Effect of Annealing on the Structure and Hardness of Electrodeposited Ni–W Alloys

Shinichiro HAYATA,1) Satoshi OUE,2) Hiroaki NAKANO2)* and Takehiro TAKAHASHI1,3)

1) Graduate School of Engineering, Kyushu University, 744, Motoooka, Nishi-ku, Fukuoka-shi, 819-0395 Japan.
2) Faculty of Engineering, Kyushu University, 744, Motoooka, Nishi-ku, Fukuoka-shi, 819-0395 Japan.
3) Hirohata R & D Lab., Nippon Steel & Sumitomo Metal, 1, Fuji-machi, Hirohata-ku, Himeji-shi, 671-1188 Japan.

(Received on November 4, 2014; accepted on January 19, 2015; originally published in Tetsu-to-Hagané, Vol. 100, 2014, No. 11, pp. 1391–1397)

The electrodeposition of Ni–W alloys was conducted from an unagitated sulfate solution containing citric acid at pH 5 and 60°C under coulostatic (3.44–6.22 × 10^5 C/m^2) and galvanostatic (30–5 000 A/m^2) conditions. Before annealing, the lattice constant of Ni increased linearly with an increase in W content up to 40.7 mass% in accordance with Vegard’s law. This shows that the W formed a supersaturated solid solution in Ni. At W contents of < 37.1 mass%, the deposit morphologies showed a field-oriented texture, with a preferred orientation of the specific plane towards the electric field during deposition, and platelet crystal edges exposed at the surface. At W contents of > 40.7 mass% of the solid solubility limit, the cross section of the deposits showed a layered morphology, with a smooth surface and small granular crystals. After annealing, Ni4W precipitated with deposits of W contents of 32.6 and 37.1 mass%. Fine precipitates of NiW and NiW formed over the entire surface with W contents of 40.7–45.3 mass%. Before annealing, the hardness of the deposits increased with W content, and the increase was particularly large at a W content of 40.7 mass%. The hardness was almost constant regardless of the current density for W contents of > 40.7 mass%. The alloy composition required to change the hardness of the deposits corresponded with that required to change the deposit structures. The hardness of the deposits increased for all W contents by annealing, with the increase being particularly large for W contents of > 40.7 mass%.

KEY WORDS: Zn; electrodeposition; Ni–W alloy; deposits; structure; hardness; annealing; morphology; orientation.

1. Introduction

Nickel-plated steel sheets, which are used in numerous applications, such as in batteries, are likely to cause the seizure of plated layers and of dies during press working. Because this prevents stable quality production, hard plated films are required that do not adhere easily to dies. When deposited hard alloy films with excellent abrasion resistance are coated onto Ni-plated steel sheets, steel sheet seizure is unlikely, and excellent consecutive press formability can be achieved. In this study, Ni–W alloys are examined as deposited hard alloy films. Many studies have been conducted on electrodeposited Ni–W alloys because of their excellent resistance to corrosion,1,2) and abrasion.2,3) The number of studies conducted has increased since the recent discovery that electrodeposited Ni–W alloys are likely to be amorphous,4,5) and that their corrosion resistance is improved significantly by the formation of homogeneous and stable passive films.6–9) The codeposition of W occurs in the presence of iron-group metals although it cannot be deposited by itself from aqueous solutions (similar to Mo).10–17) This is termed induced-type codeposition.

The hardness of deposited Ni–W alloys increases with annealing, but many ambiguities exist regarding the hardening mechanism.18–23) Therefore, in this study, the relationship between the hardness and the crystal texture of deposited Ni–W alloys before and after annealing is investigated to clarify the effect of annealing on the hardness of the deposits. As the first step to simplify the system, Ni–W alloys were deposited directly on steel sheets without initial Ni plating, and the hardness and texture were investigated.

