins of austenite. The detail will be discussed in Chapter III.

2.3.2.1.c Amount of martensite formed during deformation

Amount of martensite which formed during deformation was calculated from Eq.2.10 as a function of flow stress and was compared with the amount which was measured experimentally.

For the Fe-29%Ni-0.2%C alloy, increase in chemical driving force ($\Delta W_C(T)$) as a function of temperature difference ($\Delta T = M_S - T$) can be approximated by next equation which was shown by Patel and Cohen⁽⁹⁾.

$$\Delta W_C(T) = 5.56 \Delta T \quad (\text{J/mol}) \quad \text{-----} (2.12)$$

The amount of thermally transformed martensite by cooling below the M_S was measured experimentally as a function of ΔT . The results are shown in Fig.2.6. From Fig.2.6, the relation between the amount of martensite and ΔT is approximated as follows for the Fe-29%Ni-0.2%C alloy.

$$V_m = 1.70 \Delta T \quad (\%) \quad \text{-----} (2.13)$$

From Eq.2.12 and Eq.2.13, the following equation is derived.

$$V_m = (1.70/5.56) \Delta W_C(T) \quad \text{-----} (2.14)$$

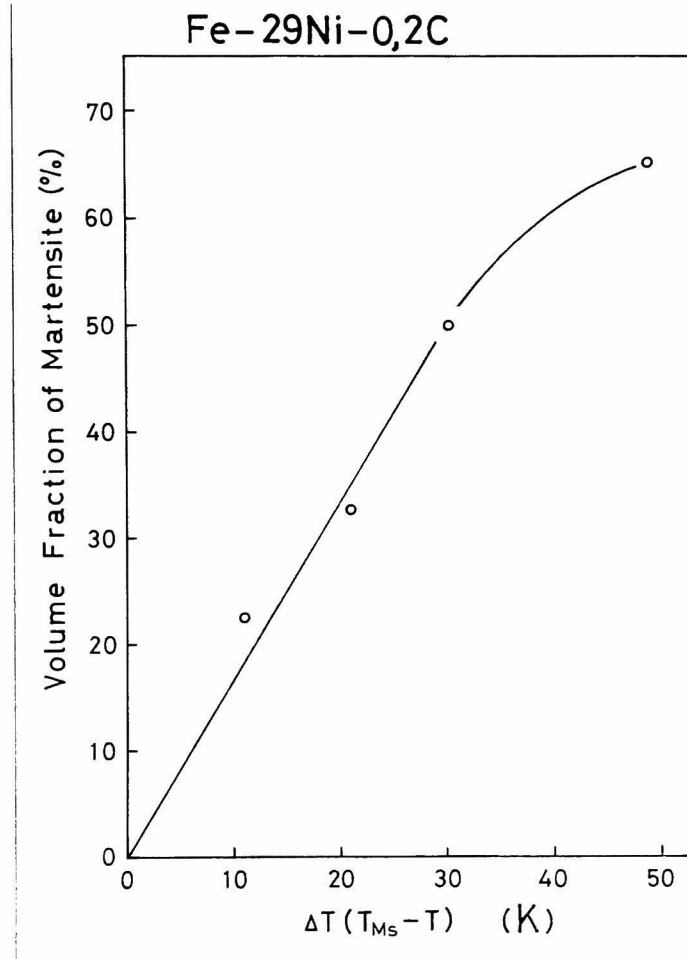


Fig.2.6 Amount of martensite formed by cooling below M_S as a function of temperature difference $\Delta T = M_S - T$.

Assuming that the increment of mechanical driving force ($\Delta W_m(\sigma', \theta, \alpha) = W_m(\sigma', \theta, \alpha) - W_m^0$) contributes to the transformation as same as $\Delta W_C(T)$, ΔW_C is displaced by $\Delta W_m(\sigma', \theta, \alpha)$ as

$$V_m(\sigma', \theta, \alpha) = k(1.70/5.56)\Delta W_m(\sigma', \theta, \alpha) \quad \text{-----(2.15)}$$

where $k=7.08$. k is necessary to convert the units of $\Delta W_m(\sigma', \theta, \alpha)$ (stress times transformation strain per unit volume of austenite : Nm/m^3) to the units of $\Delta W_C(T)$ (J/mol).

The total amount of martensite $\overline{V}_m(\sigma')$ will be obtained by averaging Eq.2.15 for θ and α .

$$\overline{V}_m(\sigma') = k(1/\alpha_0) \int_0^{\alpha_0} (2/\pi) \int_0^{\pi/2} (1.70/5.56) [(\sigma'/2)\{\gamma_0 \sin 2\theta \cos \alpha \pm \epsilon_0 (1 + \cos 2\theta)\} - W_m^0] d\theta d\alpha \quad \text{-----(2.16)}$$

where $\gamma_0=0.20^{(9)}$ and $\epsilon_0=0.04^{(9)}$, and + and - correspond to tension and compression, respectively, and α_0 is the upper limit of α , and W_m^0 is the minimum mechanical driving force which is necessary to induce martensitic transformation.

The minimum mechanical driving force (W_m^0) was decided from the stress (σ_0) where martensitic transformation started in tension. When martensitic transformation starts, the particular martensite of which orientation yields a maximum value of $W_m(\sigma_0, \theta, \alpha)$. Then, α is equal to 0, and

$$dW_m(\sigma_0, \theta, 0)/d\theta = \sigma_0 \gamma_0 \cos 2\theta + \sigma_0 \epsilon_0 (-\sin 2\theta) = 0 \quad \text{---(2.17)}$$

$$\sin 2\theta / \cos 2\theta = \tan 2\theta = (\gamma_0 / \epsilon_0) \quad \text{-----(2.18)}$$

Substituting the known components of the transformation strain ($\gamma_0=0.20$, $\epsilon_0=0.04$) into Eq.2.18 and then into Eq.2.7, it is possible to obtain W_m^0 using $\sigma_0=314(\text{MPa})$.

$$\begin{aligned}
W_m^o &= k(314/2)\{(0.20)\sin(1.374) + (0.04)(1 + \cos(1.374))\} \\
&= 7.08 \times 38.2 \\
&= 271 \text{ (J/mol)} \quad \text{-----(2.19)}
\end{aligned}$$

The upper limit α_0 was decided for each stress σ' as follows from the condition $V_m(\sigma', \bar{\theta}, \alpha) \geq 0$.

$$\begin{aligned}
V_m(\sigma', \bar{\theta}, \alpha) &= k(1.70/5.56)(2/\pi) \int_0^{\pi/2} [(\sigma'/2)\{\gamma_0 \sin 2\theta \cos \alpha \pm \epsilon_0(1 + \cos 2\theta)\} - W_m^o] d\theta \\
&= k(1.70/5.56) \{(\sigma' \gamma_0 / \pi) \cos \alpha \pm (\sigma' \epsilon_0 / 2) - W_m^o\} \quad \text{---(2.20)}
\end{aligned}$$

Equation(2.20) decreases as α increases from 0. Then, α_0 is decided from the following equation.

$$(\sigma' \gamma_0 / \pi) \cos \alpha \pm (\sigma' \epsilon_0 / 2) - W_m^o = 0 \quad \text{-----(2.21)}$$

where + and - correspond to tension and compression, respectively.

Equation(2.16) is reduced to the next equation.

$$V_m(\sigma') = (A\sigma' / \alpha_0) \sin \alpha_0 \pm B\sigma' - C \quad \text{-----(2.22)}$$

where $A=1.38 \times 10^{-1}$, $B=4.33 \times 10^{-2}$, $C=82.7$, and + and - correspond to tension and compression, respectively.

Fig.2.7 shows experimental data in this study at 247K and the result which calculated from Eq.2.22. It can be seen from Fig.2.7 that the amount of martensite induced during deformation is much larger in tension than in compression and excellent agreements are obtained in both tension and compression between the calculated value and experimental data. The difference of martensitic transformation kinetics in the Fe-29%Ni-0.2%C alloy between in tension and compression is considered to be caused by the term ϵ_0 . From this result, it can be concluded that the contribution of the applied stress on the martensitic transformation is expressed by

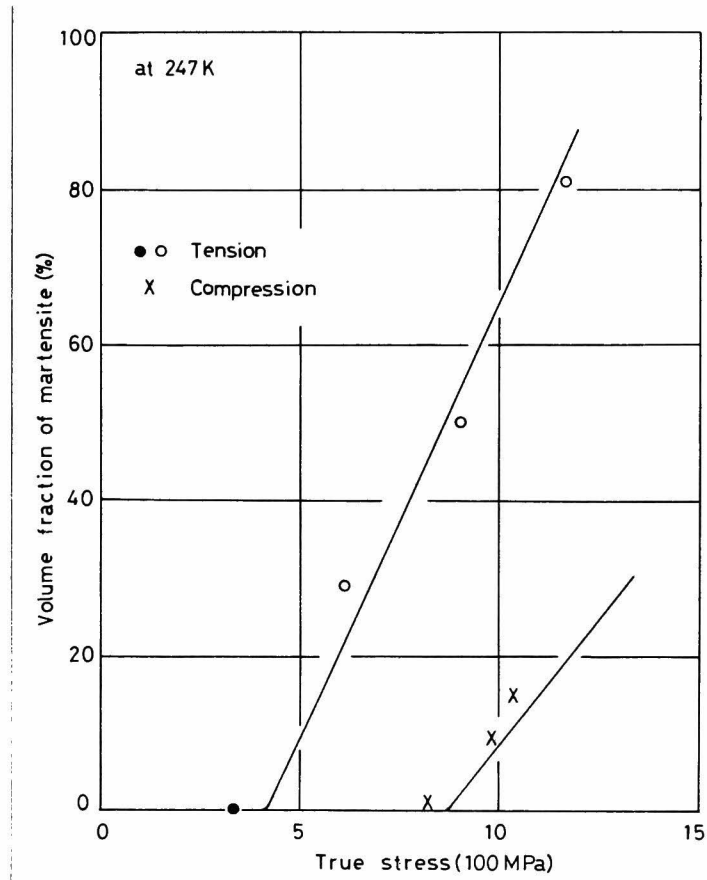


Fig.2.7 Amount of martensite induced during deformation as a function of applied stress in the Fe-29%Ni-0.2%C alloy. The calculated value is shown by full line and o, ● and x show experimental data. ● shows the datum used to determine W_m^o .

the mechanical driving force(Eq.2.7). This means that the applied stress aids the total shape strain of martensitic transformation involving the volume change.

2.3.2.2 Fe-32%Mn-0.2%C (contraction with $\gamma \rightarrow \epsilon'$)

2.3.2.2.a Stress-strain curve

Fig.2.8 shows stress-strain curves of the Fe-32%Mn-0.2%C alloy in tension and compression at 77K and 283K. The M_s temperature of this alloy is below 77K, and the M_d temperature is above 283K. At 283K, there is not a remarkable difference between tension and compression below 10% strain. Above 10% strain, the flow stress is a little higher in tension than in compression. At 283K, ϵ' -martensite is not induced in the strain range of Fig.2.8, so this difference seems to be the effect of barreling in compression^{(15),(16)}. At 77K, ϵ' -martensite is induced by deformation in the early stage of deformation, and stress-strain curve has the same tendency as at 283K. It seems that the amount of ϵ' -martensite induced is small, and the volume change in fcc $\rightarrow$ hcp transformation is small(2%), the amount of martensite induced is not quite different between tension and compression.

2.3.2.2.b Microstructure

Photo.2.3 shows a microstructure at 30% strain in tension at 77K. Photo.2.4 shows a microstructure at 29% strain in compression at 77K. ϵ' -martensite is observed in both cases, and there is no

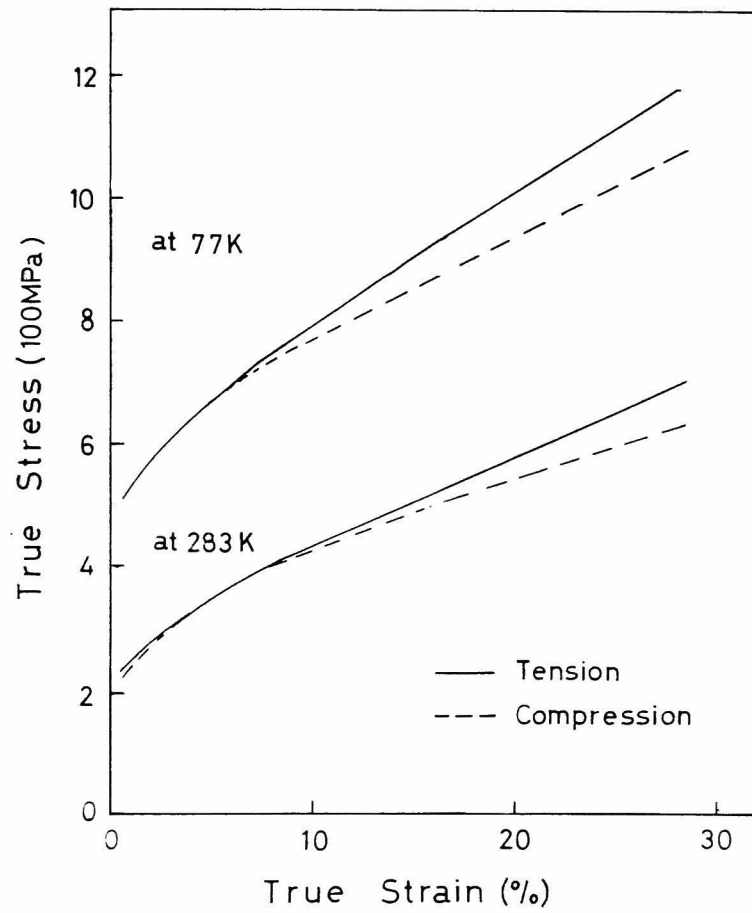


Fig.2.8 Stress-strain curves of the Fe-32%Mn-0.2%C alloy at 77K and 283K in tension and compression.



Photo.2.3 Optical microstructure of the Fe-32%Mn-0.2%C alloy strained 30% in tension at 77K.
Etched by 5% Nital. (x 350)

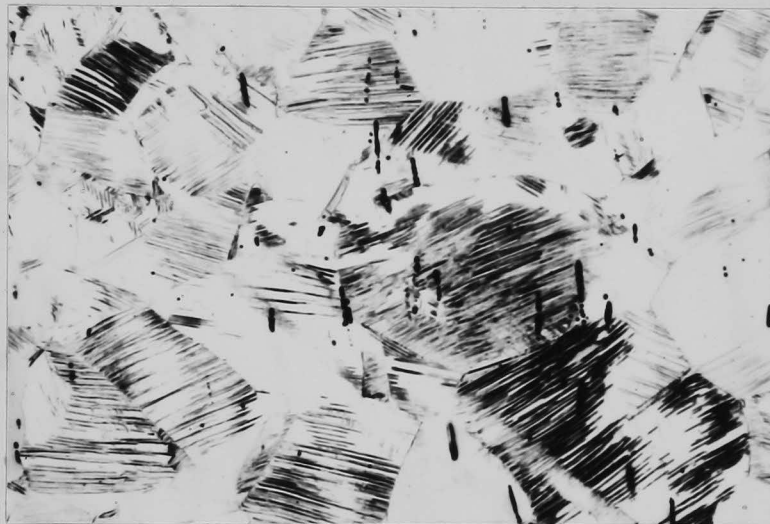


Photo.2.4 Optical microstructure of the Fe-32%Mn-0.2%C alloy strained 29% in compression at 77K.
Etched by 5% Nital. (x 350)

essential difference of morphology of martensite between tension and compression.

2.3.2.2.c Amount of martensite formed during deformation

Although the relation between V_m and $\Delta W_C(T)$ could not be determined in present study since the M_s temperature of the Fe-32%Ni-0.2%C alloy was below 77K, the relation was estimated from the data reported by Fujita and Uchiyama⁽¹⁷⁾ for several Fe-Mn alloys. Fig.2.9 shows the amount of martensite as a function of chemical driving force reported by Fujita and Uchiyama. From Fig.2.9, it is considered that the transformation curve consists of two stages. The amount of martensite first increases drastically and then increases slowly with increasing chemical driving force. As the manganese content increases, the amount of martensite which formed in the first stage decreases. The extrapolated amount of martensite for the Fe-32%Mn-0.2%C alloy is zero for the first stage. The relation between amount of martensite and chemical driving force in the second stage is almost the same for three alloys, and it is assumed that the same relation can be applied to the Fe-32%Mn-0.2%C alloy. The relation(that is shown by dotted line in Fig.2.9) is assumed by averaging the slopes of the three Fe-Mn alloys in Fig.2.9, and given as

$$V_m = 4.59 \times 10^{-2} \Delta W_C(T) \quad \text{-----} (2.23)$$

For the Fe-32%Mn-0.2%C alloy, the relation between the amount of martensite($V_m(\sigma')$) and applied stress is derived using Eq.2.23 and the known components of the transformation($\gamma_0=0.353$,

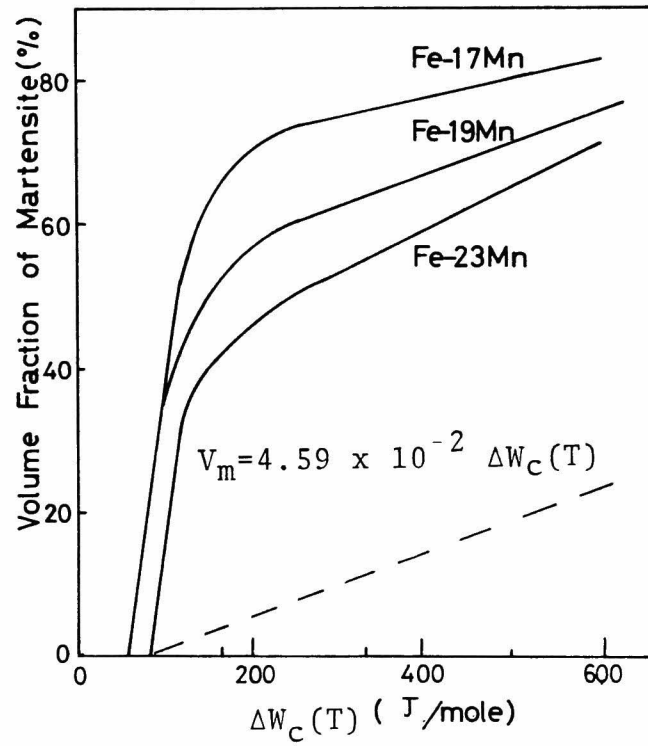


Fig.2.9 Amount of martensite as a function of chemical driving force.

(by Fujita and Uchiyama⁽¹⁷⁾)

$\epsilon_0 = -0.02$)⁽¹²⁾ as

$$V_m(\sigma') = (A\sigma'/\alpha_0)\sin\alpha_0 \pm B\sigma' - C \quad \text{----- (2.24)}$$

where $A = 3.66 \times 10^{-2}$, $B = 3.26 \times 10^{-3}$ and $C = 39.9$, and + and - correspond to compression and tension, respectively.

The upper limit(α_0) for each stress(σ') was determined from the following equation.

$$(\sigma'\gamma_0/\pi)\cos\alpha \pm (\sigma'\epsilon_0/2) - W_m^0 = 0 \quad \text{----- (2.25)}$$

where $W_m^0 = 867$ (J/mol), and + and - correspond to tension and compression, respectively. W_m^0 was determined from the stress ($\sigma_0 = 735$ (MPa)) where the transformation started in tension at 77K.

Fig.2.10 shows experimental data in this study at 77K and that calculated from Eq.2.24. The experimental data show that the amount of martensite is slightly larger in compression than in tension. The calculated value does not fit well with experimental data, but the tendency of increase with the stress is well agreed in both tension and compression. This indicates that the difference of martensitic transformation between tension and compression in the Fe-32%Mn-0.2%C alloy also results from the contribution of the term ϵ_0 . From the results obtained in section 2.3.2.1.c and 2.3.2.2.c, it can be concluded that the contribution of applied stress to the martensitic transformation is expressed by the mechanical driving force(Eq.2.7), and the stress aids the total shape strain involving the volume change during the transformation.

2.3.3 Effect of test temperature

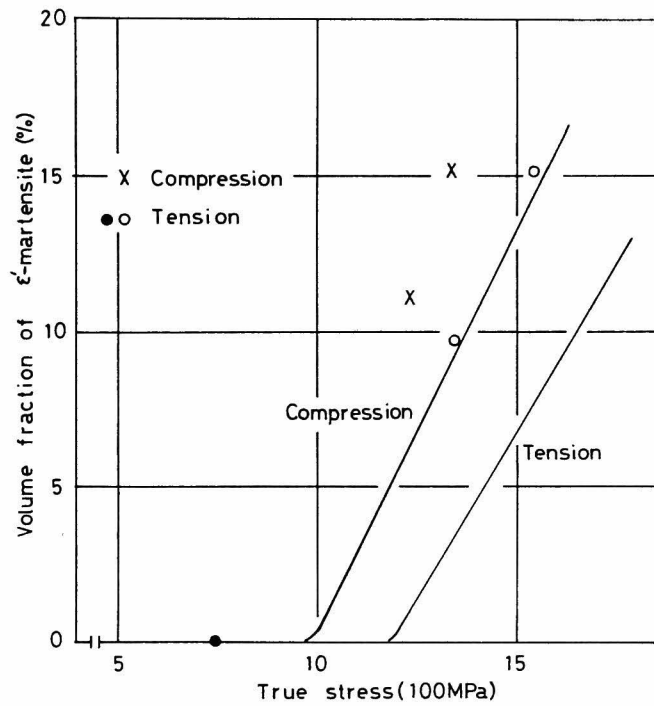


Fig.2.10 Amount of martensite induced during deformation as a function of stress in the Fe-32%Mn-0.2%C alloy at 77K. The calculated value is shown by full line and o, • and x show experimental data. • shows the datum used to determine W_m^0 .

The effect of test temperature on the kinetics of the deformation-induced martensitic transformation was examined using the Fe-25%Ni-0.7%C alloy ($M_S = 163K$). Fig.2.11 shows experimental data at 163K(o), 183K($\square$), 213K($\triangle$) and 223K($\times$) and the expected line which calculated from Eq.2.22. W_m^0 at each temperature was determined from the stress($\bullet$, $\blacksquare$, $\blacktriangle$ and $\times$) where martensitic transformation started at each temperature. The value calculated fits well with experimental data at all temperatures. From this result; it can be concluded that the model proposed in section 2.3.1 can be applied at all temperatures below M_d temperature.

2.3.4 Effect of texture

In present study, specimens were prepared by drawing, and there was a possibility that the results in sections 2.3.2 and 2.3.3 would be subjected the effect of the texture formed by drawing. To confirm such an effect, the following experiment was performed. Cubic specimens were machined from the drawn bars of the Fe-29%Ni-0.2%C alloy as shown in Fig.2.12. The amount of martensite which induced by compression in the direction(a) was compared with that induced by compression in the direction(b).

Fig.2.13 shows the results obtained at 247K with together the data in Fig.2.7. It is noted, there is not a remarkable difference in the amount of martensite between the specimens compressed along the direction (a) and (b). From this result, the effect of texture can be considered as negligible.

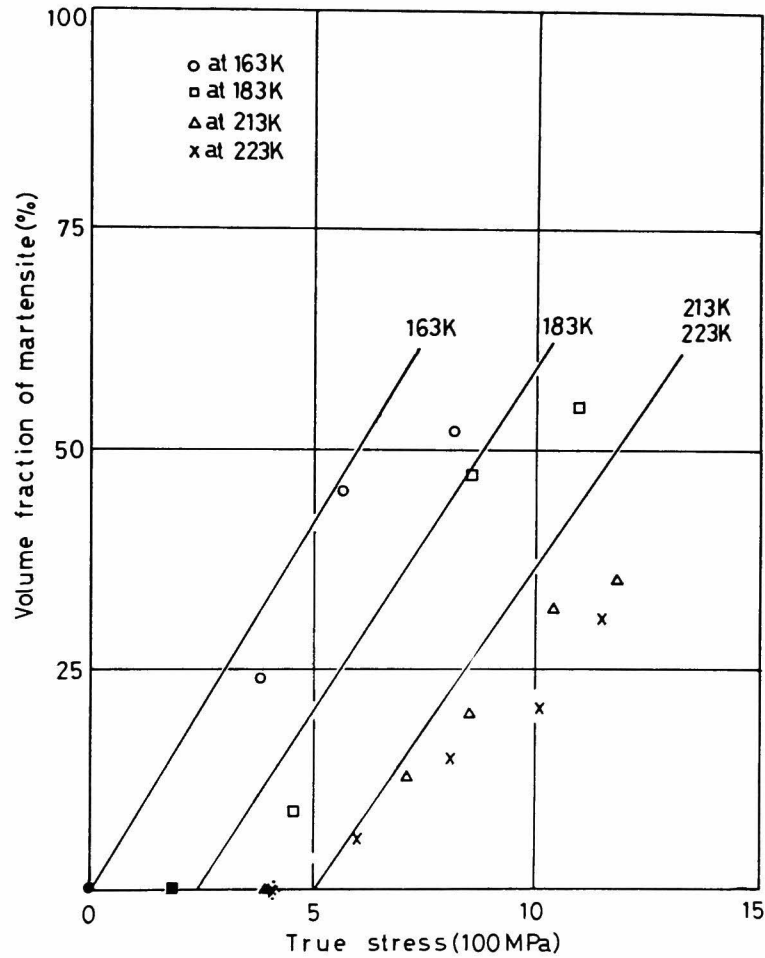


Fig.2.11 Amount of martensite induced during tensile deformation at various temperatures in the Fe-25%Ni-0.7%C alloy. The calculated value is shown by full line and $\circ, \bullet$ (163K), $\square, \blacksquare$ (183K), $\triangle, \blacktriangle$ (213K) and $x, \times$ (223K) show experimental data. $\bullet, \blacksquare, \blacktriangle$ and $\times$ show data used to determine W_m^0 at respective temperatures.

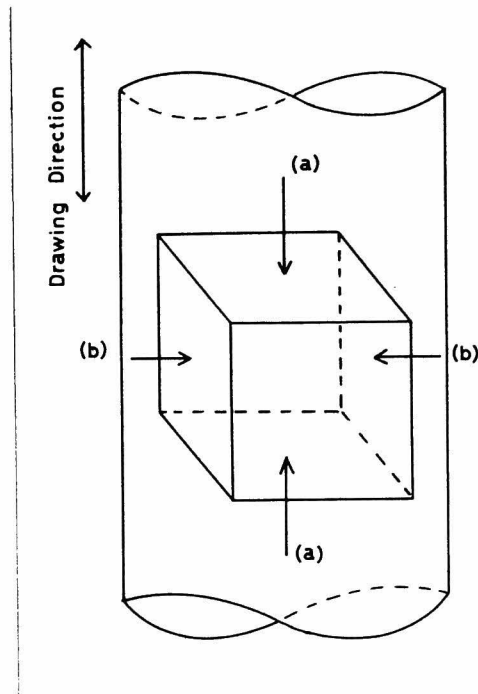


Fig.2.12 Relation between the drawing direction and stressing direction.

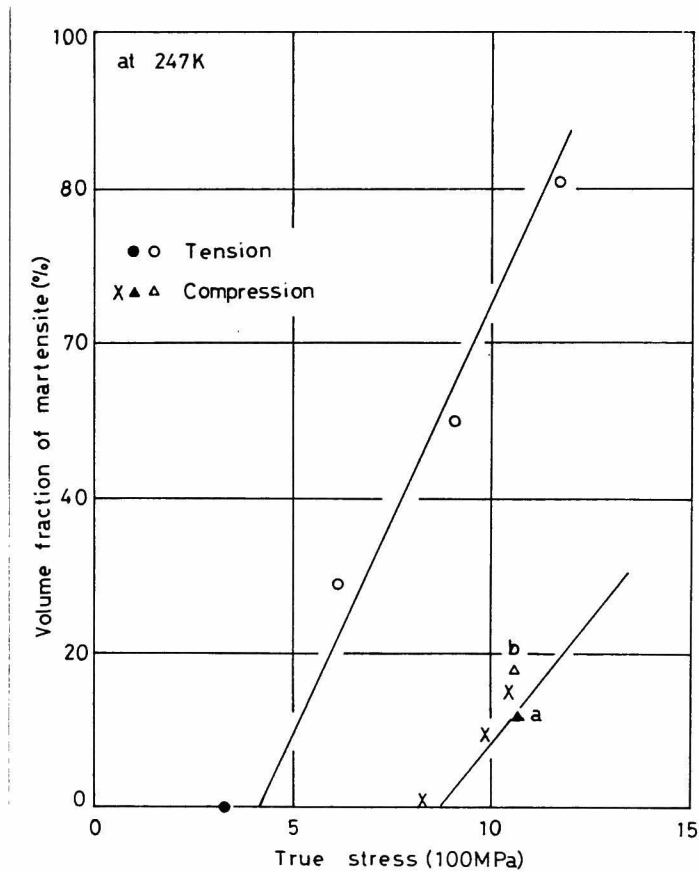


Fig.2.13 Amount of martensite induced by compression in the direction (a)(▲) and (b)(▲)(in Fig.2.12) at 247K in the Fe-29%Ni-0.2%C alloy with together the data in Fig.2.7.

2.4 Conclusion

The effect of applied stress on the martensitic transformation was studied by revealing the relation between the mechanical driving force given by the applied stress and the amount of martensite induced by deformation. The results obtained are as follows.

(1) For the Fe-29%Ni-0.2%C alloy, the amount of martensite induced during deformation was much larger in tension than in compression. For the Fe-32%Mn-0.2%C alloy, the amount of martensite induced during deformation was slightly larger in compression than in tension. Such differences are considered to result from the difference between the tensile and compressive stress in the contribution to the transformation due to the volume change with the transformation.

(2) The amount of martensite which calculated from the proposed model fitted well with the amount which measured experimentally. From this result, the model was verified. It can be concluded that the contribution of applied stress to the martensitic transformation is expressed by the mechanical driving force, and the stress aids the total shape strain involving the volume change during the transformation.

(3) The proposed model can be applied at all temperatures below M_d temperature.

(4) The effect of texture is confirmed to be negligible.

References for Chapter II

- 1) T. Angel: J. Iron Steel Inst., 177(1954), 165.
- 2) D. C. Ludwigson and J. A. Berger: JISI, 207(1969), 63.
- 3) W. W. Gerberich, G. Thomas, E. R. Parker and V. F. Zackay:
Proc. Second Int. Conf. on the Strength of Metals and Alloys,
Asilomer, (1970), 894.
- 4) G. B. Olson and M. Cohen: Met. Trans., 6(1975), 791.
- 5) J. R. C. Guimarães: Scripta Met., 6(1972), 795.
- 6) A. B. Greninger and A. R. Troiano: Trans. AIME, 145(1941), 289;
185(1949), 590.
- 7) E. J. Efsic and C. M. Wayman: Trans. AIME, 239(1967), 873.
- 8) T. Tadaki and Shimizu: Trans. JIM, 11(1970), 44.
- 9) J. R. Patel and M. Cohen: Acta Met., 1(1953), 531.
- 10) G. P. Sanderson and R. W. Honeycombe: J. Iron Steel Inst.,
200(1962), 934.
- 11) P. M. Kelly: Acta Met., 13(1965), 635.
- 12) H. Schumann: Arch. Eisenhütt., 38(1967), 647.
- 13) B. D. Cullity: X-Ray Diffraction, Addison-Wesley Co. Ltd., (196—)
- 14) P. C. Maxwell, A. Goldberg and J. C. Shyne: Met. Trans.,
5(1974), 1305.
- 15) G. C. Rauch and W. C. Leslie: Met. Trans., 3(1972), 373.
- 16) A. Goldberg and K. G. Hoge: Materials Science and Engineering,
13(1974), 211.
- 17) M. Fujita and I. Uchiyama: J. Iron and Steel Inst. Japan,
60(1974), 525.

CHAPTER III OBSERVATION OF BANDED STRUCTURE FORMED BY DEFORMATION IN FE-NI-C ALLOYS

3.1 Introduction

Maxwell et al⁽¹⁾ observed a kind of banded structure (such as shown in Photo.1.2) in Fe-Ni-C alloys, and they reported that such a structure is a new type of martensite which is formed by heavy deformation above the M_s temperature. Since this structure can only be observed when the specimen is plastically deformed, they proposed that this structure is the strain-induced martensite which should be classified from the stress-assisted one, and considered that such martensite was formed by the strain-induced nucleation mechanism proposed by Olson and Cohen⁽²⁾.

Maki et al⁽³⁾ reported that in Fe-Ni and Fe-Ni-C alloys, the morphologies of deformation-induced martensite are essentially the same as those of thermally transformed martensite. If the banded structure is martensite, the morphology is completely different from that usually observed in Fe-Ni-C alloys (lenticular martensite, thin plate martensite or lath martensite).

Maxwell et al⁽¹⁾ reported that such a structure could be formed at the temperature more than 150K above the M_s temperature in an Fe-Ni-C alloy. Since the temperature is extremely high above the M_s , it is doubtful whether or not the banded structure is martensite. Then, in this study, the banded structure was examined by optical and transmission electron microscope using Fe-Ni-C alloy

From these observations, it was shown that the banded structure, which Maxwell et al called strain-induced martensite,

was not martensite but deformation twins of austenite.

3.2 Experimental Procedure

Alloys were prepared by vacuum induction melting. Chemical compositions and M_s temperatures are shown in Table 3.1. The ingots were homogenized at 1473K for 24h, and cold-rolled to sheets of 0.8 mm thickness. Then, specimens to apply tensile deformation were machined into the shape shown in Fig.3.1. Specimens were austenitized at 1273K for 1h for both alloys, and quenched in iced brine.

Tensile deformation was applied by an Instron type testing machine(max. 2t) at various temperatures from 291K to 473K. Temperatures above room temperature were obtained in oil bath. Microstructures of deformed specimens were examined by optical and transmission electron microscope(JEOL JEM-120). 35% $\text{Na}_2\text{S}_2\text{O}_5$ solution was used as an etching solution for the optical microscopy.

Thin foils were prepared from the specimen(the thickness before deformation was 2 mm) which rolled to 30% strain at 373K, by both chemical($\text{HF}:\text{H}_2\text{O}_2:\text{H}_2\text{O} = 1:17:2$) and electro-polishing($\text{Cr}_2\text{O}_3:\text{CH}_3\text{COOH}:\text{H}_2\text{O} = 5:25:1$).

3.3 Results and Discussion

3.3.1 Optical microstructure

Photo.3.1(a) and (b) show typical banded structures observed in a specimen of the Fe-20%Ni-0.7%C alloy rolled to 30% strain at

Table 3.1 Chemical compositions(wt%) and M_s temperatures(K) of alloys.

Alloy	C	Si	Mn	P	S	Ni	Fe	M_s
Fe-20%Ni-0.7%C	0,73	0.003	< 0.003	0.004	0.004	19.50	bal.	228
Fe-23%Ni-0.5%C	0.47	0.003	< 0.003	0.003	0.004	22.60	bal.	228

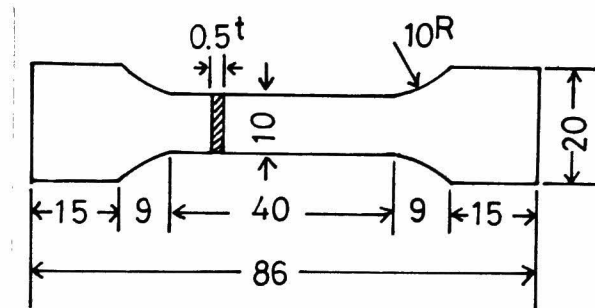
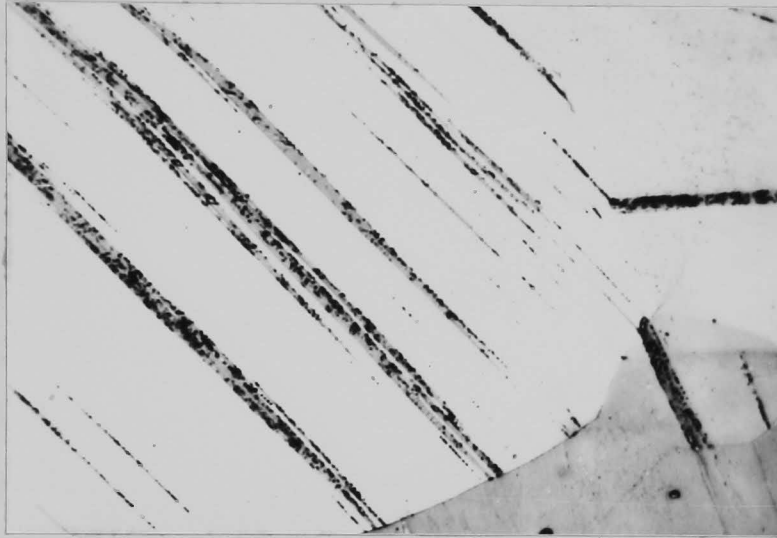


Fig.3.1 Shape and size of specimen (mm).

(a)



(b)

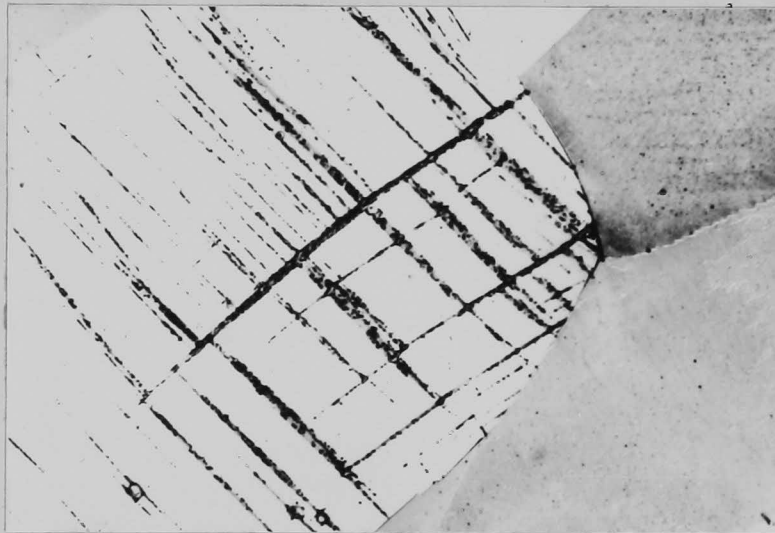


Photo.3.1 Optical microstructures of the Fe-20%Ni-0.7%C alloy($M_S = 228K$) rolled to 30% reduction at 373K, etched by 35% $Na_2S_2O_5$ solution. (x 700)

373K. Maxwell et al⁽¹⁾ called these structures as strain-induced martensite. Banded structures form in parallel along one direction (Photo.3.1(a)) or along two directions (Photo.3.1(b)). Small dark etched spots are sometimes observed in the banded area.

3.3.2 Transmission electron microscopic observation

Photo.3.2(a) shows a transmission electron microstructure of the specimen of the Fe-20%Ni-0.7%C, Photo.3.2(b) shows a selected area electron diffraction pattern taken from the region marked by square in Photo.3.2(a), and Photo.3.2(c) is its key diagram. It can be seen from Photo.3.2(a) that the banded structure consists of many thin plates. The maximum thickness of them is about 0.2 μ m. Miller indices of the diffraction spots from bands are marked with subscript (t). The key diagram shows that the two regions of different orientations are both austenite and they are twin related along $(111)_\gamma$ twinning plane. The direction of the streaks observed in Photo.3.2(b) is perpendicular to the $(111)_\gamma$ twinning plane. These streaks suggest that the twinned regions are considerably thin. Photo.3.3(a) shows the banded structure in a bright field image which is lower magnification of Photo.3.2(a). Photo.3.3(b) is a dark field image from a diffraction indexed as the $(00\bar{2})$ of the matrix. Photo.3.3(c) is a dark field image corresponding to the $(11\bar{1})_t$ of twin. The complete reversal of contrast between the two dark field images is clear.

Photo.3.4 shows an optical micrograph of a specimen(the thickness before rolling is 2 mm) deformed by rolling to 30% reduction at 373K and cooled to 253K. The banded structure are

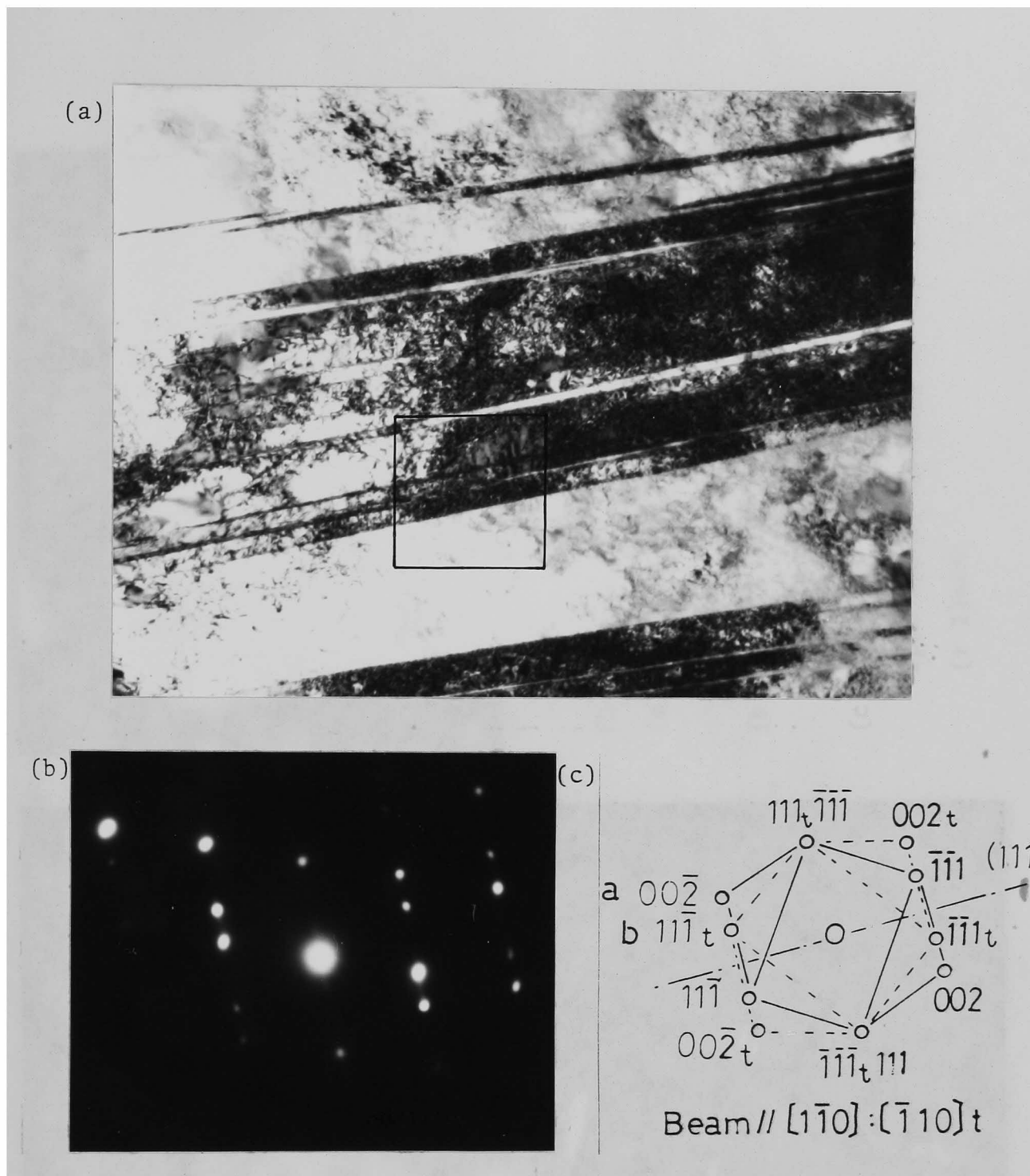


Photo.3.2 Transmission electron micrograph of the Fe-20%Ni-0.7%C alloy rolled to 30% reduction at 373K.

(a): Bright field image. (x 63000)

(b): Selected area electron diffraction pattern taken from the square in (a).

(c): Key diagram of (a).



(a)



(b)



(c)

Photo.3.3

(a):The banded structure in a bright field image.

(b):Dark field image from a diffraction spot $(00\bar{2})$ of matrix in Photo.3.2(c).

(c):Dark field image from a diffraction spot $(11\bar{1})_t$ of twin in Photo.3.2(c).
(x 21000)

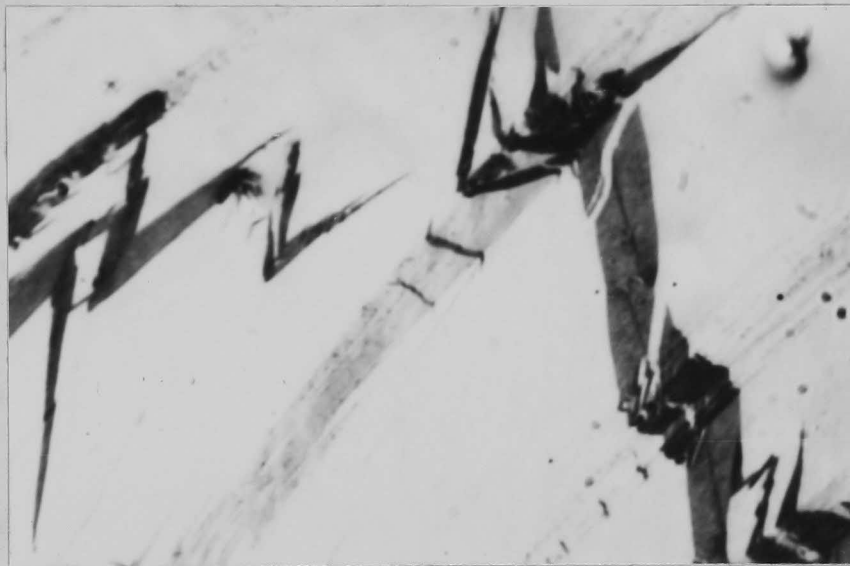


Photo.3.4 Optical micrograph of the Fe-20%Ni-0.7%C alloy rolled 30% at 373K and cooled to 253K, etched by 35% $\text{Na}_2\text{S}_2\text{O}_5$ solution. (x 700)

formed by rolling and lenticular martensites are formed by cooling. (It seems that the M_s temperature was increased due to the deformation.) Some of martensite plates grow across the banded structure. In such a case martensite plate changes its growing direction in the banded structure and shows a kinked morphology. Since the banded structure is consisted of several bands, the crossing mode of martensite plate is rather complicated but this kinking is another strong evidence that the banded structure is deformation twins of austenite.