2. Experimental

Table 1 shows the solution composition and electrolysis conditions. An electrolytic solution was prepared by dissolving fixed amounts of high-grade Na2WO4·2H2O (0.2 mol/L), NiSO4·6H2O (0.21 mol/L), Cu2H2O4·2NH3 (0.45 mol/L), and HCOONa (0.2 mol/L) in distilled deionized water. The pH was adjusted to 5 using sulfuric acid. Electrodeposition was performed in unagitated solutions under galvanostatic (30–5 000 A/m^2) conditions at 60°C. The charge was set at 3.44 × 10^4 and 6.22 × 10^4 C/m^2 to investigate the deposition behavior and to analyze the deposit structures, respectively. Steel sheets measuring 3 × 3 cm^2 and stainless steel sheets...
(SUS304) measuring 15×3 cm² were used as cathode and anode substrates, respectively. The deposits were dissolved from the cathode using a mixture of nitric acid and hydrogen fluoride (HNO₃/HF = 1/1). Both Ni and W were analyzed quantitatively by inductively coupled plasma spectrometry. The W content in the deposits and the current efficiency for Ni and W deposition were calculated. The partial current densities for Ni and W deposition and H₂ evolution were calculated by multiplying the total current density by the current efficiency. The current efficiency for H₂ evolution was calculated by subtracting the current efficiency for Ni and W deposition from 100%. The cathode potentials during deposition were measured against a saturated Ag/AgCl reference electrode (0.199 V vs. the normal hydrogen electrode, 25°C). The potentials of the polarization curves were plotted with reference to the NHE.

For annealing, the deposited Ni–W alloys were maintained at 700±5°C for 15 min in N₂ atmosphere, cooled to 200°C and then removed to the atmosphere. The deposited Ni–W alloy structure was examined by X-ray diffraction. The Ni crystallite size in the deposits was calculated using Scherrer’s equation and the half-width of the X-ray diffraction peak of the 111 reflection. The cross-sectional structure in the micro-areas of the deposited alloys was observed using a transmission electron microscope (TECNAI-20, FEI Co., Ltd). The specimens were cut with a diamond knife on an ultramicrotome. The surfaces and cross-sectional morphologies of the deposits were observed by scanning electron microscopy (SEM). The specimen was analyzed from an electron backscatter diffraction (EBSD) pattern after mirror polishing and removing the damaged layer by Ar sputtering. In EBSD, the crystallographic orientation of the tangential direction with respect to the cross section of the deposited Ni–W alloys (i.e., normal to the surface of the deposited Ni–W alloy) was examined as the reference direction. To investigate the diffusion of Ni, W, and Fe by annealing, the cross-sectional distribution of each element in the deposited Ni–W alloys was examined by energy dispersive X-ray analysis (EDX). The hardness of the deposited Ni–W alloys was measured on the surface by nanoindentation for a load of 10 mN, and with 500 partitions and a step interval of 20 ms.

### Table 1. Electrolysis conditions.

| Bath composition     | Concentration (mol/L) |
|----------------------|------------------------|
| Na₂WO₄·2H₂O          | 0.2                    |
| NiSO₄·6H₂O           | 0.2                    |
| C₆H₈O₇·2NH₃         | 0.45                   |
| HCOONa               | 0.2                    |
| pH                   | 5                      |

| Operating conditions | Current density (A/m²) | Amount of charge (C/m²) |
|----------------------|------------------------|-------------------------|
| Cathode              | Fe (3×3 cm²)           | 3.44×10⁴, 6.22×10⁴     |
| Anode                | SUS304 (15×3 cm²)      |                         |
| Quiescent bath       |                        |                         |

### Fig. 1. Polarization curves for Ni–W deposition.

### Fig. 2. Effect of current density on W content in deposits and current efficiency for Ni–W alloy deposition.

#### 3. Results and Discussion

### 3.1. Deposition Behavior of Ni–W Alloy

Figure 1 shows the total and partial polarization curves for the Ni–W alloy and Ni and W deposition, respectively. In the total polarization curve, the current density rose at approximately −0.5 V, and the cathode potential shifted to a less noble direction above 600 A/m². At approximately −0.85 V, the current density began to increase again. The partial polarization curves for Ni and W deposition showed the same tendency as the total polarization curve.