From these observations, it is now clear that the banded structure(Photo.3.1) is a group of thin deformation twins of austenite.

3.3.3 Stacking fault energy of Fe-Ni-C austenite

It has been considered that the stacking fault energy of Fe-Ni-C austenite is too high to form deformation twins, and only Jan Kounicky⁽⁴⁾ reported that in the Fe-25%Ni-0.5%C austenite, the deformation twins were observed by 50% reduction at room temperature. In this study, it was confirmed that the deformation twinning could occur in Fe-Ni-C austenite.

Venables⁽⁵⁾ reported equation(3.1) as the relationship between twinning stress(τ_T) and stacking fault energy(γ).

$$(1/\sqrt{3})\{1/3 + K(\tau_T/\mu)\}(\tau_T/\mu) = \gamma/\mu b \quad \text{-----}(3.1)$$

where K: constant (700 ~ 900)

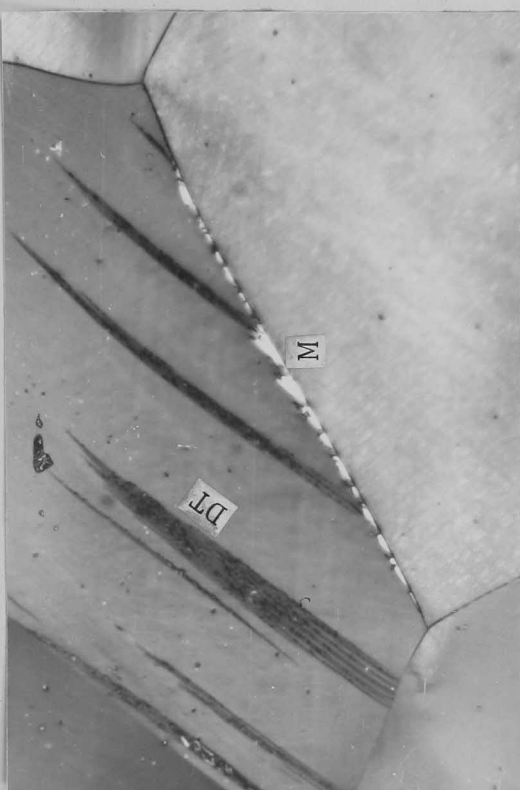
μ : shear modulus (7.4×10^4 MPa)

b: burgers vector of perfect dislocation

The stacking fault energy of the Fe-20%Ni-0.7%C and the Fe-23%Ni-0.5%C austenite was calculated by Eq.3.1. About $5.0 \times 10^{-2} \text{ J/m}^2$ was obtained for both alloys from the twinning stress (σ_T = about 500MPa) which was determined as the stress at which twinning was first observed by optical microscope at room temperature. The twinning stress (σ_T) was transferred to shear stress (γ_T) by Taylor's relation⁽⁶⁾ ($\sigma_T = 3.06 \gamma_T$). In the case of stainless steels, in which deformation twins occur, the stacking fault energy is $2.0 \times 10^{-2} \sim 3.0 \times 10^{-2} \text{ J/m}^2$. Comparing the estimated stacking fault energy for Fe-Ni-C alloys with this value, the stacking fault energy ($5.0 \times 10^{-2} \text{ J/m}^2$) obtained in this study seems to be not too large.

3.3.4 Corelation between deformation twinning and martensitic transformation

In the temperature range below the M_d temperature, there are possibilities that both twinning of austenite and martensitic transformation occur by deformation. Photo.3.5(a), (b), (c) and (d) show optical microstructures taken from the Fe-23%Ni-0.5%C specimens deformed to about 30% strain in tension at 291K, 323K, 373K and 473K, respectively. In the temperature range near the M_s , martensite formed by deformation can easily be identified as shown in Photo.3.5(a) and (b). However, as the deformation temperature increases, the amount of martensite formed by deformation becomes less and difficult to be identified. In the case of Fe-23%Ni-0.5%C alloy, small dark etched spots are seen in the banded area as shown



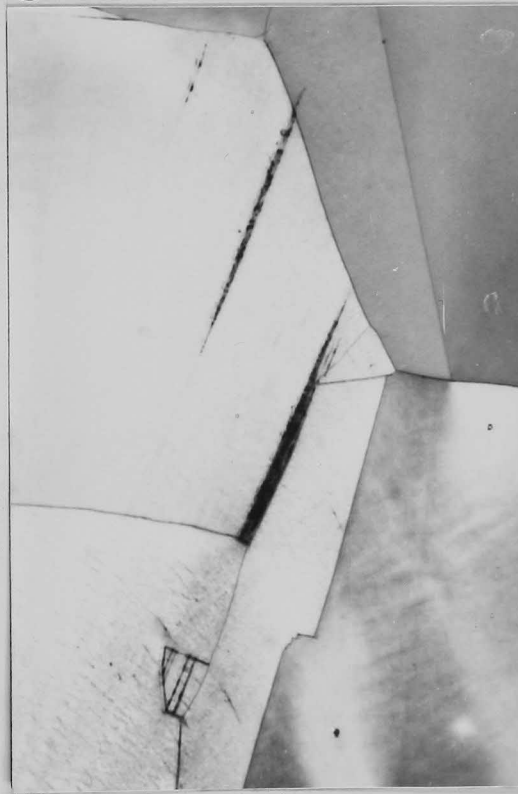
(a)



(b)



(c)



(d)

Photo.3.5 Optical microstructures taken from the Fe-23%Ni-0.5%C specimens deformed to about 30% strain in tension at 291K(a), 323K(b), 373K(c) and 473K(d), respectively, etched by 35% $\text{Na}_2\text{S}_2\text{O}_5$ solution. (x 700)

in Photo.3.5(c) and (d). Such small dark etched spots are also seen in an Fe-20%Ni-0.7%C alloy(Photo.3.1) more clearly.

To examine the magnetic characteristic of these spots, an experiment was carried out. The specimen(Photo.3.6(a)) which had been deformed to 30% reduction at 373K and formed banded structure was magnetized and immersed in alcohol where fine iron particles were dispersed. Fine iron particles were found to be preferentially attached to the small dark etched spots(Photo.3.6(b)). The microstructures of Photo.3.6(a) and (b) are taken from the same region. This indicates that the small dark etched spots are ferromagnetic while the matrix is paramagnetic. From this experiment, there is a possibility that the small dark etched spots are martensite. However, Georgiyeva et al⁽⁷⁾ reported that in Fe-Ni-C austenite, the clustering of carbon atoms occurs at the imperfections introduced by deformation, and the Curie temperature of these regions is remarkably increased. This suggests another possibility that the small dark etched spots are ferromagnetic austenite with higher Curie temperature than the matrix.

3.4 Conclusion

From the optical and electron microscopic observations of the banded structure which formed by deformation in Fe-Ni-C alloys, the following results are obtained.

(1) The banded structure formed by deformation, which Maxwell et al called strain-induced martensite, is not martensite but a group of thin deformation twins of austenite.

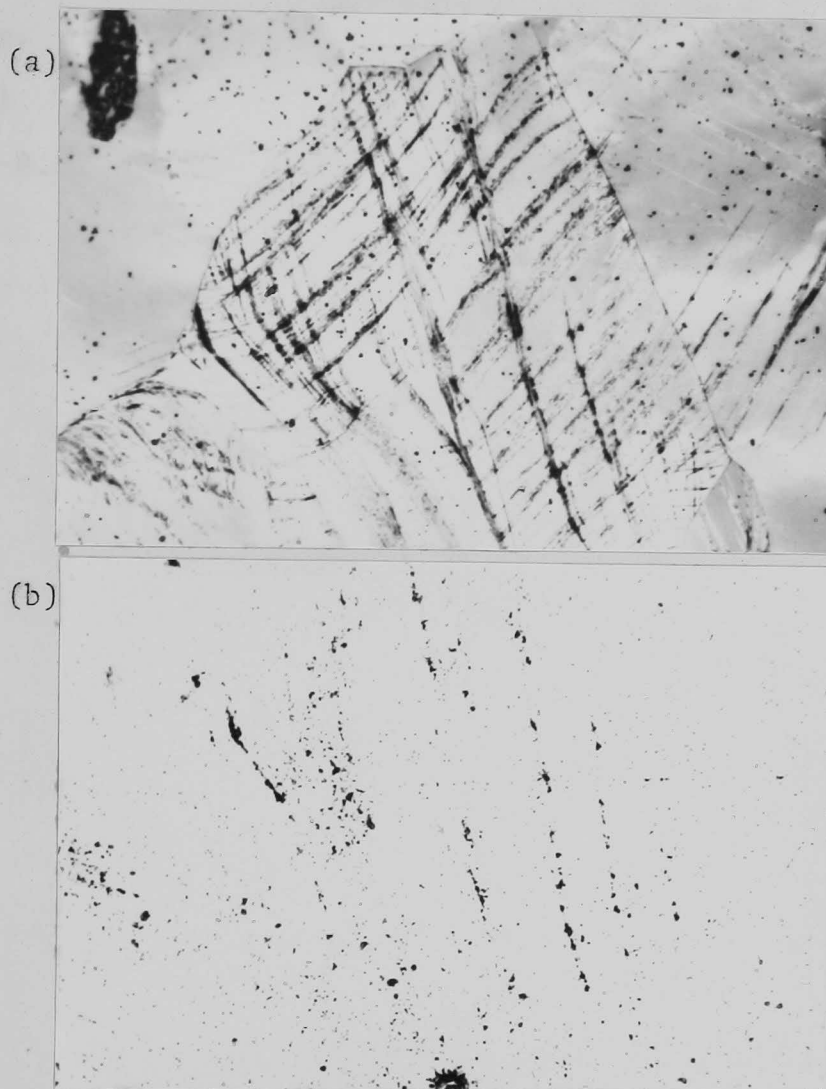


Photo.3.6 (a): Banded structures which formed in the Fe-20%Ni-0.7%C specimen by 30% reduction at 373K. (x 700)

(b): Fine iron particles which attached to the small dark etched spots in the banded area in (a). (b) is the same area as (a). (x 700)

(a) was etched by 35% $\text{Na}_2\text{S}_2\text{O}_5$ solution.

(2) The small dark etched spots in the banded area is ferromagnetic and it can be either martensite or ferromagnetic austenite.

(3) The stacking fault energy of the Fe-20%Ni-0.7%C and the Fe-23%Ni-0.5%C austenite was estimated to be about $5.0 \times 10^{-2} \text{ J/m}^2$.

References for Chapter III

- 1) P. C. Maxwell, A. Goldberg and J. C. Shyne: Met. Trans., 5(1974), 1305.
- 2) G. B. Olson and M. Cohen: J. Less-Common Metals, 28(1972), 107.
- 3) T. Maki, S. Shimooka, M. Umemoto and I. Tamura: Trans. JIM, 13(1972), 400.
- 4) J. Kounicky: Phil. Mag., 18(1968), 205.
- 5) J. A. Venables: Phil. Mag., 7(1962), 35.
- 6) G. I. Taylor: J. Inst. Metals, 62(1938), 307.
- 7) I. Ya. Georgiyeva, G. V. Kurdjumov, O. P. Maksimova and V. V. Nemirovskiy: Fiz. Met. Metalloved, 23(1967), 1070.

CHAPTER IV EFFECT OF STRAIN ON DEFORMATION-INDUCED MARTENSITIC TRANSFORMATION

4.1 Introduction

The purpose of this Chapter is to make clear the effect of plastic strain on the deformation-induced martensitic transformation. There are two different views about the effect of strain on the deformation-induced martensitic transformation. One^{(1)~(3)} is that the concentrated stress at obstacles (grain boundary or twin boundary etc) induces the transformation, and the other is the strain-induced nucleation proposed by Olson and Cohen⁽⁴⁾.

About the effect of the concentrated stress on the deformation-induced martensitic transformation, Suzuki et al⁽²⁾ and Ueda and Fujita⁽³⁾ studied the interaction between slip bands and deformation-induced martensitic transformation. Since there are 24 slip systems, it is difficult to determine the operating slip systems from Schmid factor. So, in these experiments^{(2),(3)}, the relation between the orientation of martensite and the direction of internal concentrated stress was not made clear.

Venables⁽⁵⁾ observed α' -martensite which was formed at the intersection of two ϵ' -martensite bands in deformed 18-8 stainless steel. Olson and Cohen⁽⁴⁾ suggested that such a martensite was formed by intersection of two shear bands corresponding to those of Bogers and Burgers⁽⁶⁾ double shear mechanism. Olson and Cohen also reported that such a martensite was formed in TRIP steel by deformation. Lecroisey and Pineau⁽⁷⁾ observed martensite which nucleated at the twin boundary and grew along the intersecting

deformation bands in deformed Fe-Ni-Cr-C alloy, and suggested almost the same mechanism as that of Olson and Cohen. However, in these studies^{(4),(6),(7)}, the systems of two intersecting deformation bands were not analyzed crystallographically. So, it is not clear whether or not such martensites are formed from the embryo which suggested by Olson and Cohen.

Shear plane and shear direction of deformation twin can be determined by calculating Schmid factor. So, the direction of local concentrated stress at the grain boundary or the annealing twin boundary on which deformation twins impinged can be determined. In this study, the relation between the direction of local concentrated stress at the grain boundaries or annealing twin boundaries on which deformation twins impinged and the orientation of martensite was studied to make clear the effect of local concentrated stress on the deformation-induced martensitic transformation. The martensite which formed at the junction of two deformation twins was examined whether or not such martensite was formed by the intersection of two shear bands of Bogers and Burgers type suggested by Olson and Cohen. Operating twinning systems are different between tensile and compressive deformation. By tension, deformation twinning does not occur when tensile direction is near $[001]_{\gamma}$. On the other hand, by compression, deformation twinning occurs only when compressive direction is near $[001]_{\gamma}$ ^{(8)~(10)}. The difference between the contribution of tensile and compressive deformations on the strain-induced embryo formation model which was proposed by Olson and Cohen was also discussed.

4.2 Experimental Procedure

4.2.1 Specimen preparations

An Fe-25%Ni-0.7%C alloy and an Fe-25%Ni-0.4%C alloy were prepared by vacuum induction melting. Chemical compositions and M_s and M_d temperatures were listed in Table 4.1. To examine the difference of effect between tensile and compressive deformation on the deformation-induced martensitic transformation, both tensile and compressive tests were performed with an Instron type testing machine (max. load 2t). Compressive tests were performed using cubic specimens (3x3x3 mm) of an Fe-25%Ni-0.4%C alloy to compare with the results of tensile tests of an Fe-25%Ni-0.7%C alloy. Other results were obtained from tensile experiments on an Fe-25%Ni-0.7%C alloy.

The ingot of an Fe-25%Ni-0.7%C alloy was hot-rolled at 1473K and homogenized in vacuum furnace at 1473K for 24h, and then cold-rolled to make a sheet of 0.8 mm thickness. Specimens for tensile tests as shown in Fig.4.1 were machined from the sheet. Specimens (a) were used for metallographic examinations after tensile deformation. Specimens(b) were used to determine the 0.2% proof stress and the critical stress for martensite formation. All the specimens were austenitized at 1423K for 1h in a vacuum furnace, and were quenched into iced brine. The specimens were deformed to various strain at various temperatures (183K ~ 303K), and examined by optical microscope. 35% $\text{Na}_2\text{S}_2\text{O}_5$ solution was used as an etching solution. Temperatures below the room temperature were obtained in a cold chamber which cooled by controlling evaporation of liquid nitrogen. The heat treatment of an Fe-25%Ni-0.4%C alloy was performed in the

Table 4.1 Chemical compositions (wt%) and M_S and M_d temperatures (K) of alloys used.

Alloy	C	Si	Mn	P	S	Ni	Fe	M _S	M _d
Fe-25%Ni-0.7%C	0.65	0.067	0.17	0.010	0.009	25.31	bal.	163	253
Fe-25%Ni-0.4%C	0.36	0.02	0.01	0.002	0.007	24.54	bal.	223	—

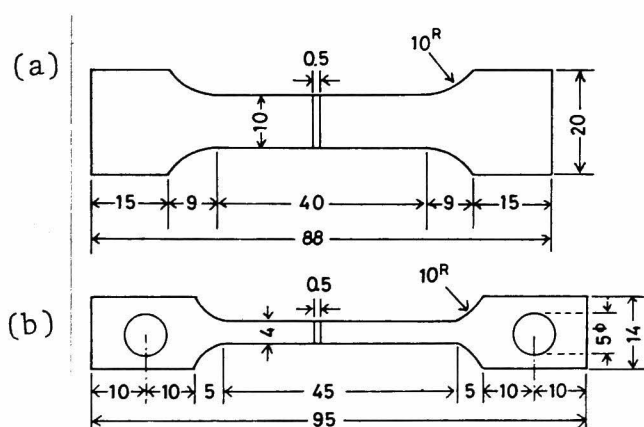


Fig.4.1 Shape and size of specimen (mm).

same way as the case of an Fe-25%Ni-0.7%C alloy. Compressive tests were performed at temperatures from 298K to 333K. The temperatures were obtained in hot water bath.

4.2.2 Determination of martensite variants

The orientation of austenite was determined by measuring the four angles (δ , α , β and γ) formed by any two out of 4 $\{111\}_\gamma$ annealing or deformation twin traces. Photo.4.1 shows a typical example of structure which has four $\{111\}_\gamma$ traces. Four $\{111\}_\gamma$ traces were moved in parallel to cross at one point as shown in Fig.4.2. Fig.4.2 shows the case of Photo.4.1. The maximum angle among the four angles was called as δ , and the others were called as α , β and γ along counterclockwise from δ .

The orientation of surface normal $P(P_1P_2P_3)$ was determined using the following equations suggested by Shugo et al⁽¹¹⁾.

$$\begin{aligned}
 P_1 &= \sqrt{(A^2 - BC) / (A^2 + B^2 + C^2 - AB - BC - CA)} \\
 P_2 &= \sqrt{(B^2 - CA) / (A^2 + B^2 + C^2 - AB - BC - CA)} \quad \text{----- (4.1)} \\
 P_3 &= \sqrt{(C^2 - AB) / (A^2 + B^2 + C^2 - AB - BC - CA)} \\
 A &= -\cos\delta\cos\beta \\
 B &= \cos(\alpha + \beta)\cos(\beta + \gamma) \\
 C &= \cos\alpha\cos\gamma \\
 \alpha + \beta + \gamma + \delta &= \pi
 \end{aligned}$$

The variant of martensite was determined by two surface analysis. In the Fe-25%Ni-0.7%C alloy, lenticular martensite was

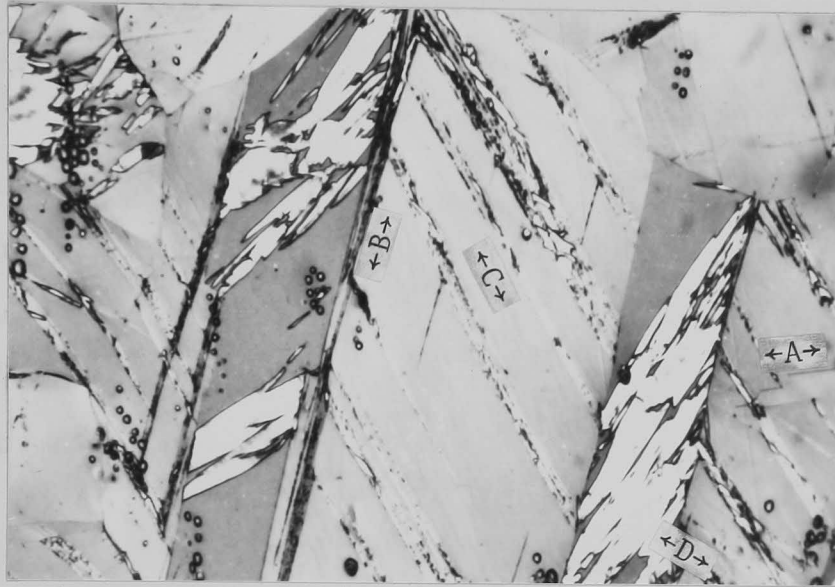


Photo.4.1 Typical example of structure which has four $\{111\}_{\gamma}$ traces. (x 700)

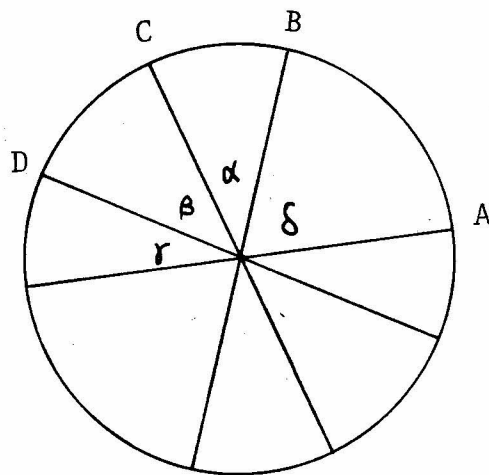


Fig.4.2 Four $\{111\}_{\gamma}$ twin traces which were moved in parallel to cross at one point. A, B, C and D are corresponded to the same symbols in Photo.4.1.

formed by deformation, and the habit plane was reported as $\{3\ 15\ 10\}_\gamma^{(12)}$. Photo.4.2 is a typical example of two surface analysis. Arrow indicates the trace of a mid-rib of martensite plate on each surface. The angle between two surface was measured by comparing the relative angles between directions of $\{111\}_\gamma$ traces on each surface on stereographic projection, and the deviation from 90° was within $\pm 2^\circ$.

4.3 Results

4.3.1 Temperature dependence of critical stress for martensite formation by tensile stress

Fig.4.3 shows critical applied tensile stress of martensite formation (σ_m) and 0.2% proof stress ($\sigma_{0.2}$) at various temperatures above the M_s . σ_m was determined as the stress where the first stress drop occurred in stress-strain curve. Below M_s^σ (about 210K), martensitic transformation is induced before austenite is yielded, and 0.2% proof stress decreases with decrease in temperature. This decrease is due to the fact that martensitic transformation is induced by elastic stress. Such a phenomenon has been reported by Bolling and Richman⁽¹⁾ and also many other workers^{(13),(14)}. Below the M_s , 0.2% proof stress increases with decrease in temperature, since amount of martensite transformed thermally before applying the tensile stress increases with decrease in testing temperature.

The critical stress for martensite formation increases linearly with increase in temperature in the range between M_s and

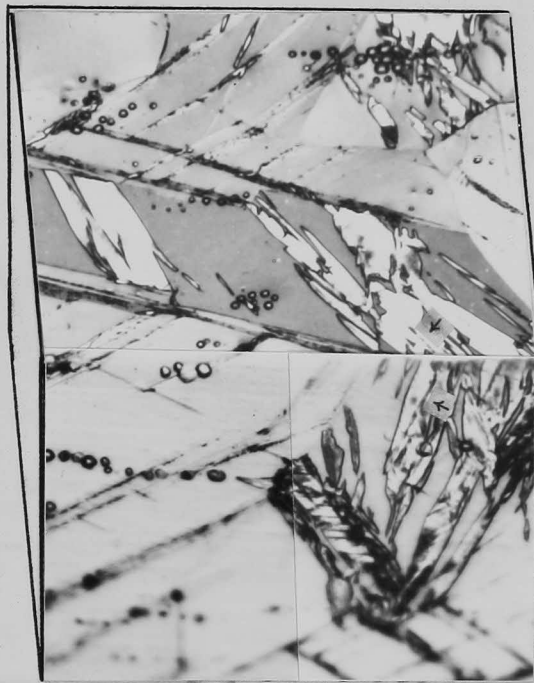


Photo.4.2 Typical example of two surface analysis. Arrow indicates the trace of mid-rib on each plane.
(x 700)

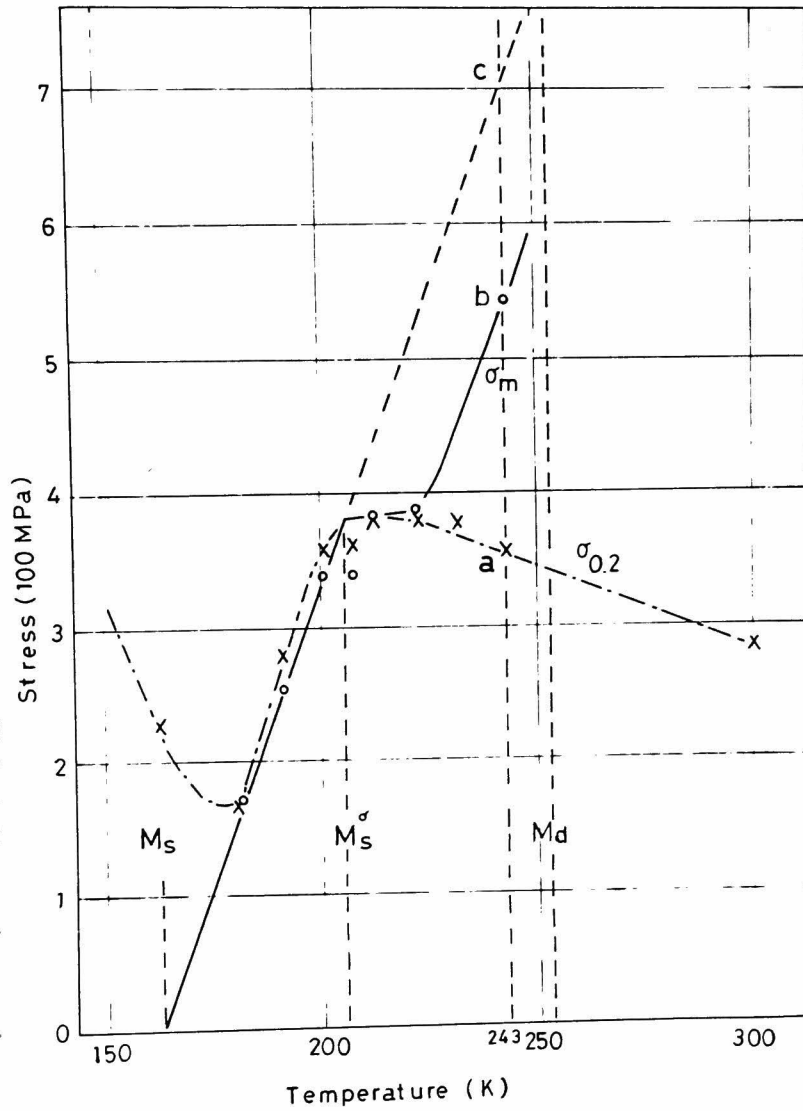


Fig.4.3 Temperature dependence of critical tensile stress for martensite formation, o (σ_m) and 0.2% proof stress, x ($\sigma_{0.2}$) for the Fe-25%Ni-0.7%C alloy.

M_S^σ . However, above M_S^σ (243K in this case), austenite begins to deform at stress (a, 350 MPa) and is required to strain-harden up to (b, 550 MPa), to transform martensite. The stress (b) is considerably lower than (c, about 700 MPa) expected value by extrapolating the stress-temperature line between M_S and M_S^σ . The same tendency was reported by Tokizane⁽¹⁵⁾. This decrease of critical applied stress for martensite formation (c-b) by plastic strain prior to the formation of martensite has been observed by many researchers.

4.3.2 Optical microstructure

Photo.4.3 shows an optical microstructure taken from the specimen of the Fe-25%Ni-0.7%C alloy deformed to 5% strain at 183K (between M_S and M_S^σ). The martensite is lenticular and the same morphology as the one of thermally transformed. The same tendency was also reported by Maxwell et al⁽¹⁶⁾.

Photo.4.4 shows optical microstructures taken from the specimen of the Fe-25%Ni-0.7%C alloy deformed to 22% strain in tension at 243K. At 243K, both slip and deformation twinning are introduced in the parent austenite. Above the stress (b, in Fig.4.3), the martensite was observed at the intersection of two deformation twins (Photo.4.4(a)) and at the annealing twin boundary (Photo.4.4(b)) or the grain boundary (Photo.4.4(c)) on which deformation twins impinged.

The stress concentration can be produced at the grain boundary or the annealing twin boundary on which deformation twins impinged. From the above observations, the concentrated stress is considered to be quite effective for inducing martensite above M_S^σ temperature.

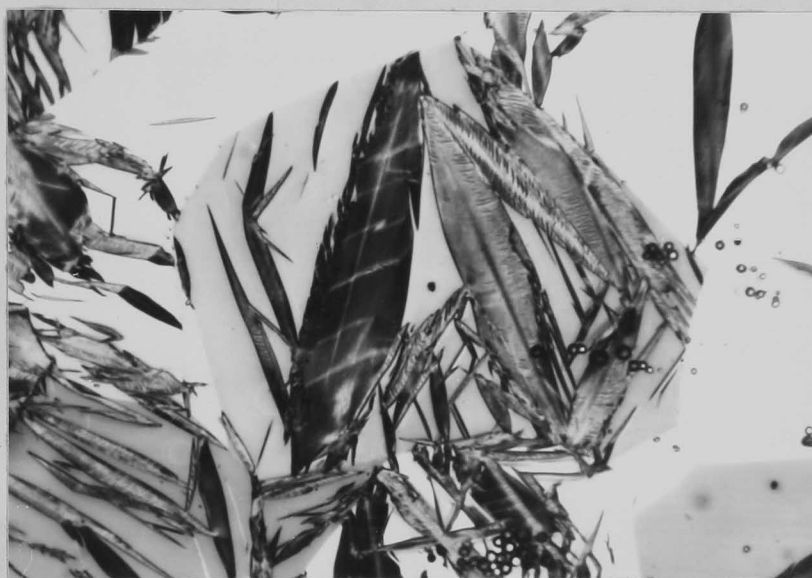


Photo.4.3 Optical microstructure of the specimen
of Fe-25%Ni-0.7%C alloy deformed to 5%
strain in tension at 183K (below M_S^σ).
Etched by 35% $\text{Na}_2\text{S}_2\text{O}_5$ solution. (x 700)

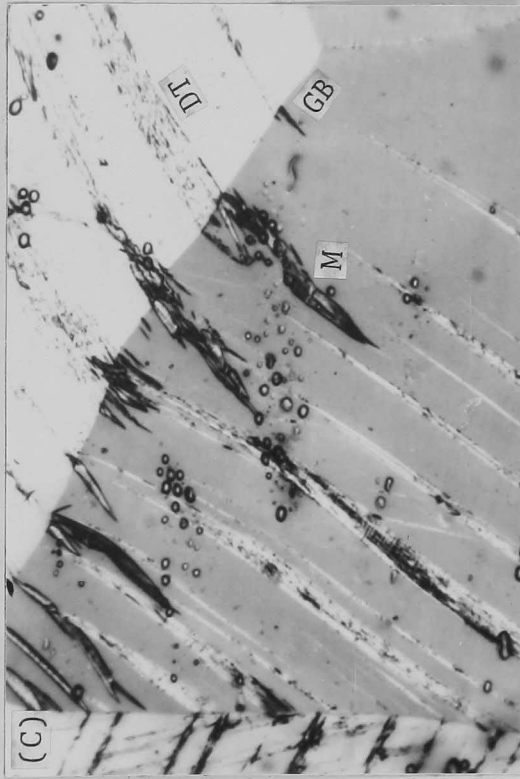
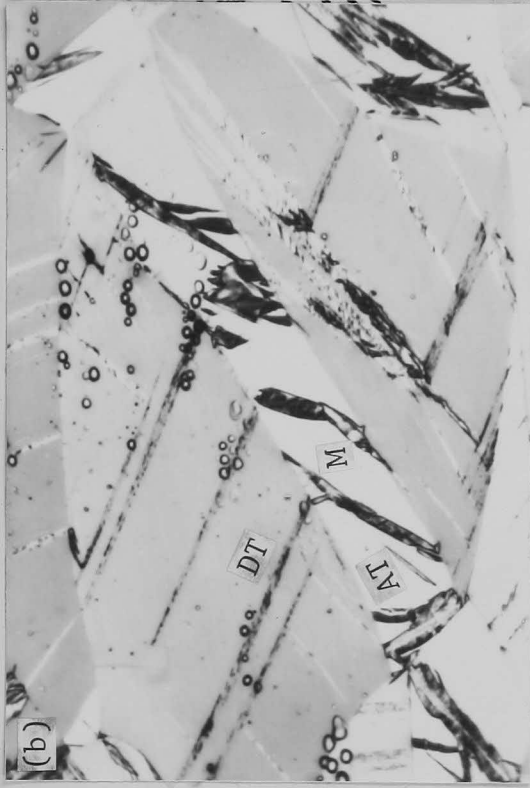


Photo.4.4 Microstructures of the Fe-25%Ni-0.7%C specimen which deformed to 22% strain in tension at 243K.
 (a): martensite(M) which formed at the intersection of deformation twins. (b): martensite(M) which formed from the annealing twin boundary(AT) on which deformation twins(DT) impinged. (c): martensite (M) which formed from the grain boundary(GB) on which deformation twins(DT) impinged. (x700)

Although the critical stress for martensite formation is (b) at 243K, the effective concentrated stress is considered to be (c) in Fig.4.3.

4.4 Discussion

4.4.1 Role of stress concentration

Photo.4.5 shows an optical microstructure taken from the specimen deformed to 22% strain at 243K. Martensite(M) was formed and grown from the annealing twin boundary(AT) at which deformation twins(DT) impinged. If martensite is formed by the aid of the concentrated stress due to the impingement of deformation twins, the variant of martensite should depend upon the shear direction and the shear plane of the impinging deformation twin.

According to Livingston and Chalmers⁽¹⁷⁾, if deformation twinning system(i) (which has an unit vector e_i as its twinning plane normal and g_i as its twinning direction) has occurred in crystal A and impinges on the grain boundary between crystal A and crystal B, and if the acting stress is P_i , the resultant shear stress P_j on twinning system(j) (which has an unit vector e_j as its twinning plane normal and g_j as its twinning direction) will be as follows.

$$P_j = P_i \times N_{ij} = P_i [(e_i \cdot e_j)(g_i \cdot g_j) + (e_i \cdot g_j)(e_j \cdot g_i)] \quad \text{---(4.2)}$$

In the case of Photo.4.5, the surface normal was calculated from equation(4.1) using four twin traces(A, B, C and D). Fig.4.4 is the stereographic projection showing the relation among the surface normal(P), the tensile direction(T) and four twin traces

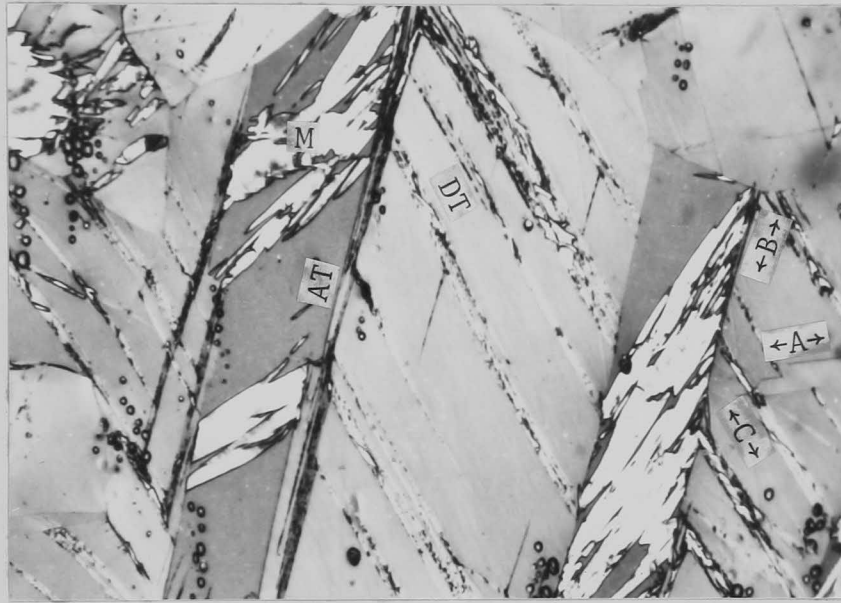


Photo.4.5 Optical microstructure of the Fe-25%Ni-0.7%C specimen deformed to 22% strain in tension at 243K. Martensite(M) in the annealing twin is considered to be formed from the junction of annealing twin boundary(AT) and deformation twins(DT). (x 700)

(A, B, C and D) in Photo.4.5. The annealing twin boundary(AT) was taken as $(1\bar{1}1)_\gamma$. The shear plane of the impinging deformation twin (DT) is $(\bar{1}11)_\gamma$. Calculated Schmid factor for the twinning system $\{111\}_\gamma$, $\langle 112 \rangle_\gamma$ is shown in Table 4.2. The twinning direction(DT) on the $(\bar{1}11)_\gamma$ plane was determined as $[\bar{1}\bar{2}1]_\gamma$ since it has the highest Schmid factor(0.33) among three directions on the $(\bar{1}11)_\gamma$.

N_{ij} value(stress coupling factor) for the twinning system $(\bar{1}11)_\gamma$, $[\bar{1}\bar{2}1]_\gamma$ and the shear of martensitic transformation on the habit plane was calculated for 24 martensite variants. The calculated values are listed in Table 4.3. In Table 4.3, Schmid factor for the shear of martensitic transformation on the habit plane and the direction of applied tensile stress is also shown. The shear system $\{3\ 15\ 10\}_\gamma$, $\langle .24\ .60\ .77 \rangle_\gamma$ was used as the shear system of martensitic transformation which was calculated by W-L-R⁽¹⁸⁾. Since martensite was formed inside $(1\bar{1}1)_\gamma$ annealing twin, the shear system was transferred to the indices of matrix by the equation(4.3).

$$[h\ k\ 1]_M = (1/3)[h\ k\ 1]_T \begin{pmatrix} \bar{1} & \bar{2} & 2 \\ \bar{2} & \bar{1} & \bar{2} \\ 2 & \bar{2} & \bar{1} \end{pmatrix} \text{-----(4.3)}$$

The observed martensite variant was determined by the two surface analysis which was shown in Photo.4.2. Fig.4.5 shows the key diagram. The habit plane of martensite in Photo.4.5 was taken as $(11\ \bar{2}\bar{6}\ 47)_\gamma$ which was nearest to H in Fig.4.5. The observed variants were marked by -o in Table 4.3. In Table 4.3, martensite variants are listed in order of the larger value of N_{ij} . Observed martensite variants have relatively larger N_{ij} values and Schmid factors. The same analysis as above were done for 7 martensite

Table 4.2 Calculation of the Schmid factor for
the shear system $\{111\}_\gamma$, $\langle 112 \rangle_\gamma$ when
the tensile directions are $[\bar{0}4 \bar{5}6 .53]_\gamma$
(Photo.4.5) and $[\bar{5}5 \bar{7}3 .41]_\gamma$ (Photo.4.6).

Shear Plane	Shear Direction	Schmid Factor	
		Photo.4.5	Photo.4.6
(111)	$[11\bar{2}]$ C δ	0.36 -D	0.43 -A
	$[1\bar{2}1]$ A δ	-0.16	-0.27
	$[\bar{2}11]$ B δ	-0.20	-0.16
$(\bar{1}11)$	$[\bar{1}\bar{2}1]$ C β	0.33 -C	0.13 -C
	$[211]$ D β	-0.19	-0.07
	$[\bar{1}1\bar{2}]$ A β	-0.14	-0.05
$(1\bar{1}1)$	$[\bar{2}\bar{1}1]$ C α	0.25 -B	0.31 -B
	$[1\bar{1}\bar{2}]$ B α	-0.12	-0.09
	$[121]$ D α	-0.13	-0.22
$(\bar{1}\bar{1}1)$	$[2\bar{1}1]$ A γ	-0.08	0.02
	$[\bar{1}\bar{1}\bar{2}]$ D γ	0.06	0.18
	$[\bar{1}21]$ B γ	0.02	-0.20

Table 4.3 N_{ij} value and Schmid factor for 24 martensite variants. -o: observed variants.

Habit Plane	Shear Direction	Index Plane	in Matrix Direction	N_{ij}	Schmid Factor	
(15 3 10)	[.60 .24 .77]	($\bar{1}$ $\bar{53}$ 14)	[.55 .19 .81]	0.65	0.47	-O
(15 $\bar{3}$ 10)	[.60 .24 .77]	(11 $\bar{47}$ 26)	[.87 .03 .50]	0.48	0.48	-O
(15 10 3)	[.60 .77 .24]	(29 $\bar{46}$ 7)	[.15 .02 .99]	0.44	0.42	
(3 15 10)	[.24 .60 .77]	($\bar{13}$ $\bar{41}$ $\bar{34}$)	[.83 .47 .30]	0.37	0.03	
(15 $\bar{10}$ 3)	[.60 .77 .24]	(11 $\bar{26}$ 47)	[.87 .50 .03]	0.21	0.48	-O
($\bar{15}$ 10 3)	[.60 .77 .24]	(1 14 $\bar{53}$)	[.55 .81 .19]	0.18	0.47	
(3 $\bar{15}$ 10)	[.24 .60 .77]	(47 $\bar{11}$ 26)	[.03 .87 .50]	0.17	0.04	
($\bar{15}$ $\bar{10}$ 3)	[.60 .77 .24]	(41 34 $\bar{13}$)	[.47 .30 .83]	0.12	0.29	
(3 10 15)	[.24 .77 .60]	(7 $\bar{46}$ $\bar{29}$)	[.99 .02 .15]	0.11	-0.06	
(10 15 3)	[.77 .60 .24]	($\bar{34}$ $\bar{41}$ $\bar{13}$)	[.30 .47 .83]	0.10	-0.35	
($\bar{15}$ 3 10)	[.60 .24 .77]	(29 7 $\bar{46}$)	[.15 .99 .02]	0.09	0.38	
(10 3 15)	[.77 .24 .60]	(14 $\bar{53}$ $\bar{1}$)	[.81 .19 .55]	0.05	-0.34	
(10 $\bar{15}$ 3)	[.77 .60 .24]	(26 $\bar{11}$ 47)	[.50 .87 .03]	-0.03	-0.21	
($\bar{15}$ $\bar{3}$ 10)	[.60 .24 .77]	(41 13 $\bar{34}$)	[.47 .83 .30]	-0.04	0.30	
(10 $\bar{3}$ 15)	[.77 .24 .60]	(26 $\bar{47}$ 11)	[.50 .03 .87]	-0.07	-0.22	
($\bar{3}$ 15 10)	[.24 .60 .77]	($\bar{7}$ $\bar{29}$ $\bar{46}$)	[.99 .15 .02]	-0.11	-0.04	
($\bar{3}$ 10 15)	[.24 .77 .60]	(13 $\bar{34}$ $\bar{41}$)	[.83 .30 .47]	-0.13	0.02	
(3 $\bar{10}$ 15)	[.24 .77 .60]	(47 $\bar{26}$ 11)	[.03 .50 .87]	-0.14	0.03	
($\bar{3}$ $\bar{10}$ 15)	[.24 .77 .60]	(53 $\bar{14}$ $\bar{1}$)	[.19 .81 .55]	-0.24	-0.14	
($\bar{3}$ $\bar{15}$ 10)	[.24 .60 .77]	(53 1 14)	[.19 .55 .81]	-0.26	-0.12	
($\bar{10}$ $\bar{15}$ 3)	[.77 .60 .24]	(46 29 7)	[.02 .15 .99]	-0.42	-0.48	
($\bar{10}$ 15 3)	[.77 .60 .24]	($\bar{14}$ $\bar{1}$ $\bar{53}$)	[.81 .55 .19]	-0.43	-0.32	
($\bar{10}$ $\bar{3}$ 15)	[.77 .24 .60]	(46 $\bar{7}$ $\bar{29}$)	[.02 .99 .15]	-0.54	-0.49	
($\bar{10}$ 3 15)	[.77 .24 .60]	(34 $\bar{13}$ $\bar{41}$)	[.30 .83 .47]	-0.55	-0.34	

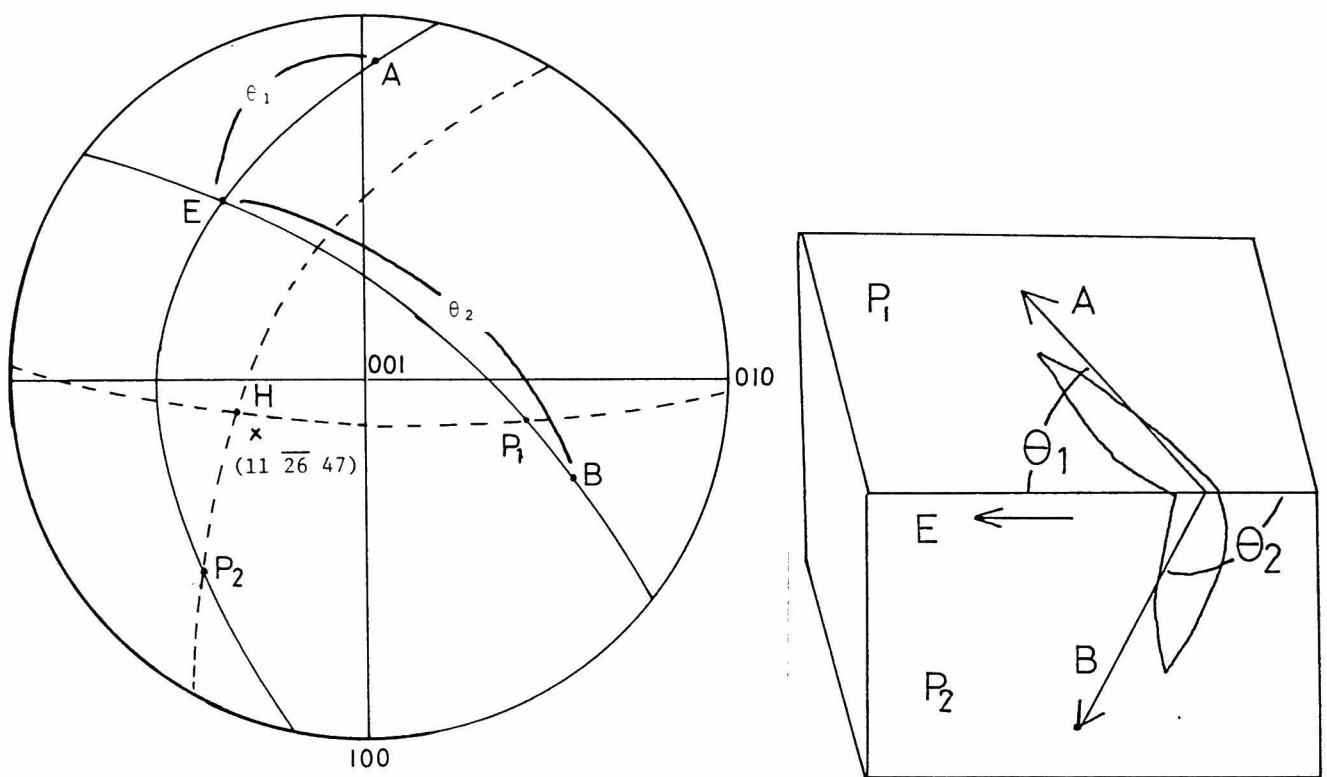


Fig.4.5 Stereographic projection of two surface analysis for Photo.4.5. The habit plane H for martensite in Photo.4.5 is determined as the point which is 90° from both A and B. $\{3\ 15\ 10\}_\gamma$ planes are transferred to matrix indices by Eq.4.3.