Figure 2 shows the effect of current density on the W content in the deposits and the current efficiency for Ni–W alloy deposition. The W content in the deposits increased with current density, and reached a maximum of approximately 42 mass% at 600 A/m². It then decreased gradually with increasing current density. The current efficiency for Ni–W alloy deposition increased initially with current density to a maximum of approximately 49% at 600 A/m², and then decreased gradually with increasing current density. The current density dependence of W content in the deposits showed the same tendency as that of the current efficiency for alloy deposition. From Figs. 1 and 2, the current density in the total and partial polarization curves for Ni–W alloy and Ni and W deposition, respectively, can be seen to start increasing again beyond the current density, at which point the W content in the deposits and the current efficiency for alloy deposition reach a maximum.
3.2. Effect of Annealing on Deposited Ni–W Alloy Structure

Figure 3 shows the X-ray diffraction patterns before annealing of the Ni–W alloys with various compositions deposited at each current density. The alloys of low W content obtained at low current densities showed diffraction patterns of a single Ni phase, which means that a solution of W formed in Ni. As the W content in the deposits increased with current density, the diffraction patterns broadened, which shows a significant decrease in crystal grain size. Alloys with W contents above 45.2 mass% deposited at current densities above 1 000 A/m² showed halo patterns in their diffraction spectra, which is a characteristic of an amorphous structure.

Figure 4 shows the X-ray diffraction patterns after annealing of the Ni–W alloys with various compositions deposited at each current density. After annealing, all the deposited Ni–W alloys, including those with broad diffraction patterns before annealing, showed the sharp peaks of the Ni phase. X-ray diffraction patterns for alloys with W contents of 32.6 and 37.1 mass% obtained at 100 and 200 A/m², respectively, showed peaks resulting from Ni₄W. Alloys with W contents from 40.7 to 45.3 mass% obtained at 500 to 3 000 A/m², respectively, showed peaks of NiW in addition to Ni₄W. According to the equilibrium phase diagram of the Ni–W binary system, Ni₄W and NiW are the stable phases for W contents of 40.7–42.7 and 75.3–75.8 mass%, respectively. Therefore, in this study, the metallic compounds precipitated by annealing correspond to the stable phase expected from the equilibrium phase diagram of the Ni–W binary system.

Figure 5 shows the electron diffraction patterns before and after annealing of the Ni–W alloy with W content of 32.6 mass% deposited at 100 A/m². After annealing (b), spots corresponding to Ni₄W were observed inside the 111 reflection of Ni, which confirms the X-ray diffraction result (Fig. 4).

Figure 6 shows the electron diffraction patterns before and after annealing of the Ni–W alloy with W content of 44.9 mass% deposited at 3 000 A/m². After annealing (b), spots corresponding to Ni₄W and NiW were observed inside the 111 reflection of Ni, which confirms the X-ray diffraction result.
tion result (Fig. 4). Before annealing (a), the deposited Ni–W alloy is composed of obscure wide rings. The presence of W was not observed from the radius of the rings, showing that the deposited Ni–W alloy is the solid solution of Ni containing W. Although the deposited Ni–W alloys showed halo patterns in the diffraction spectra (Fig. 3), which are characteristic of an amorphous structure, assuming that these are amorphous, a single weak ring should appear in the electron diffraction patterns. Because multiple obscure wide rings appeared, the deposited Ni–W alloys were concluded to be a crystallite of Ni solid solution.

As shown in Fig. 3, the diffraction peaks shifted to the low diffraction angle region with an increase in W content in the deposits. The lattice constants of Ni were calculated from the shift of the main diffraction peak corresponding to the 111 reflection of Ni to a lower angle. Figure 7 shows the dependence of the lattice constant of Ni on the W content in the deposits. In accordance with Vegard’s law, the lattice constant of Ni increased linearly with an increase in W content up to 40.7 mass%. From the equilibrium phase diagram of the Ni–W alloy, the maximum amount of W in the Ni solid solution is seen to be approximately 29% at 50°C, indicating that electrodeposition increases the solubility of W in a Ni solid solution. After annealing, the solubility of W in a Ni solid solution decreases irregularly because of the precipitation of Ni₄W and NiW. As a result, the linear relationship between the lattice constant of Ni and W contents in the deposits is incomplete.

Figure 8 shows the relationship between the current density and the Ni crystallite size before and after annealing of the Ni–W alloys deposited at various current densities. Before annealing, the crystallite size of Ni in the deposited Ni–W alloys decreased with increasing current density. Because the W content in the deposits increased with current density as shown in Fig. 2, the decrease in Ni crystallite size is attributed to an increase in W content in the deposits and the current density. In contrast, after annealing, the Ni crystallite size was larger than that before annealing at all current densities. The effect of current density on Ni crystallite size has rarely been observed after annealing.
Figure 9 shows the secondary and backscattered electron images of the Ni–W alloy surface before and after annealing and deposited at 100, 200, and 500 A/m². Before annealing, the deposit (a) obtained at 100 A/m² showed an exposure of platelet crystals edges, and small granular crystals were present at the boundaries between the platelet crystals. At 200 A/m² (c), the number of platelet crystals decreased and that of the granular crystals increased. When the current density was increased to 500 A/m² (e), the platelet crystals disappeared, and the deposits were completely composed of small granular crystals, which resulted in a smooth surface. The backscattered electron images from energy selective backscatter electron detector (EsB) before annealing showed that a slight contrast resulting from the surface roughness was observed in the deposits (b, d) obtained at 100 and 200 A/m², but all the deposits, including the one (f) obtained at 500 A/m², exhibited an almost uniform surface. This indicates that the distribution of W in the deposits is uniform. After annealing, the platelet crystals in the deposits (g, i) obtained at 100 and 200 A/m², respectively, decomposed and changed to granular crystals. In the backscattered electron images (h, j), the white object resulting from the W compounds was present in scattered form in the deposits. Increasing the current density to 500 A/m² (l) led to the presence of a linear white object resulting from the W compounds was observed over the entire surface of the deposits. With an increase in current density to 500 A/m² (l), the white object from the W compounds was linear. The W compounds were distributed unevenly with an aggregated state in the deposits (h, j) obtained at 1 000 and 2 000 A/m², while they were small and dispersed in deposit (l) obtained at 3 000 A/m².

Figure 10 shows enlargements of the SEM images in Fig. 10 for the surface before annealing the Ni–W alloys deposited at 2 000 and 3 000 A/m². The deposits obtained at high current densities were composed of granular crystals, and small pinholes and cracks were observed at the boundary between the clusters formed by the small granular crystals.

Figure 11 shows the secondary and backscattered electron images before and after annealing of Ni–W alloys deposited at 1 000, 2 000, and 3 000 A/m². Before annealing, the deposit (a) obtained at 100 A/m² showed an exposure of platelet crystals edges, and small granular crystals were present at the boundaries between the platelet crystals. At 200 A/m² (c), the number of platelet crystals decreased and that of the granular crystals increased. When the current density was increased to 500 A/m² (e), the platelet crystals disappeared, and the deposits were completely composed of small granular crystals, which resulted in a smooth surface. The backscattered electron images from energy selective backscatter electron detector (EsB) before annealing showed that a slight contrast resulting from the surface roughness was observed in the deposits (b, d) obtained at 100 and 200 A/m², but all the deposits, including the one (f) obtained at 500 A/m², exhibited an almost uniform surface. This indicates that the distribution of W in the deposits is uniform. After annealing, the platelet crystals in the deposits (g, i) obtained at 100 and 200 A/m², respectively, decomposed and changed to granular crystals. In the backscattered electron images (h, j), the white object resulting from the W compounds was present in scattered form in the deposits. Increasing the current density to 500 A/m² (l) led to the presence of a linear white object resulting from the W compounds was observed over the entire surface of the deposits. With an increase in current density to 500 A/m² (l), the white object from the W compounds was linear. The W compounds were distributed unevenly with an aggregated state in the deposits (h, j) obtained at 1 000 and 2 000 A/m², while they were small and dispersed in deposit (l) obtained at 3 000 A/m².