$(11\ \overline{26}\ 47)_\gamma$ is the nearest to H among them, and is selected as the habit plane of the martensite.

A,B: direction of the trace of habit plane on each plane.

E: edge direction.

P_1, P_2 : direction of the surface normal of each surface.

plates. All the observed martensite variants are listed in Table 4.4.

In Table 4.3 and Table 4.4, all the observed martensite variants have both large N_{ij} value and large Schmid factor. From above results, it can be concluded that martensite was formed by the aid of the concentrated stress due to the impingement of deformation twins plus by the aid of applied tensile stress.

In this study, stress was considered to aid the total shape strain on the habit plane. However, as described in Chapter I, Kato et al^{(19),(20)} considered that martensitic transformation took place by double shears(i.e., the 1st shear is $(a_\gamma/18)[11\bar{2}]_\gamma$ on $(111)_\gamma$, 2nd shear is $(a_\gamma/12)[1\bar{1}\bar{2}]_\gamma$ on $(1\bar{1}1)_\gamma$) suggested by Bogers and Burgers⁽⁶⁾, and that the applied stress aided the 1st shear.

BCC structure which is obtained by B-B double shear requires rotation to satisfy the orientation relationship such as K-S, G-T or N. The orientation relationship in the Fe-25%Ni-0.7%C alloy is considered to be G-T relationship⁽¹²⁾. The difference in orientation is within 5° among K-S, G-T and N. There is no essential trouble in considering the orientation relationship in this alloy to be K-S, G-T or N relationship. Then, all K-S and N variants were associated with B-B double shears. An example of such an association was shown in Appendix. All the martensite variants and corresponding B-B double shears are listed in Table 4.5.

N_{ij} value for the impinging deformation twinning system and the 1st shear of B-B double shear, and Schmid factor for the 1st shear and the direction of applied tensile stress are calculated for the 7 observed martensite variants. The results are also shown in Table 4.4. In Table 4.4, No.1, 4 and 6 have negative N_{ij} values for the 1st shear and all the observed martensite variants have

Table 4.4 Observed martensite variants.

No.	Observed Variant	Twin Crossing	For the total shape strain		First Shear	For the 1st shear	
			N _{ij}	Schmid Factor		N _{ij}	Schmid Factor
1.	(11 $\overline{26}$ 47), [$\overline{.87}$ $\overline{.50}$ $\overline{.03}$]	($\overline{111}$), [$\overline{121}$]	0.48	0.41	(1 $\overline{11}$), [$\overline{121}$]	-0.44	0.07
2.	($\overline{1}$ $\overline{53}$ 14), [$\overline{.55}$ $\overline{.19}$ $\overline{.81}$]	($\overline{111}$), [$\overline{121}$]	0.65	0.49	($\overline{151}$), [$\overline{712}$]	0.40	0.28
3.	(11 $\overline{47}$ 26), [$\overline{.87}$ $\overline{.03}$ $\overline{.50}$]	($\overline{111}$), [$\overline{121}$]	0.48	0.48	(1 $\overline{11}$), [$\overline{112}$]	0.39	0.06
4.	(11 $\overline{26}$ 47), [$\overline{.87}$ $\overline{.50}$ $\overline{.03}$]	($\overline{111}$), [$\overline{121}$]	0.21	0.48	(1 $\overline{11}$), [$\overline{121}$]	-0.44	0.05
5.	($\overline{11}$ $\overline{47}$ 26), [$\overline{.87}$ $\overline{.03}$ $\overline{.50}$]	(111), [$\overline{121}$]	0.48	0.44	($\overline{111}$), [$\overline{112}$]	0.39	0.07
6.	($\overline{11}$ $\overline{26}$ 47), [$\overline{.87}$ $\overline{.50}$ $\overline{.03}$]	(111), [$\overline{121}$]	0.21	0.43	($\overline{111}$), [$\overline{121}$]	-0.44	0.09
7.	($\overline{11}$ $\overline{47}$ 26), [$\overline{.87}$ $\overline{.03}$ $\overline{.50}$]	(111), [$\overline{121}$]	0.48	0.43	($\overline{111}$), [$\overline{112}$]	0.39	0.05

Table 4.5 The Bogers-Burgers double shear system (1st shear-2nd shear) associated with N and K-S variant. Symbols are shown in Table 4.2.

	K-S and N Relationship Variant			Bogers-Burgers	Habit Plane
	Plane	N Direction	K-S Direction		
1-1	$(111)_\gamma // (011)_\alpha$	$[\bar{1}\bar{1}2]_\gamma // [0\bar{1}1]_\alpha$	$[\bar{1}01]_\gamma // [\bar{1}\bar{1}1]_\alpha$	C δ -B α	$(3 \ 15 \ 10)_\gamma$
1-2			$[0\bar{1}1]_\gamma // [1\bar{1}1]_\alpha$	C δ -A β	$(15 \ 3 \ 10)_\gamma$
1-3		$[2\bar{1}\bar{1}]_\gamma // [0\bar{1}1]_\alpha$	$[1\bar{1}0]_\gamma // [\bar{1}\bar{1}1]_\alpha$	B δ -A γ	$(10 \ 3 \ 15)_\gamma$
1-4			$[10\bar{1}]_\gamma // [1\bar{1}1]_\alpha$	B δ -C α	$(10 \ 15 \ 3)_\gamma$
1-5		$[\bar{1}2\bar{1}]_\gamma // [0\bar{1}1]_\alpha$	$[0\bar{1}\bar{1}]_\gamma // [\bar{1}\bar{1}1]_\alpha$	A δ -C β	$(15 \ 10 \ 3)_\gamma$
1-6			$[\bar{1}\bar{1}0]_\gamma // [1\bar{1}1]_\alpha$	A δ -B γ	$(3 \ 10 \ 15)_\gamma$
2-1	$(\bar{1}\bar{1}1)_\gamma // (011)_\alpha$	$[\bar{2}\bar{1}\bar{1}]_\gamma // [0\bar{1}1]_\alpha$	$[\bar{1}0\bar{1}]_\gamma // [\bar{1}\bar{1}1]_\alpha$	D β -A γ	$(\bar{1}0 \ 15 \ 3)_\gamma$
2-2			$[\bar{1}\bar{1}0]_\gamma // [1\bar{1}1]_\alpha$	D β -C α	$(\bar{1}0 \ 3 \ 15)_\gamma$
2-3		$[1\bar{1}2]_\gamma // [0\bar{1}1]_\alpha$	$[0\bar{1}1]_\gamma // [\bar{1}\bar{1}1]_\alpha$	A β -C δ	$(\bar{1}5 \ 3 \ 10)_\gamma$
2-4			$[101]_\gamma // [1\bar{1}1]_\alpha$	A β -D γ	$(\bar{3} \ 15 \ 10)_\gamma$
2-5		$[12\bar{1}]_\gamma // [0\bar{1}1]_\alpha$	$[110]_\gamma // [\bar{1}\bar{1}1]_\alpha$	C β -D α	$(\bar{3} \ 10 \ 15)_\gamma$
2-6			$[0\bar{1}\bar{1}]_\gamma // [1\bar{1}1]_\alpha$	C β -A δ	$(\bar{1}5 \ 10 \ 3)_\gamma$
3-1	$(\bar{1}\bar{1}1)_\gamma // (011)_\alpha$	$[112]_\gamma // [0\bar{1}1]_\alpha$	$[101]_\gamma // [\bar{1}\bar{1}1]_\alpha$	D γ -A β	$(\bar{3} \ \bar{1}5 \ 10)_\gamma$
3-2			$[011]_\gamma // [1\bar{1}1]_\alpha$	D γ -B α	$(\bar{1}5 \ \bar{3} \ 10)_\gamma$
3-3		$[\bar{2}1\bar{1}]_\gamma // [0\bar{1}1]_\alpha$	$[\bar{1}10]_\gamma // [\bar{1}\bar{1}1]_\alpha$	A γ -B δ	$(\bar{1}0 \ \bar{3} \ 15)_\gamma$
3-4			$[\bar{1}0\bar{1}]_\gamma // [1\bar{1}1]_\alpha$	A γ -D β	$(\bar{1}0 \ \bar{1}5 \ 3)_\gamma$
3-5		$[1\bar{2}\bar{1}]_\gamma // [0\bar{1}1]_\alpha$	$[0\bar{1}\bar{1}]_\gamma // [\bar{1}\bar{1}1]_\alpha$	B γ -D α	$(\bar{1}5 \ \bar{1}0 \ 3)_\gamma$
3-6			$[1\bar{1}0]_\gamma // [1\bar{1}1]_\alpha$	B γ -A δ	$(\bar{3} \ \bar{1}0 \ 15)_\gamma$
4-1	$(1\bar{1}1)_\gamma // (011)_\alpha$	$[21\bar{1}]_\gamma // [0\bar{1}1]_\alpha$	$[10\bar{1}]_\gamma // [\bar{1}\bar{1}1]_\alpha$	C α -B δ	$(10 \ \bar{1}5 \ 3)_\gamma$
4-2			$[110]_\gamma // [1\bar{1}1]_\alpha$	C α -D β	$(10 \ \bar{3} \ 15)_\gamma$
4-3		$[\bar{1}12]_\gamma // [011]_\alpha$	$[011]_\gamma // [\bar{1}\bar{1}1]_\alpha$	B α -D γ	$(15 \ \bar{3} \ 10)_\gamma$
4-4			$[\bar{1}01]_\gamma // [1\bar{1}1]_\alpha$	B α -C δ	$(3 \ \bar{1}5 \ 10)_\gamma$
4-5		$[\bar{1}\bar{2}\bar{1}]_\gamma // [0\bar{1}1]_\alpha$	$[\bar{1}\bar{1}0]_\gamma // [\bar{1}\bar{1}1]_\alpha$	D α -C β	$(3 \ \bar{1}0 \ 15)_\gamma$
4-6			$[0\bar{1}\bar{1}]_\gamma // [1\bar{1}1]_\alpha$	D α -B γ	$(15 \ \bar{1}0 \ 3)_\gamma$

small Schmid factors for the 1st shear. From these results, both the locally concentrated and applied stress is considered to assist the total shape strain of martensitic transformation rather than the 1st shear of the B-B double shears.

4.4.2 Martensite formed at the intersection of two deformation twins

As described in section 4.4.1, martensitic transformation was induced by the concentrated stress due to the impingement of deformation twins.

On the other hand, Olson and Cohen⁽⁴⁾ suggested that the intersection of shear bands (slip bands, ϵ' -bands and deformation twins etc.) could form embryo of martensite if the two shear systems corresponded to B-B double shears. Deformation twins are one of such shear bands, and when two deformation twins intersect, there is a possibility to operate O-C mechanism. In present study, the shear systems of two intersecting deformation twins were examined. The operating twinning systems are different between tensile deformation and compressive deformation, the examinations were performed both by tension and compression.

4.4.2.1 In the case of tensile deformation

Photo.4.6 shows an optical microstructure of Fe-25%Ni-0.7%C specimen deformed to 22% strain in tension at 243K. The martensite (M) is formed at the intersection of two deformation twins (A and B). The surface normal of austenite was determined from four twin traces

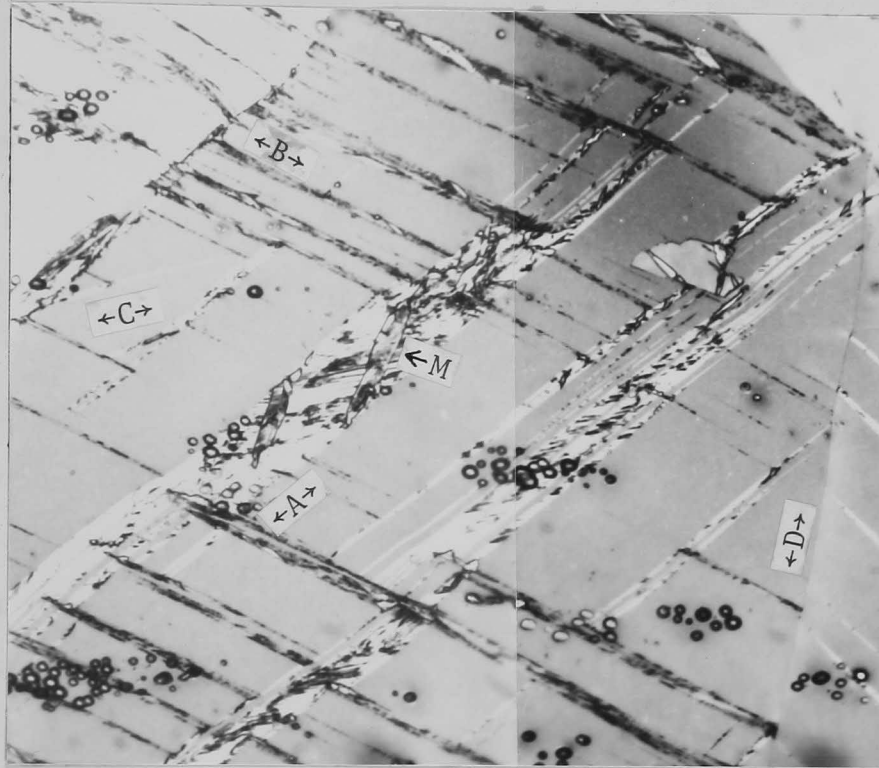


Photo.4.6 Microstructure of the Fe-25%Ni-0.7%C specimen which deformed to 22% strain in tension at 243K. Martensite(M) formed at the intersection of deformation twins(A) and (B). (x 700)

(A, B, C and D). The relation among the surface normal (P), the tensile direction (T) and four twin traces (A, B, C and D) is shown in the stereographic projection (Fig.4.6). The shear plane of A and B were determined as $(111)_\gamma$ and $(1\bar{1}1)_\gamma$, respectively. Calculated Schmid factor for each shear system is shown in Table 4.2. From Table 4.2, the shear systems of A and B are $(111)_\gamma$, $[11\bar{2}]_\gamma$ and $(1\bar{1}1)_\gamma$, $[\bar{2}11]_\gamma$, respectively. However, the combination of these two shear systems does not correspond to B-B double shears. If martensite which formed at the intersection of two deformation twins was formed by O-C mechanism, the combination of two intersecting twinning systems must correspond to B-B double shears. Martensites formed at the intersection of deformation twins in Photo.4.6 are not considered to be formed from an embryo suggested by Olson and Cohen because the shear systems of two intersecting deformation twins does not correspond to B-B double shears.

The other 5 martensite plates were analyzed in the same way. In all 5 cases, the combination of two shear systems of observed deformation twins did not correspond to B-B double shears. Then, it can be concluded that martensites which observed in this study at the intersections of two deformation twins were not considered to be formed by O-C mechanism.

Generally, the operating twinning system depends on stressing direction, and it is considered that the specific stressing direction is required to introduce two special twinning systems which correspond to B-B double shears. In the case of tension, the Schmid factor for these two twinning systems are positive for the tensile direction in the region of ADF in Fig.4.7. One of the Schmid factor for these two twinning systems is negative in the region of DFG.

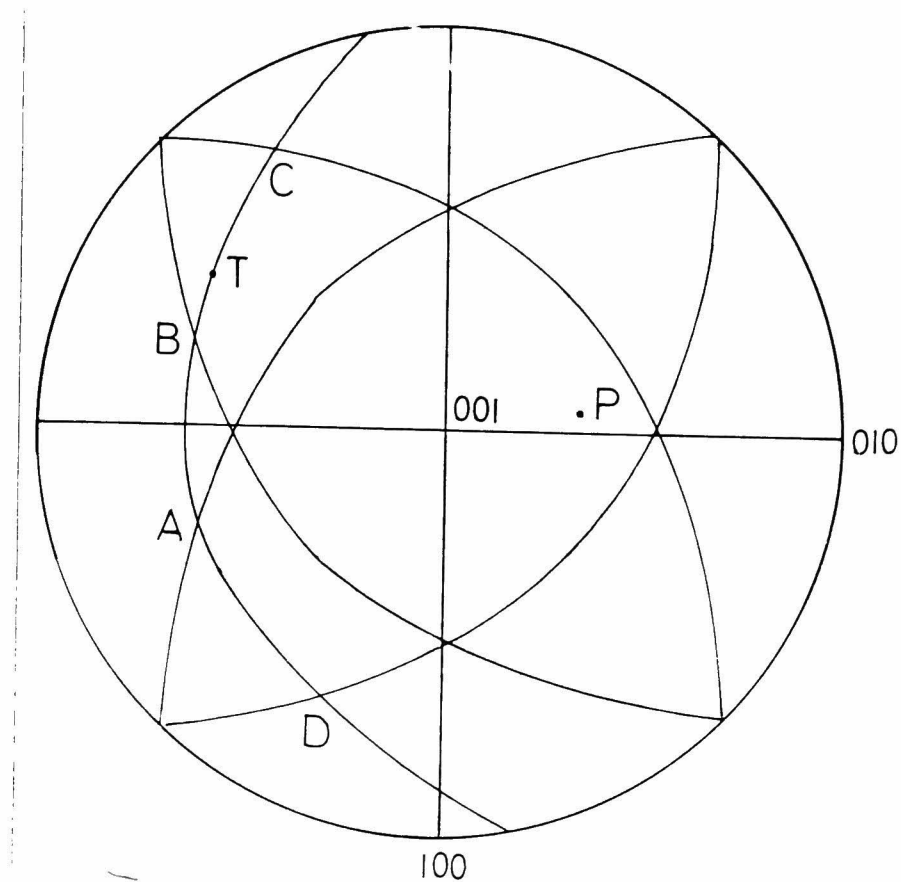


Fig.4.6 The stereographic projection showing the relation among the surface normal P, the tensile direction T and four directions of twin traces(A, B, C and D) of Photo.4.6.

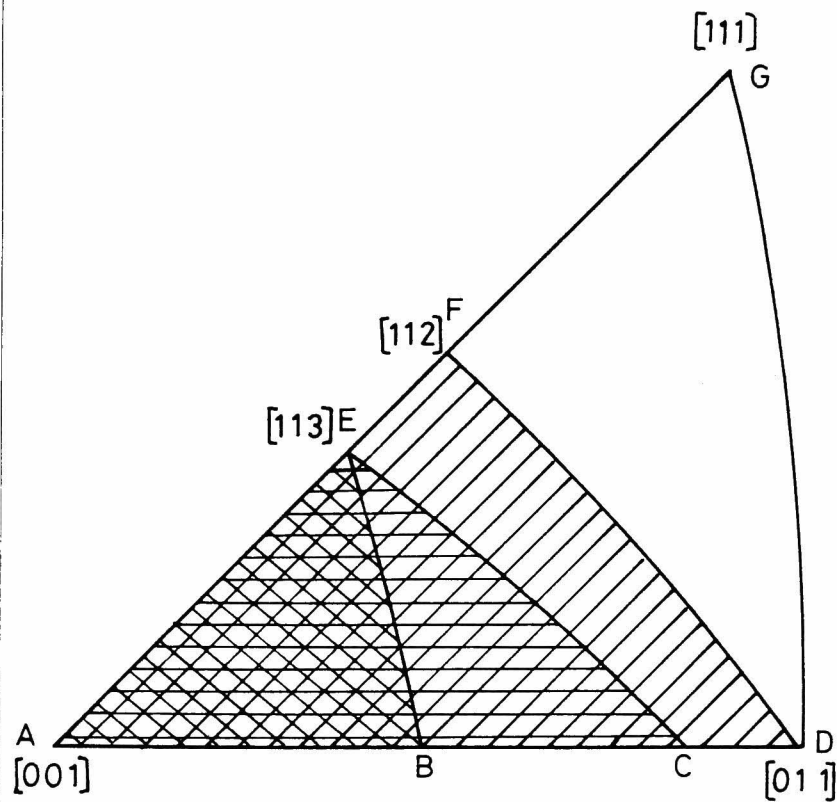


Fig.4.7 Stereographic unit triangle showing the region where two twinning system corresponding to the Bogers-Burgers double shears can be operated in tension.

ADF: where the Schmid factor for these two twinning systems are positive.

DFG: where one of the Schmid factor for these two twinning systems are negative.

ACE: where the Schmid factor for these two twinning systems are larger than 0.1.

AEB: where deformation twinning is difficult to occur in tension.

If the twinning system can operate when the Schmid factor for the system is positive, both of two twinning systems which correspond to B-B double shears can operate in the region of ADF. If the twinning system can operate when the Schmid factor is larger than 0.1, these two twinning systems can operate in the region of ACE which is smaller than ADF. The critical value of Schmid factor for twinning system operation is not clear. In the case of Fe-25%Ni-0.7%C alloy, critical twinning stress is about 500 MPa, and tensile strength is about 1000 MPa. A Schmid factor(μ) of operating twinning system must be larger than $(500/1000)\mu_m$ (μ_m is the maximum Schmid factor among 12 twinning systems). μ_m takes the minimum value(0.24) when tensile direction is $[001]\gamma$. Schmid factor of operating twinning system must be larger than $0.12 (= 0.24 \times 500 \div 1000)$. It is reasonable that twinning system, which has Schmid factor lower than 0.1, is difficult to operate in this alloy. There are some reports^{(8),(9)} describing that deformation twinning is difficult to occur in the region of ABE. From above consideration, the tensile direction which can introduce two twinning systems corresponding to B-B double shears is considered to be quite restricted.

4.4.2.2 In the case of compressive deformation

Operating twinning system depends on deformation method. In the case of compressive deformation, Narita and Takamura⁽¹⁰⁾ reported that deformation twinning occur only when the compressive direction was in the area ABE in Fig.4.7. In the area ABE, the combination of two twinning systems, each of which has the largest

Schmid factor and the second largest Schmid factor, correspond to B-B double shears. When two twinning systems operate in compression, the combination of the systems are considered to correspond to the B-B double shears. Photo.4.7 shows optical microstructures taken from the Fe-25%Ni-0.4%C($M_s = 223K$) specimen which deformed to 44% strain in compression at 333K. Martensite(M) is formed at the intersection of the grain boundary(GB) and deformation twins(DT)(Photo.4.7(a) and (b)) and within deformation twins(DT)(Photo.4.7(c)). On the other hand, martensite which can be identified clearly is not observed at the intersection of two deformation twins(Photo.4.7(b) and (d)). As described in Chapter III, there is a possibility that the small dark etched spots in deformation twins are martensite. However, martensite, which can be identified clearly, is not tend to form at the intersection of two deformation twins preferentially. The combination of such two deformation twinning systems are considered to correspond to the B-B double shears. Then, from the results in section 4.4.2.1 and above, it can be concluded that the strain-induced embryo formation mechanism which proposed by Olson and Cohen is not prevailing mechanism for deformation-induced martensitic transformation.

It is considered that stress is concentrated at the intersection when deformation twins impinged on another deformation twins(since twinning dislocation can not move two different twinning planes). However, martensite was not tend to form at the intersection of two deformation twins. When impinging deformation twin is thin, the concentrated stress at the intersection of two deformation twins is considered to be small. In Photo.4.7(b) and

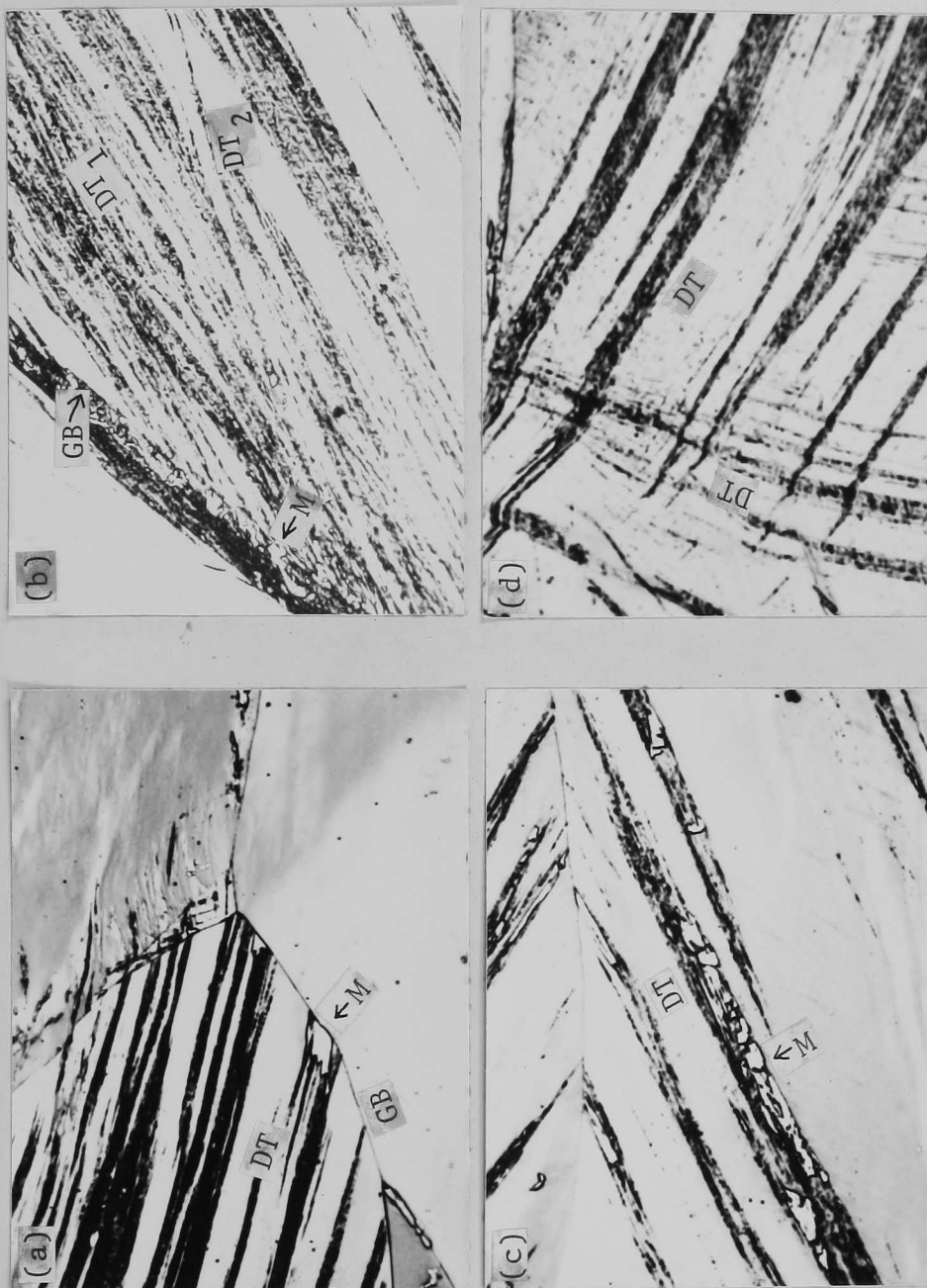


Photo.4.7 Optical microstructures of the Fe-25%Ni-0.4%C specimen which deformed to 44% strain in compression at 333K.
 (a), (b): martensite (M) which formed at the grain boundary (GB) on which deformation twins (DT) impinged.
 (c): martensite (M) which formed inside the deformation twin (DT).
 (d): intersections of two deformation twins (DT). (x 700)

(d), one of two intersecting deformation twins is a group of thin twins. It is considered that the concentrated stress at the intersection of two deformation twins is smaller than that at the grain boundary(Photo.4.7(a) and (b)) on which deformation twins impinged.

4.5 Conclusion

In an Fe-25%Ni-0.7%C austenite and an Fe-25%Ni-0.4%C austenite, both twinning and martensitic transformation can be induced during deformation in a suitable temperature range. In the present study, the effect of deformation twin on the deformation-induced martensitic transformation was examined to reveal the effect of applied stress and plastic strain on the martensitic transformation. The results obtained are as follows.

(1) In the temperature range between M_S and M_S^σ , the critical stress for martensite formation increased linearly with increase in temperature. In the temperature range above M_S^σ , the critical stress for martensite formation was considerably lower than that expected by extrapolating the stress-temperature line in the temperature range between M_S and M_S^σ .

(2) In the temperature range between M_S and M_S^σ , the morphology of martensite was the same as the one of thermally transformed(i.e., lenticular). In the temperature range above M_S^σ , the martensite was frequently observed at the intersection of two deformation twins, and the grain boundaries or the annealing twin boundaries on which deformation twins impinged.

(3) Such martensites had large N_{ij} value for the impinging twinning system and the shear of martensitic transformation on the habit plane. And they also had large Schmid factor for the direction of applied tensile stress. It was concluded that such martensite were formed by the aid of the concentrated stress due to the impingement of deformation twins plus by the aid of applied tensile stress.

(4) Martensite which observed at the junction of two deformation twins formed in tension in this study was considered not to be formed from the embryo produced by the intersection of two shear bands suggested by Olson and Cohen. From the crystallographic analysis, the combination of these two deformation twins did not correspond to Bogers and Burgers double shears.

(5) It was shown that in the case of tension, the tensile direction which can introduce two deformation twinning systems corresponding to B-B double shears was considered to be quite restricted. In the case of compression, martensite which can be identified clearly is not tend to form preferentially at the intersection of two deformation twins that are considered to be corresponding to B-B double shears. From the results (4) and (5), it was considered that the strain-induced embryo formation mechanism which proposed by Olson and Cohen is not prevailing mechanism for deformation-induced martensitic transformation.

Appendix

First, each K-S variant was associated with each N variant. In Table 4.5, the associated K-S and N variant has the nearest orientation. Next, each B-B variant was associated with each K-S variant. For example, $C\delta - B\alpha$ (symbols are shown in Table 4.2) variant is associated with corresponding K-S variant.

In Fig.A.1, the process of $C\delta - B\alpha$ shear mechanism is shown. During the 1st shear ($C\delta : (a_\gamma/18)[11\bar{2}]_\gamma, (111)_\gamma$), only the $(111)_\gamma$ plane is the invariant plane among four $\{111\}_\gamma$ planes. Then, $(111)_\gamma$ plane was fixed during the transformation. After the transformation, each $\{111\}_\gamma$ plane becomes $\{011\}_\alpha$ plane. Then, the following relationship is obtained.

$$(111)_\gamma // (011)_\alpha$$

During the transformation, $[\bar{1}01]_\gamma$ is invariant direction by shear of both $C\delta$ and $B\alpha$. After the transformation, $[\bar{1}01]_\gamma$ becomes $[\bar{1}\bar{1}1]_\alpha$ or $[1\bar{1}1]_\alpha$ direction on the $(011)_\alpha$ plane. After the transformation, $[\bar{1}01]_\gamma$ becomes (a) direction on $(011)_\alpha$ plane as shown in Fig.A.1. (a) direction and (b) direction makes 70° angle after the transformation, then (a) direction is decided as $[\bar{1}\bar{1}1]_\alpha$ direction. Then, the following orientation relationship is obtained.

$$(111)_\gamma // (011)_\alpha$$

$$[\bar{1}01]_\gamma // [\bar{1}\bar{1}1]_\alpha.$$

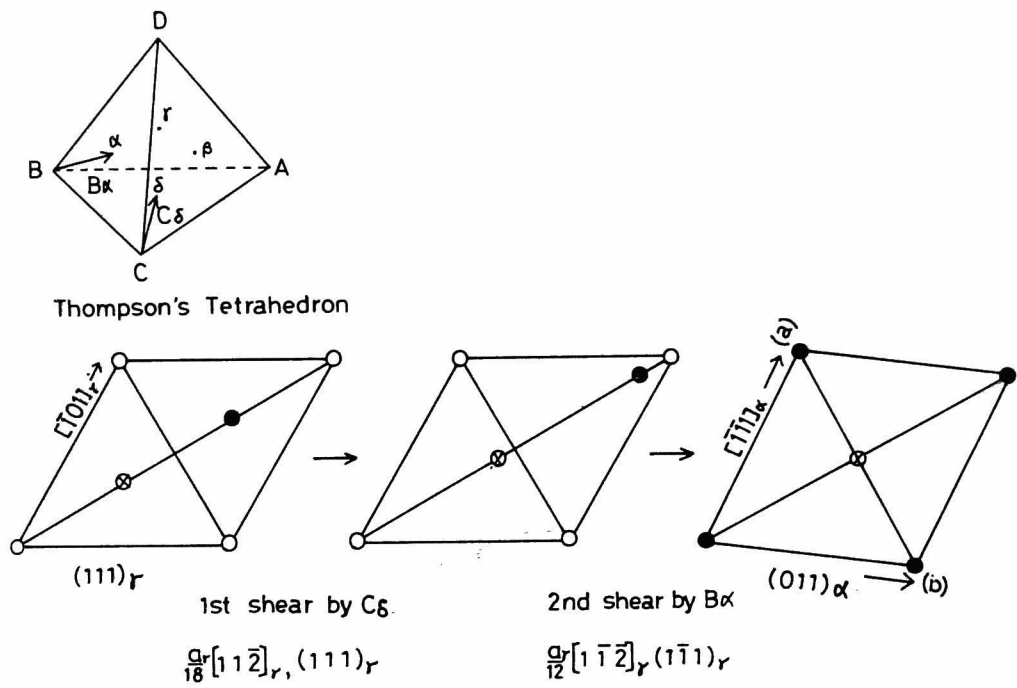


Fig.A.1 Double shear mechanism by Bogers and Burgers⁽⁶⁾.

References for Chapter IV

- 1) G. F. Bolling and R. H. Richman: *Phil. Mag.*, 19(1969), 247.
- 2) T. Suzuki, H. Kojima, K. Suzuki, T. Hashimoto and M. Ichihara: *Acta Met.*, 25(1977), 1162.
- 3) S. Ueda and H. Fujita: *J. Japan Institute of Metals*, 41(1977), 1007.
- 4) G. B. Olson and M. Cohen: *J. Less-Common Metals*, 28(1972), 107.
- 5) J. A. Venables: *Phil. Mag.*, 7(1962), 35.
- 6) A. J. Bogers and W. G. Burgers: *Acta Met.*, 12(1964), 255.
- 7) F. Lecroisey and A. Pineau: *Met. Trans.*, 3(1973), 387.
- 8) T. H. Blewitt, R. R. Coltman and J. K. Redman: *J. Appl. Phys.*, 28(1957), 651.
- 9) S. Miura, J. Takamura and N. Narita: *Trans., JIM*, 9(1968), Supplement 555.
- 10) N. Narita and J. Takamura: *Scripta Met.*, 9(1975), 819.
- 11) Y. Shugo, K. Hashizume, S. Kato and T. Honma: *Research Institut of Mineral Dressing and Metallurgy Report, Tohoku University*, 32(1976), 109.
- 12) A. B. Greninger and A. R. Troiano: *Trans. AIME*, 145(1941), 289; 185(1949), 590.
- 13) J. F. Breedis and W. D. Robertson: *Acta Met.*, 11(1963), 547.
- 14) Y. Higo, T. Mori and T. Nakamura: *Trans. JIM*, 11(1971), 1176.
- 15) M. Tokizane: *Scripta Met.*, 10(1976), 459.
- 16) P. C. Maxwell, A. Goldberg and J. C. Shyne: *Met. Trans.*, 5(1974), 1305.
- 17) J. D. Livingston and B. Chalmers: *Acta Met.*, 5(1957), 322.
- 18) M. S. Wechsler, D. S. Lieberman and T. A. Read: *Trans. AIME*, 197(1953), 1503.
- 19) M. Kato, R. Monzen and T. Mori: *Acta Met.*, 26(1978), 605.
- 20) M. Kato and T. Mori: *Acta Met.*, 24(1976), 853.

CHAPTER V EFFECT OF PRIOR DEFORMATION ON THE MARTENSITIC TRANSFORMATION

5.1 Introduction

When martensitic transformation is induced during deformation, it is difficult to separate the effect of deformation into the total applied stress and local concentrated stress. In Chapter IV, it was observed that during deformation, concentrated stress formed by the impingement of deformation twins on the grain boundaries or the annealing twin boundaries induced martensitic transformation. From that, it was suggested that the effect of plastic strain is to form the stress concentrated regions during deformation. However, when the martensite is induced during deformation, the effect of total applied stress was also acting. To separate the effect of local concentrated stress from the effect of applied stress on the martensitic transformation, it seems to be important to examine the effect of prior deformation on the subsequent thermal martensitic transformation.

In the former studies^{(1)~(7)} on the effect of prior deformation on the martensitic transformation, as shown in Chapter I, the M_s temperature increased with increasing degree of prior deformation, and then decreased by further prior deformation. In these studies, it was suggested that the local residual stress, which was introduced by prior deformation, promoted the transformation. However, the defects which were introduced in these studies were mainly dislocations, and microstructural considerations of the effects of these defects were not made enough from its experimental difficulties.

Since it has not been made clear where martensitic transformation initiated and how the local residual stress acts to induce the transformation, the relation between the local residual stress and formation of martensite has not been made clear.

In Chapter III, it was found that deformation twinning could occur in austenite of Fe-Ni-C alloys. Deformation twin can be observed by optical microscope, and it is the most suitable defects to examine the effect of prior deformation on the martensitic transformation. In this study, the effect of deformation twins which was introduced by prior deformation above the M_d temperature on the martensitic transformation was examined using an Fe-25%Ni-0.7%C alloy. An Fe-32%Ni alloy, in which deformation twinning hardly occurs during prior deformation, was also used. The results were compared with the results obtained for the Fe-25%Ni-0.7%C alloy.

From this study, it was found that martensites formed preferentially at the grain boundaries or the annealing twin boundaries on which deformation twins (that were introduced by prior deformation) impinged. From the crystallographic analysis, it was concluded that such martensites were formed by the aid of residual local stress at the boundaries on which deformation twins impinged. Furthermore, since such local residual stress is formed during deformation, the result in this study supports the result in Chapter IV.

5.2 Experimental Procedure

An Fe-25%Ni-0.7%C alloy and an Fe-32%Ni alloy were prepared by vacuum induction melting. Chemical compositions and M_s and M_d temperatures are shown in Table 5.1. The ingots of alloys were

Table 5.1 Chemical compositions (wt%) and M_s and M_d temperatures (K) of alloys used.

Alloy	C	Si	Mn	P	S	Ni	Al	Fe	M_s	M_d
Fe-25%Ni-0.7%C	0.65	0.067	0.17	0.010	0.009	25.31	-	bal.	163	253
Fe-32%Ni	<0.005	0.006	0.051	<0.005	0.004	32.18	0.094	bal.	190	<298

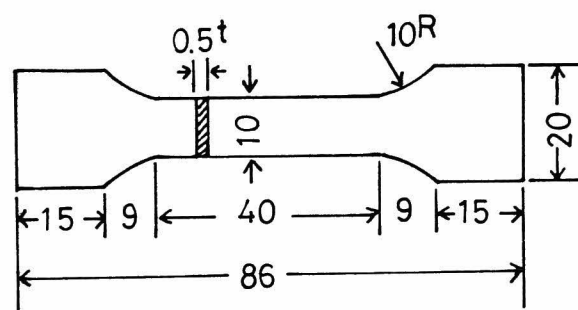


Fig. 5.1 Shape and size of specimen (mm).

hot-rolled at 1473K and homogenized in a vacuum furnace at 1473K for 24h, and then cold-rolled to make sheets of 0.8 mm thickness. Specimens for tensile deformation as shown in Fig.5.1 were machined from the sheets. Specimens were austenitized at 1423K for 1h in a vacuum furnace and were quenched in iced brine(I.B.Q.) for both alloys. Some of specimens of the Fe-32%Ni alloy were cooled in furnace(F.C.) from the austenitizing temperature to avoid the thermal stress during quenching which was considered to be affected M_S temperature of the alloy. The result obtained for furnace cooled specimens was compared with those obtained for iced brine quenched specimen.

Each specimen was pre-strained at 298K(above the M_d temperature for both alloys) to various strain in tension. In the specimens of the Fe-25%Ni-0.7%C alloy, deformation at 298K proceeded only by slip to about 15% strain and by both slip and deformation twinning above 15% strain. In the case of the Fe-32%Ni, deformation proceeded only by slip.

Deformed specimens were cut to three pieces to measure the M_S temperature. The M_S temperature was measured by recording the change in electrical resistance during continuous cooling of the specimen between room temperature and the temperature of a liquid nitrogen.

Optical microscopic observation was performed on the specimen which formed a few martensite plates after cooling below the M_S temperature to examine where martensites formed. 35%Na₂S₂O₅ was used as an etching solution.

The variant of martensite plate formed by cooling below the M_S temperature was determined by the method described in 4.2.2

(Chapter IV). The morphologies of martensite formed in the Fe-25%Ni 0.7%C alloy were lenticular or thin plate. The habit plane was reported as $\{3\ 15\ 10\}_\gamma$ for both martensites^{(8),(9)}. Maki et al⁽¹⁰⁾ reported that the morphology of martensite depends mainly on the formation temperature and 170K is the temperature at which both morphologies can be formed.

5.3 Results

5.3.1 Effect of prior deformation on the M_s temperature

Typical electrical resistance vs temperature curves of initial deformed specimens are shown in Fig.5.2(a), (b) and (c). In the specimen of the Fe-25%Ni-0.7%C alloy deformed to small degree at 298K, the burst transformation occurred and the M_s was clearly detected as shown in Fig.5.2(a). While in the specimen of the Fe-25%Ni-0.7%C alloy which was initially deformed to a certain degree, the burst was suppressed, and the M_s was characterized by a relatively small change in slope of the curve as shown in Fig.5.2(b). In the case of the Fe-32%Ni alloy, large bursts were observed for all specimens such as shown in Fig.5.2(c).

Fig.5.3 ~ Fig.5.5 show the variation of the M_s temperature with increasing degree of prior deformation. For the Fe-25%Ni-0.7%C alloy(I.B.Q.)(Fig.5.3), the M_s first rises up to about 10% strain and then falls with increasing degree of prior deformation. The maximum increase of the M_s by prior deformation is about 30K. When austenite is plastically deformed, the regions which have local

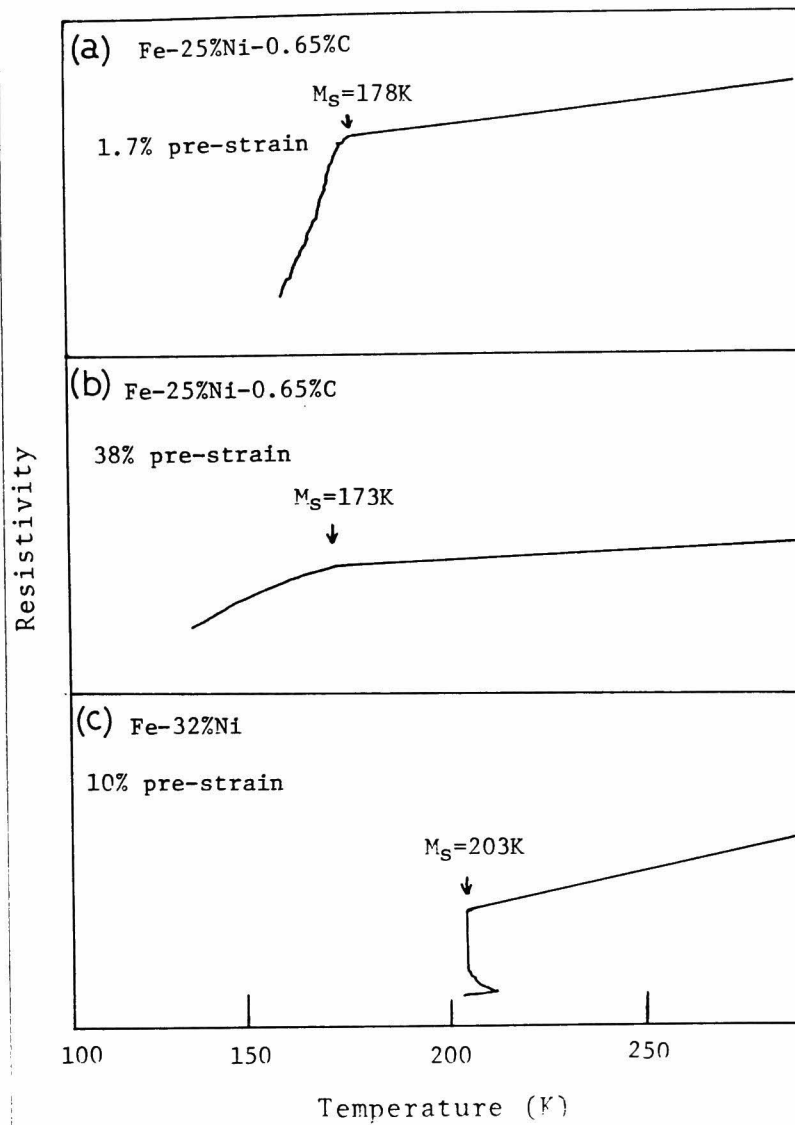


Fig.5.2 Typical transformation curves.

- (a): For the Fe-25%Ni-0.7%C alloy which was pre-strained 1.7% at 298K. Bursts are detected.
- (b): For the Fe-25%Ni-0.7%C alloy which was pre-strained 38% at 298K. Small change in slope is observed at M_S .
- (c): For the Fe-32%Ni alloy which was pre-strained 10% at 298K. Large burst is observed.