Figure 11 shows enlargements of the SEM images in Fig. 10 for the surface before annealing the Ni–W alloys deposited at 2 000 and 3 000 A/m². The deposits obtained at high current densities were composed of granular crystals, and small pinholes and cracks were observed at the boundary between the clusters formed by the small granular crystals.

Figure 12 shows the SEM results with angle selective backscattered electron detector (AsB) and EBSD orientation mapping images before and after annealing of the Ni–W alloys deposited at 100 and 200 A/m². Before annealing, the EBSD orientation mapping images revealed that the Ni–W alloys deposited at various current densities obtained a field-oriented texture, in which the preferred orientation of the specific planes is towards the electric field during deposition. After annealing, granular crystals were deposited over the entire surface, regardless of the current density as can be seen from the SEM–AsB images. These precipitates are expected to be Ni₄W from the X-ray diffraction patterns in Fig. 4. Many areas existed where the orientation cannot be analyzed by EBSD because of the presence of precipitates.
after annealing.

Figure 13 shows the SEM–AsB and EBSD orientation mapping images of the cross section before and after annealing of the Ni–W alloys deposited at 500 and 3 000 A/m². Before annealing, both the deposits obtained at 500 and 3 000 A/m² had a layered texture as seen from the SEM–AsB images. This layered texture may be attributed to a variation in solution constituent concentration close to the cathode during deposition. Because the crystals of the deposits with layered texture are small, the orientation of Ni in the layered texture cannot be analyzed by EBSD. At 500 A/m², the deposit was columnar in the initial stage but changed to a layered texture halfway through the deposition as can be seen from the SEM–AsB and EBSD images. Figures 9, 10, and 13 show that the surface of the deposits were completely composed of small granular crystals and became smooth at current densities above 500 A/m² when they became layered in texture. Considering that the W content...
in the deposits obtained at 500 A/m² was 40.7 mass% of the solid solubility limit and the deposits changed to a layered texture halfway through the deposition, the texture of the deposits is expected to change significantly around this W content in the deposits. After annealing, granular precipitates occurred in the deposits obtained at all current densities as shown in the SEM–AsB images. A large number of small precipitates occurred particularly in the vicinity of the iron substrate. The Ni orientation cannot be analyzed by EBSD in the layered-texture area before annealing, whereas it can be analyzed after annealing because of the increase in Ni crystallite size. A granular or cluster-like orientation was observed, which is different from the field-oriented texture of the preferred orientation of specific planes.

Figure 14 shows the cross-sectional SEM–AsB images and EDX spectra after annealing of the Ni–W alloy deposited at 200 A/m². The contrast in the image occurred from the deposit (a) and steel substrate (b). Iron was detected in deposit [(1), (2), (3)], which shows that diffusion of iron occurred from the substrate to the deposit. Nickel was detected in the steel substrate [(4), (5)], but W was seldom observed. This shows the preferential diffusion of Ni in the deposits to the steel substrate. The precipitation of Ni₄W and NiW in the vicinity of the steel substrate after annealing, shown in Fig. 13, was possibly caused by the increase in W content in the deposits because of the diffusion of Ni in the deposits toward the steel substrate.