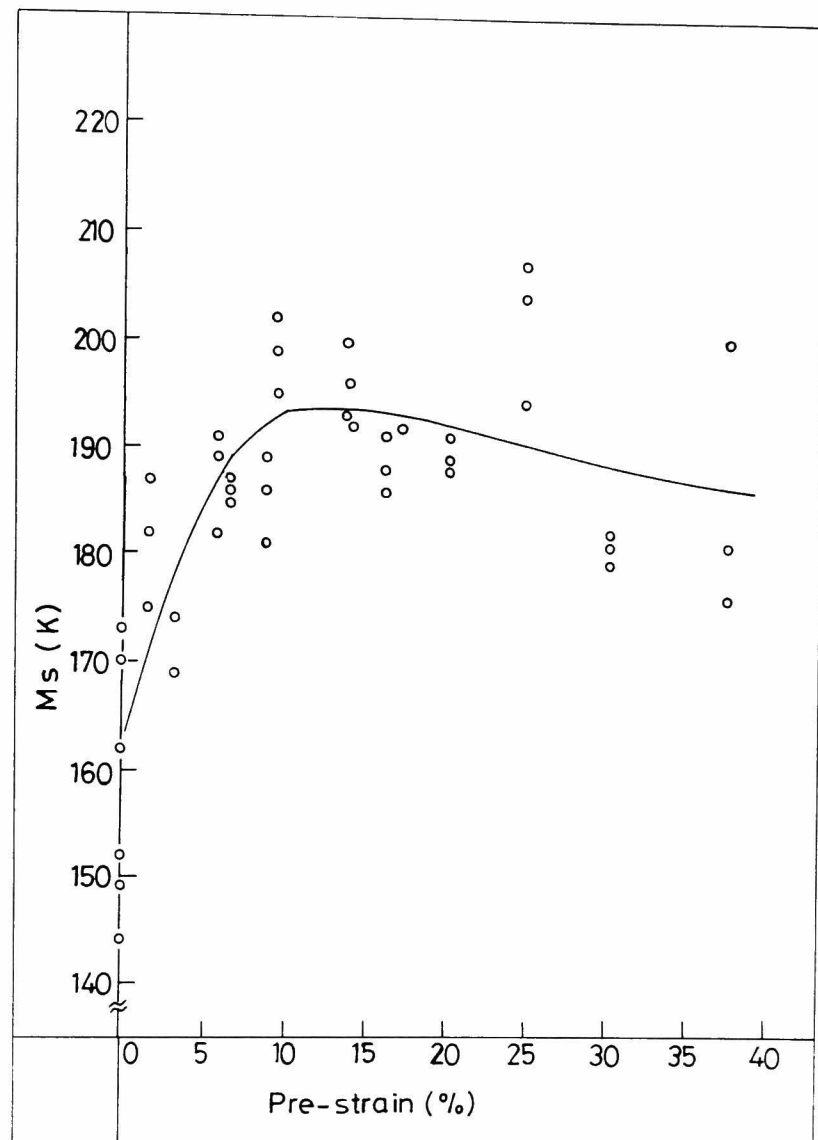


Fig.5.3 Variation of the M_s temperature with increasing degree of prior deformation for the Fe-25%Ni-0.7%C specimen which was quenched in iced brine(I. B. Q.) from austenitizing temperature.

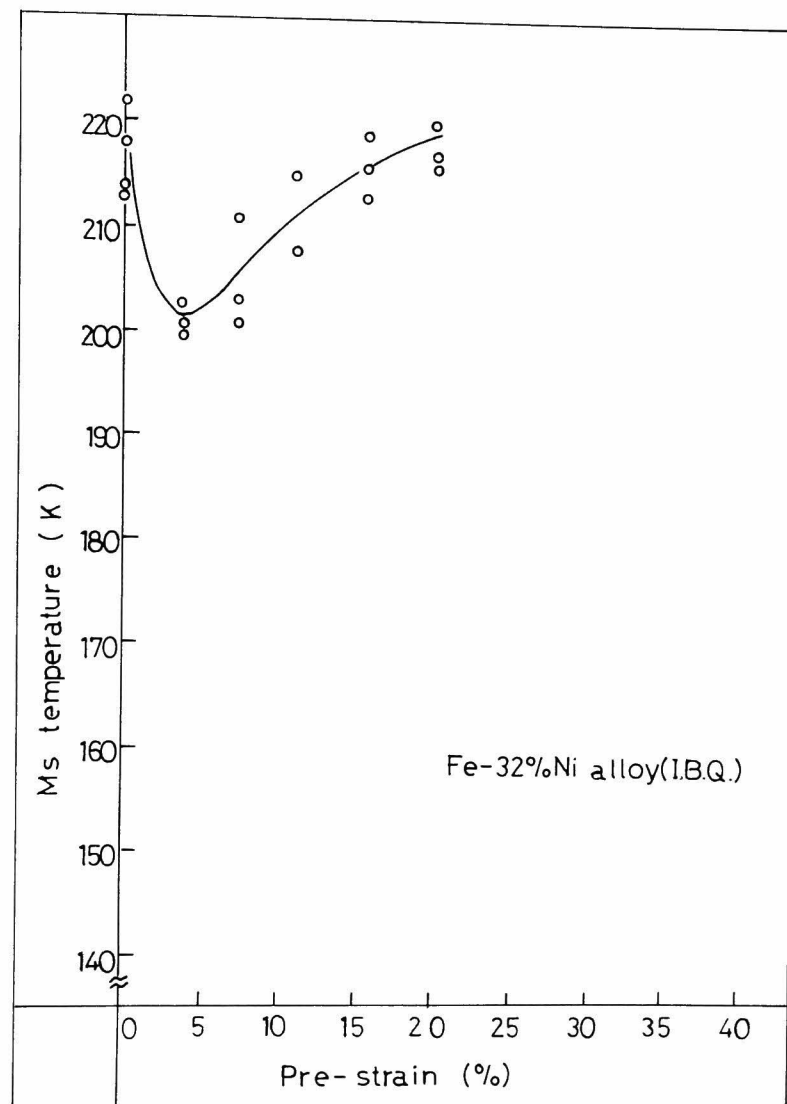


Fig.5.4 Variation of the M_s temperature with increasing degree of prior deformation for the Fe-32%Ni specimen which was quenched in iced brine(I.B.Q.) from austenitizing temperature.

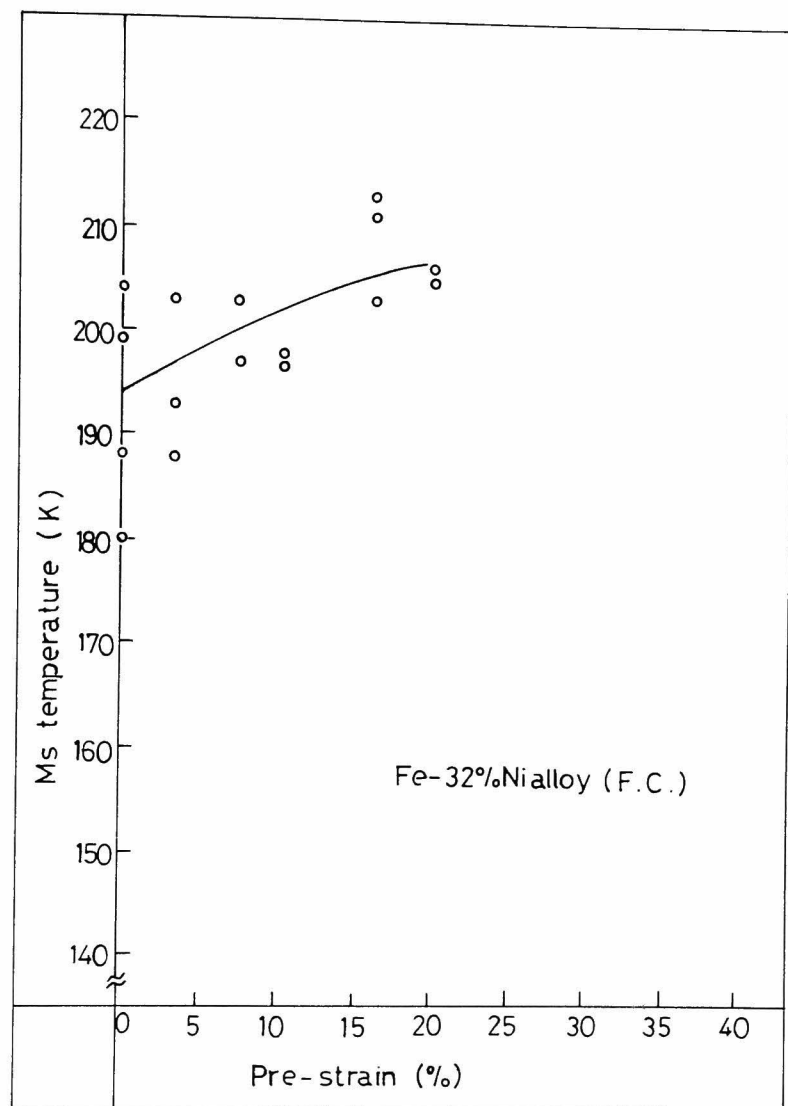


Fig.5.5 Variation of the M_s temperature with increasing degree of prior deformation for the Fe-32%Ni specimen which was cooled in furnace(F. C.) from austenitizing temperature.

high concentrated stress are introduced. Such local stress is considered to aid the martensitic transformation. Plastic deformation also results in the hardening of the matrix austenite. If the austenite is work-hardened, the M_s falls^{(6),(11)}. The results in Fig.5.3 can be explained as the total of both effects. For the specimens of the Fe-32%Ni alloy which were quenched in iced brine(Fig.5.4), the M_s first decreased to about 5% pre-strain and then increased with increasing pre-strain. Such decrease of the M_s at small pre-strain was not observed for the specimens of the Fe-32%Ni alloy which was cooled in furnace(Fig.5.5). The decrease of the M_s in Fig.5.4 was considered to be due to the thermal residual stress which was created by quenching from austenitizing temperature. It was considered that the M_s of the Fe-32%Ni alloy(I. B. Q.) was high in the quenched state due to the thermal residual stress created by quenching from austenitizing temperature, and the M_s was decreased with prior deformation by eliminating the residual stress. Such decrease of the M_s at small strain was not observed for the Fe-25%Ni-0.7%C alloy. It was considered that the thermal residual stress due to quenching was smaller in the Fe-Ni-C alloy than in the Fe-Ni alloy, because the yield strength of the Fe-Ni-C alloy is higher than that of the Fe-Ni alloy.

Fig.5.6 shows true stress-strain curves for the Fe-25%Ni-0.7%C alloy and the Fe-32%Ni alloy at 298K(above M_d for both alloys). For the Fe-25%Ni-0.7%C alloy, deformation proceeds by slip up to 15% strain and deformation twins begins to form from about 15% strain. For the Fe-32%Ni alloy, deformation proceeds only by slip. The serration, which was observed in the Fe-25%Ni-0.7%C alloy, was due to P-L effect due to carbon atom because the serration was

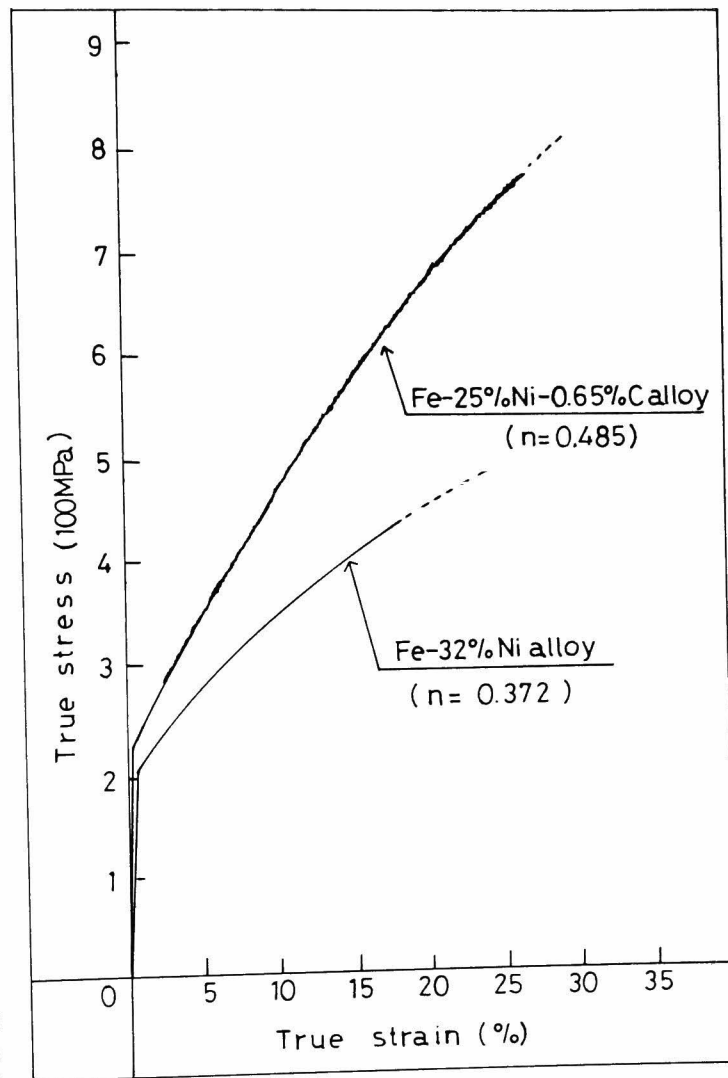


Fig.5.6 True stress-true strain curves for the Fe-25%Ni-0.7%C specimen and the Fe-32%Ni specimen at 298K.

disappeared when the strain rate was changed. At 298K, the equation of $\sigma = K\epsilon^n$ (where σ is true stress, ϵ is true strain, n is work-hardening exponent and K is constant) could be applied to fit the stress-strain curves for both alloys. When deformation twinning occurs, the value of n changes. In the case of Fe-25%Ni-0.7%C alloy it was considered that the equation can be applied since amount of deformation twin was small. The work-hardening exponent(n) is 0.49 for the Fe-25%Ni-0.7%C alloy and 0.37 for the Fe-32%Ni alloy, respectively.

The lowering of the M_s observed for the Fe-25%Ni-0.7%C alloy above 10% strain(Fig.5.3) was not observed for the Fe-32%Ni alloy (Fig.5.4 and Fig.5.5). The lack of such lowering of the M_s for the Fe-32%Ni alloy seems to be due to the smaller work-hardening($n=0.37$) comparing with the Fe-25%Ni-0.7%C alloy($n=0.49$).

5.3.2 Optical microstructure observation

The optical microscopic observation was performed on the Fe-25%Ni-0.7%C specimen cooled just below the M_s to reveal the effect of deformation twins of austenite on the martensitic transformation.

Photo.5.1(a), (b) and (c) show optical microstructures of the Fe-25%Ni-0.7%C specimen which deformed to 30% strain at 298K and then cooled to 130K. Martensite(M) are frequently observed at the grain boundry(GB)(Photo.5.1(a)), and annealing twin boundary(AT) (Photo.5.1(b)) deformation twins(DT) impinged, and at the junction of two deformation twins(Photo.5.1(c)). From these observations, it was found that martensite was tend to form at the region of high local stress created by the impingement of deformation twins.



Photo.5.1 Microstructures of the Fe-25%Ni-0.7%C specimen which was deformed to 30% strain at 298K and cooled to 130K.

- (a) Martensite(M) which formed from the grain boundary(GB) on which deformation twins(DT) impinged.
- (b) Martensite(M) which formed from the annealing twin boundary(AT) on which deformation twins(DT) impinged.
- (c) Martensite(M) which formed from the junction of two deformation twins(DT). (x 700)

Photo.5.2(a) shows optical microstructure of the Fe-25%Ni-0.7%C specimen which was deformed to 30% strain at 298K which showing deformation twins of austenite. After the deformation, specimen was cooled to 120K just below the M_s to form small amount of martensite. Photo.5.2(b) shows optical microstructure of the same region as Photo.5.2(a) after cooled to 120K. Martensite(A) and (B) seems to be formed from the grain boundary(GB) on which deformation twins(DT;A) and (DT;B) impinged, respectively. Martensite(C) and (D) are considered to be formed from the annealing twin boundary(AT) on which deformation twins(DT;C) impinged.

From these observations, it can be concluded that local residual stress at the grain boundaries or the annealing twin boundaries on which deformation twins impinged is quite effective to induce the martensitic transformation.

5.4 Discussion

The concentrated stress at the intersection of grain boundary or annealing twin boundary and deformation twins which is created during deformation can not be released completely by unloading. The direction of such residual shear stress is fixed by the shear system of impinging twin, and the variants of martensites which form by the aid of such residual stress will be restricted by the direction of residual shear stress. Here, the relation between the variant of martensite and the direction of residual shear stress was examined.

Photo.5.3 shows an optical microstructure which was deformed to 30% strain in tension at 298K and then cooled to 130K. Martensites (M_1 , M_2 and M_3) are formed at the intersections of the annealing

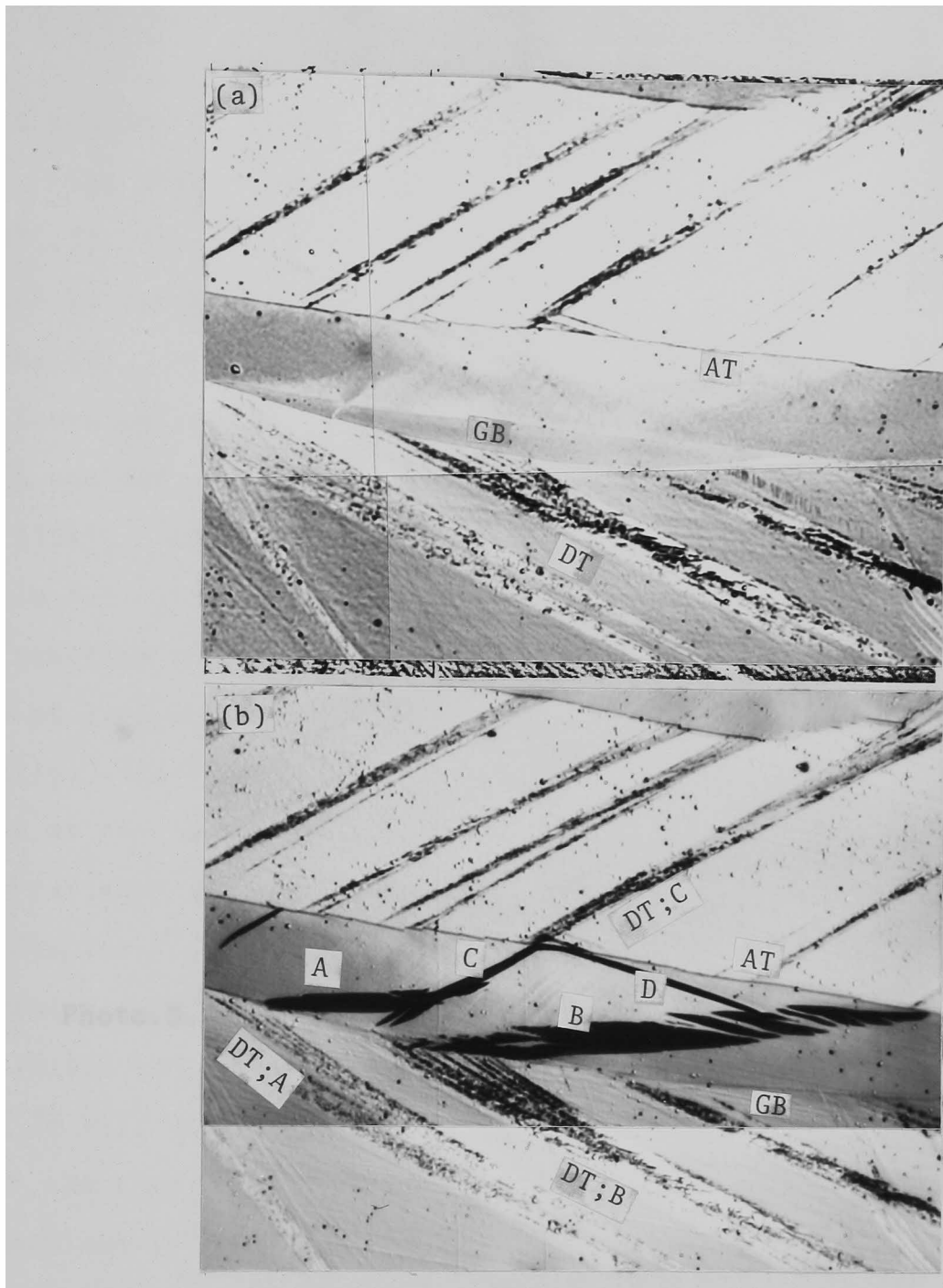


Photo.5.2 Microstructures of the Fe-25%Ni-0.7%C specimen showing the nucleation site for martensite. (x 700)

(a) Microstructure of the austenitic specimen which deformed to 30% strain at 298K.

(b) Microstructure of the same area as (a) which was cooled to 120K.



Photo.5.3 Microstructure of the Fe-25%Ni-0.7%C specimen which was deformed to 30% strain at 298K and cooled to 130K. Martensite(M₁, M₂ and M₃) are considered to be formed from the annealing twin boundary(AT) on which deformation twins(DT) impinged. (x 930)

twin boundary(AT) and deformation twins(DT).

The surface normal of the specimen was calculated as $[\overline{41} \ .58 \ .71]_{\gamma}$ by Eq.4.1 using four twin traces(A, B, C and the other which is not seen in Photo.5.3). The relation among the plane normal(P), tensile direction(T) and directions of four twin traces is shown in Fig.5.7. From Fig.5.7, twin plane of the annealing twin(AT) was determined as $(\overline{111})_{\gamma}$, and that of the deformation twins(DT) as $(111)_{\gamma}$. Schmid factor for each twinning system is listed in Table 5.2. From Table 5.2, the twinning direction of impinging deformation twins(DT) was determined as $[1\overline{2}1]_{\gamma}$ on the $(111)_{\gamma}$ as the one of largest Schmid factor(0.32) among three directions on the $(111)_{\gamma}$. The direction of residual shear stress was assumed the same as the twinning direction of the impinging deformation twins. The variants of martensites(M_1 , M_2 and M_3) were measured as described in Chapter IV(4.2.2). Photo.5.4 shows microstructures of two surfaces. Fig.5.8 shows two surface analysis in the case of M_1 in Photo.5.3 and 5.4. The habit plane of (M_1) is determined as $(\overline{11} \ 2\overline{6} \ 47)_{\gamma}$ which is nearest to (H) among 24 variants in Fig.5.8. From the same analysis, the habit planes of (M_2 and M_3) were determined as $(\overline{11} \ 4\overline{7} \ 26)_{\gamma}$ and $(\overline{1} \ 14 \ 5\overline{3})_{\gamma}$, respectively. Since martensites were formed inside $(\overline{111})_{\gamma}$ annealing twin(AT), $\{3 \ 15 \ 10\}_{\gamma}$ habit planes were transfered to indices of matrix by equation(5.1).

$$(h \ k \ 1)_M = (1/3)(h \ k \ 1)_T \begin{vmatrix} \overline{1} & 2 & \overline{2} \\ 2 & \overline{1} & \overline{2} \\ \overline{2} & \overline{2} & \overline{1} \end{vmatrix} \quad \text{-----(5.1)}$$

To study the relation between the residual concentrated stress formed by the impingement of deformation twins and the martensitic

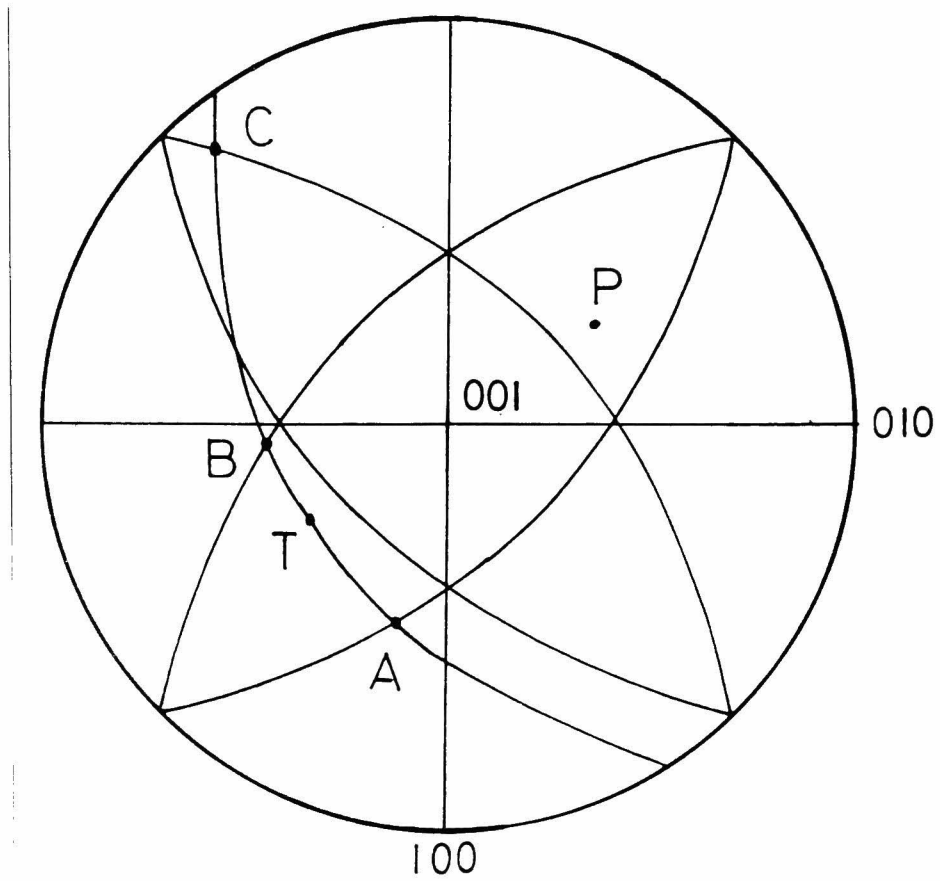


Fig.5.7 The stereographic projection showing the surface normal P, tensile direction T and three twin traces(A, B and C) of Photo.5.3.

Table 5.2 Calculation of the Schmid factor for the shear system $\langle 112 \rangle_r$, $\{111\}_r$ when the tensile direction is $[\frac{1}{2} \frac{\sqrt{3}}{2} \frac{1}{2}]_r$ (Photo.5.3).

Shear Plane	Shear Direction	Schmid Factor
(111)	$[11\bar{2}]$	-0.22
	$[1\bar{2}1]$	0.32
	$[\bar{2}11]$	-0.11
$(\bar{1}11)$	$[\bar{1}\bar{2}1]$	-0.09
	$[211]$	-0.07
	$[\bar{1}1\bar{2}]$	0.17
$(1\bar{1}1)$	$[\bar{2}\bar{1}1]$	0.14
	$[1\bar{1}\bar{2}]$	-0.17
	$[121]$	0.03
$(\bar{1}\bar{1}1)$	$[2\bar{1}1]$	0.41
	$[\bar{1}\bar{1}\bar{2}]$	-0.25
	$[\bar{1}21]$	-0.16

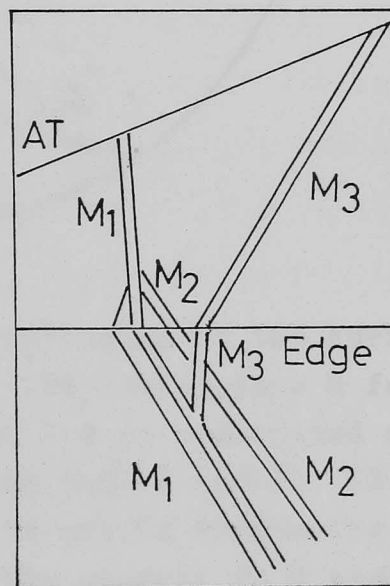


Photo.5.4 Two surface analysis for Photo.5.3.

Habit planes of martensites (M_1 , M_2 and M_3)
were determined by two surface analysis. (x 700)

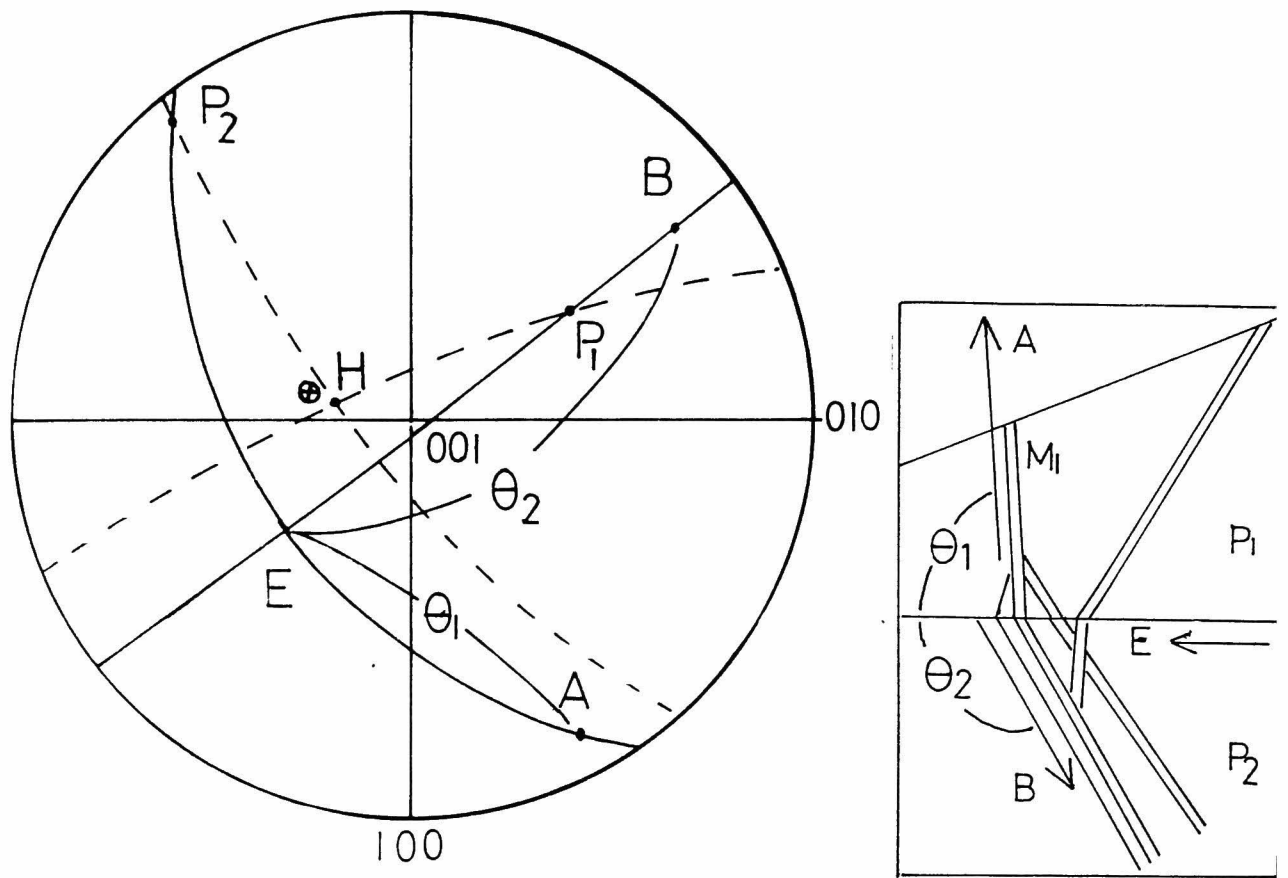


Fig.5.8 Stereographic projection of two surface analysis for Photo.5.3. The habit plane H for martensite(M_1) in Photo.5.3 and 5.4 is determined as the point which is 90° from both A and B. $\{3\ 15\ 10\}_\gamma$ planes are transferred to matrix indices by Eq.5.1. $(\overline{11}\ \overline{26}\ 47)_\gamma$ is the nearest to H among them, and is selected as habit plane of martensite(M_1).
A,B: direction of the trace of habit plane(M_1) on each plane.
E: edge direction.
 P_1, P_2 : direction of the surface normal of each surface.

transformation, N_{ij} value (stress coupling factor, Eq.4.2) between the impinging twinning system $[1\bar{2}1]_{\gamma}$, $(111)_{\gamma}$ and the shear of martensitic transformation on the habit plane was calculated for 24 martensite variants. The calculated results are listed in Table 5.3. The shear system $\langle .\bar{2}4 .60 .77 \rangle_{\gamma}$, $\{3 15 10\}_{\gamma}$ was used for martensitic transformation which was calculated by W-L-R⁽¹²⁾. The shear system was also transferred to the indices of matrix by Eq.5.1.

In Table 5.3, martensite variants are listed in order from the larger value of N_{ij} . Observed martensite variants marked by -o have relatively large N_{ij} values. The same analysis as above were performed on other three martensite plates. The results are listed in Table 5.4. In Table 5.3 and 5.4, all the observed variants have large N_{ij} values. From these analysis, it can be concluded that martensite was formed by the aid of the residual shear stress created by the impingement of deformation twins.

Such local residual stress can be also introduced by piled-up dislocations at the grain boundaries or other obstacles, and the residual stress is considered to have the same contribution to the martensitic transformation as the residual stress formed by the impingement of deformation twins. The increase of the M_s temperature which observed in the case of the Fe-32%Ni alloy is considered to be due to the residual stress formed by piled-up dislocations.

5.5 Conclusion

To separate the effect of local concentrated stress from the effect of total applied stress on the martensitic transformation, the effect of prior deformation on the martensitic transformation

Table 5.3 N_{ij} value for 24 martensite variants
in the case of Photo.5.3.

-o: observed variants.

Indices in A.T.		Indices in Matrix		S.C.F.
Habit Plane	Shear Direction	Habit Plane	Shear Direction	
($\bar{15}$ 3 10)	[.60 .24 .77]	(1 $\bar{53}$ 14)	[.55 .19 .81]	0.65
($\bar{15}$ 10 3)	[.60 .77 .24]	(29 $\bar{46}$ 7)	[.15 .02 .99]	0.59
($\bar{15}$ $\bar{3}$ 10)	[.60 .24 .77]	($\bar{11}$ $\bar{47}$ 26)	[.87 .03 .50]	0.48 ○
($\bar{3}$ 15 10)	[.24 .60 .77]	(13 $\bar{41}$ $\bar{34}$)	[.83 .47 .30]	0.37
($\bar{15}$ $\bar{10}$ 3)	[.60 .77 .24]	($\bar{11}$ $\bar{26}$ 47)	[.87 .50 .03]	0.21 ○
(15 10 3)	[.60 .77 .24]	($\bar{1}$ 14 $\bar{53}$)	[.55 .81 .19]	0.18 ○
($\bar{3}$ $\bar{15}$ 10)	[.24 .60 .77]	($\bar{47}$ $\bar{11}$ 26)	[.03 .87 .50]	0.17
($\bar{10}$ 15 3)	[.77 .60 .24]	(34 $\bar{41}$ $\bar{13}$)	[.30 .47 .83]	0.10
(15 3 10)	[.60 .24 .77]	($\bar{29}$ 7 $\bar{46}$)	[.15 .99 .02]	0.09
($\bar{10}$ 3 15)	[.77 .24 .60]	($\bar{14}$ $\bar{53}$ $\bar{1}$)	[.81 .19 .55]	0.05
($\bar{10}$ $\bar{15}$ 3)	[.77 .60 .24]	($\bar{26}$ $\bar{11}$ 47)	[.50 .87 .03]	-0.03
(15 $\bar{3}$ 10)	[.60 .24 .77]	($\bar{41}$ 13 $\bar{34}$)	[.47 .83 .30]	-0.04
($\bar{10}$ $\bar{3}$ 15)	[.77 .24 .60]	($\bar{26}$ $\bar{47}$ 11)	[.50 .03 .87]	-0.07
(3 10 15)	[.24 .77 .60]	($\bar{13}$ $\bar{34}$ $\bar{41}$)	[.83 .30 .47]	-0.11
(3 15 10)	[.24 .60 .77]	(7 $\bar{29}$ $\bar{46}$)	[.99 .15 .02]	-0.11
($\bar{3}$ $\bar{10}$ 15)	[.24 .77 .60]	($\bar{47}$ $\bar{26}$ 11)	[.03 .50 .87]	-0.13
(3 $\bar{10}$ 15)	[.24 .77 .60]	($\bar{53}$ $\bar{14}$ $\bar{1}$)	[.19 .81 .55]	-0.24
(3 $\bar{15}$ 10)	[.24 .60 .77]	($\bar{53}$ 1 14)	[.19 .55 .81]	-0.26
(15 $\bar{10}$ 3)	[.60 .77 .24]	($\bar{41}$ 34 $\bar{13}$)	[.47 .30 .83]	-0.33
(10 $\bar{15}$ 3)	[.77 .60 .24]	($\bar{46}$ 29 7)	[.02 .15 .99]	-0.42
(10 15 3)	[.77 .60 .24]	(14 $\bar{1}$ $\bar{53}$)	[.81 .55 .19]	-0.43
($\bar{3}$ 10 15)	[.24 .77 .60]	($\bar{7}$ $\bar{46}$ $\bar{29}$)	[.99 .02 .15]	-0.51
(10 $\bar{3}$ 15)	[.77 .24 .60]	($\bar{46}$ $\bar{7}$ $\bar{29}$)	[.02 .99 .15]	-0.54
(10 3 15)	[.77 .24 .60]	($\bar{34}$ $\bar{13}$ $\bar{41}$)	[.30 .83 .47]	-0.55

Table 5.4 N_{ij} value for 24 martensite variants.

-o: observed variants.

Indices in A.T.		Indices in Matrix		S.C.F.
Habit Plane	Shear Direction	Habit Plane	Shear Direction	
(15 10 3)	[.60 .77 .24]	($\bar{1}$ 14 $\bar{53}$)	[.55 .81 .19]	0.65
($\bar{15}$ $\bar{10}$ 3)	[.60 .77 .24]	($\bar{11}$ $\bar{26}$ 47)	[.87 .50 .03]	0.48
(15 $\bar{10}$ 3)	[.60 .77 .24]	(41 34 $\bar{13}$)	[.47 .30 .83]	0.46
(15 3 10)	[.60 .24 .77]	(29 7 46)	[.15 .99 .02]	0.44
(3 10 15)	[.24 .77 .60]	($\bar{13}$ $\bar{34}$ 41)	[.83 .30 .41]	0.32
($\bar{15}$ 10 3)	[.60 .77 .24]	(29 46 7)	[.15 .02 .99]	0.29
($\bar{15}$ $\bar{3}$ 10)	[.60 .24 .77]	($\bar{11}$ 47 26)	[.87 .03 .50]	0.21
($\bar{15}$ 3 10)	[.60 .24 .77]	(1 $\bar{53}$ 14)	[.55 .19 .81]	0.18
($\bar{3}$ $\bar{10}$ 15)	[.24 .77 .60]	(47 26 11)	[.03 .50 .87]	0.17
(15 $\bar{3}$ 10)	[.60 .24 .77]	(41 13 34)	[.47 .83 .30]	0.12
(3 15 10)	[.24 .60 .77]	(7 29 46)	[.99 .15 .02]	0.11
(10 3 15)	[.77 .24 .60]	(34 $\bar{13}$ 41)	[.30 .83 .47]	0.10
(10 15 3)	[.77 .60 .24]	(14 $\bar{1}$ $\bar{53}$)	[.81 .55 .19]	0.05
($\bar{10}$ $\bar{3}$ 15)	[.77 .24 .60]	(26 47 11)	[.50 .03 .87]	-0.03
($\bar{10}$ $\bar{15}$ 3)	[.77 .60 .24]	(26 $\bar{11}$ 47)	[.50 .87 .03]	-0.07
($\bar{3}$ 10 15)	[.24 .77 .60]	(7 46 29)	[.99 .02 .15]	-0.12
($\bar{3}$ 15 10)	[.24 .60 .77]	(13 41 34)	[.83 .47 .30]	-0.13
($\bar{3}$ $\bar{15}$ 10)	[.24 .60 .77]	(47 $\bar{11}$ 26)	[.03 .87 .50]	-0.14
(3 $\bar{15}$ 10)	[.24 .60 .77]	(53 1 14)	[.19 .55 .81]	-0.24
(3 $\bar{10}$ 15)	[.24 .77 .60]	(53 $\bar{14}$ $\bar{1}$)	[.19 .81 .55]	-0.26
(10 $\bar{3}$ 15)	[.77 .24 .60]	(46 7 29)	[.02 .99 .15]	-0.42
($\bar{10}$ 3 15)	[.77 .24 .60]	($\bar{14}$ $\bar{53}$ $\bar{1}$)	[.81 .19 .55]	-0.43
(10 $\bar{15}$ 3)	[.77 .60 .24]	(46 29 7)	[.02 .15 .99]	-0.54
($\bar{10}$ 15 3)	[.77 .60 .24]	(34 41 $\bar{13}$)	[.30 .47 .83]	-0.55

was studied using the Fe-25%Ni-0.7%C alloy and the Fe-32%Ni alloy. The results obtained are as follows.

(1) The M_s temperature of both the Fe-25%Ni-0.7%C alloy and the Fe-32%Ni alloy was increased by prior deformation at 298K (above M_d temperature for both alloys).

(2) In the Fe-25%Ni-0.7%C specimen cooled just below the M_s , martensites were frequently observed at the grain boundaries or the annealing twin boundaries on which deformation twins (that were introduced by prior deformation) impinged.

(3) For such martensites, the N_{ij} value between the twinning system which impinged on the boundaries and the shear of martensitic transformation on the habit plane is relatively large. From the analysis, it was concluded that such martensite were formed by the aid of residual stress at the grain boundaries or the annealing twin boundaries on which deformation twins impinged.

From the results (2) and (3), the increase of the M_s temperature by prior deformation was considered to be due to the local residual stress introduced by impingement of deformation twins on the grain boundaries or annealing twin boundaries or piled-up dislocations during deformation.

The results obtained in this Chapter support the conclusion in Chapter IV that the effect of plastic strain for deformation-induced martensitic transformation is to introduce the localized regions where stress is concentrated.

References for Chapter V

- 1) T. G. Gooch and D. R. F. West: J. Iron Steel Inst., 205(1967), 555.
- 2) I. Tamura, T. Maki, M. Nakanishi and H. Oka: Suiyokaishi, 17(1970).75.
- 3) I. Ya. Georgiyeva, G. V. Kurdjumov, O. P. Maksimova and V. V. Nemirovskiy: Fiz. Met. Metalloved, 23(1967), 23.
- 4) J. R. C. Guimarães and J. C. Shyne: Scripta Met., 4(1970), 1019.
- 5) J. R. C. Guimarães and J. C. Shyne: Met. Trans., 2(1971), 2063.
- 6) D. R. Wallman and J. R. C. Guimarães: Scripta Met., 7(1973), 355.
- 7) M. Tokizane: Scripta Met., 10(1976), 459.
- 8) A. B. Greninger and A. R. Troiano: Trans. AIME, 145(1941), 289.
- 9) T. Maki and C. M. Wayman: Proc. of the First JIM Int. Symposium (1976), 69.
- 10) T. Maki, S. Shimooka, M. Umemoto and I. Tamura: Trans. JIM, 13(1972), 400.
- 11) R. H. Richman and G. F. Bolling: Met. Trans., 2(1971), 2451.
- 12) M. S. Wechsler, D. S. Lieberman and T. A. Read: Trans. AIME, 197(1953), 1503.

CHAPTER VI SUMMARY

TRIP(Transformation-induced plasticity) is considered to be very effective to make steels which achieve both high toughness and high strength, and to improve formability of austenitic stainless steels. However, deformation-induced martensitic transformation, which is the fundamental point of TRIP phenomenon, has not been made clear. This study was performed to obtain a clear understanding of the role of applied stress and plastic strain during deformation for the deformation-induced martensitic transformation. The results obtained in each Chapter are summarized as follows.

In Chapter I, the subjects on the deformation-induced martensitic transformation were reviewed and summarized, and the purpose of this study was described.

In Chapter II, the effect of applied stress on the deformation-induced martensitic transformation was studied. A quantitative model to relate the mechanical driving force and the amount of martensite which was induced by the applied stress was proposed. The model was verified by comparing the amount of martensite which was calculated from the model with that measured experimentally. It was concluded that the effect of applied stress on the martensitic transformation was expressed by the mechanical driving force, and the stress aided the total shape strain accompanying the transformation.

In Chapter III, the banded structure, which Maxwell et al called strain-induced martensite, was examined by optical and electron microscope. It was shown that the banded structure was not martensite but deformation twins of austenite.

In Chapter IV, the effect of deformation twinning of austenite on the deformation-induced martensitic transformation was examined to reveal the effect of plastic strain on the transformation using Fe-Ni-C alloys. Martensites were frequently observed at the grain boundaries or the annealing twin boundaries on which deformation twins impinged. Such martensites had large N_{ij} (stress coupling factor) value for the impinging twinning system and the shear of martensitic transformation, and at the same time they had large Schmid factor for the direction of applied stress. It was concluded that such martensites were formed by the aid of the concentrated stress due to the impingement of deformation twins of austenite plus by the aid of applied stress.

Martensite which observed at the junction of two deformation twins in tension was not formed from the embryo produced by the intersection of two shear bands which suggested by Olson and Cohen, since from the crystallographic analysis, the combination of these twins did not correspond to Bogers and Burgers double shears. In compression, martensite which could be identified clearly was not tend to form preferentially at the intersection of two deformation twins that were considered to correspond to Bogers and Burgers double shears. It was concluded that the strain-induced embryo formation mechanism, which was proposed by Olson and Cohen, was not prevailing mechanism for deformation-induced martensitic transformation.

In Chapter V, the effect of prior deformation on the martensitic transformation was studied to separate the effect of local concentrated stress from the effect of total applied stress

on the martensitic transformation during deformation. Martensites were frequently observed at the grain boundaries or the annealing twin boundaries on which deformation twins (that were introduced by prior deformation) impinged. Such martensites had relatively large N_{ij} values for the twinning system which impinged on the boundaries and the shear of martensitic transformation. It was concluded that such martensites were formed by the aid of residual local stress at the grain boundaries or the annealing twin boundaries on which deformation twins (that were introduced by prior deformation) impinged. The increase of the M_s temperature by prior deformation was considered to be due to the local residual stress introduced by prior deformation.

The results obtained in this Chapter support the conclusion in Chapter IV that the effect of plastic strain in deformation-induced martensitic transformation is to introduce the localized region where stress is concentrated.

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