3.3. Effect of Annealing on the Hardness of Deposited Ni–W Alloys

Figure 15 shows the hardness before and after annealing of the Ni–W alloys with various compositions deposited at each current density. Before annealing, the hardness of the deposits increased with W content in the deposits, and increased particularly at a W content of 40.7 mass%. However, the hardness of the deposits was almost constant regardless of current density at W contents above 40.7 mass%. At a W content of 40.7 mass% of the solid solubility limit, the deposits changed to a layered texture halfway through the deposition, which resulted in a smooth surface. The alloy composition at which the hardness of deposit changed significantly corresponded to that at which the texture of the deposits changed. The hardness of the deposits was increased by annealing at all alloy compositions. This increase in hardness is attributed to the precipitation of W compounds. The extent of increase in hardness became large at W contents above 40.7 mass%. Metallic compounds such as Ni₄W and NiW precipitated finely over the entire surface, and this contributed to an increase in the hardness of the deposits.

Pure W, Ni₄W, and an intermediate phase with unknown composition have been found to precipitate by annealing between 450 and 550°C of the deposited Ni–W alloys with W contents of 40 and 44 mass%. For the annealing conditions used in this study, metallic compounds, such as Ni₄W and NiW, were found to precipitate in the deposits.
Because the hardness increases by annealing, the Ni4W precipitates are considered to contribute significantly to the increase in hardness by annealing. In this study, iron diffusion with annealing because of the direct deposition of Ni–W alloys onto the steel substrate. As a result, the iron possibly affects the precipitation behavior of Ni4W and NiW. As mentioned in the introduction, for Ni–W alloy deposition onto the steel substrate with initial Ni plating, the deposited Ni–W alloys are seldom affected by iron diffusion with annealing. Under the conditions for which the deposited Ni–W alloys are seldom affected by iron diffusion with annealing, the precipitation behavior of Ni4W and NiW, and the effect of their precipitation on hardness are areas for future investigation.

4. Conclusions

The electrodeposition of Ni–W alloys was conducted from sulfate solution containing citric acid, and the structure and hardness of the deposits were investigated before and after annealing. Before annealing, the Ni lattice constant increased linearly with an increase in W content up to 40.7 mass% in accordance with Vegard’s law. This shows that the W formed a supersaturated solid solution in Ni. At W contents of <37.1 mass%, the deposits showed a morphology with field-oriented texture, with a preferred orientation of the specific plane towards the electric field in deposition, and platelet crystal edges exposed at the surface. At W contents of >40.7 mass% of the solid solubility limit, the cross section of the deposits showed a layered morphology, and the surface became smooth with small granular crystals. After annealing, Ni4W precipitated with W contents of 32.6 and 37.1 mass%, and Ni4W and NiW precipitated over the entire surface in the fine deposits with W contents of 40.7 to 45.3 mass%. Before annealing, the hardness of the deposits increased with W content, and this increase was particularly large at a W content of 40.7 mass%. The hardness was almost constant regardless of the current density at W contents of >40.7 mass%. The alloy composition required to change the hardness of the deposits agreed closely with that required to change the structure of the deposits. The hardness of the deposits increased for all W contents by annealing, and the extent of increase was particularly large at W contents of >40.7 mass%.

REFERENCES

1) T. Kudo and K. Mizuno: J. Met. Finish. Soc. Jpn., 13 (1962), 162.
2) T. Misakihige and R. Suzuki: J. Met. Finish. Soc. Jpn., 22 (1971), 216.
3) E. Eskin, O. Berklii, G. Rogalsky and A. Zahavi: Plat. Surf. Finish., 85 (1998), 79.
4) T. Watanabe: Tetsu-to-Hagané, 73 (1987), 2180.
5) K. Masumoto and T. Watanabe: Amorphous Plating Method and its Application, Nikkan Kogyo Shinbun Co., Ltd, Tokyo, (1990), 263.
6) H. Ōnomoto and T. Omii: Alloy Plating, Nikkan Kogyo Shinbun Co., Ltd, Tokyo, (1997), 130.
7) T. Omii and H. Takagi: Surf. Finish. Soc. Jpn., 40 (1989), 1432.
8) M. Li and M. Kowaka: J. Surf. Finish. Soc. Jpn., 44 (1993), 748.
9) H. Somekawa, T. G. Nieh and K. Higashi: Scr. Materialia, 50 (2004), 1361.
10) A. Brenner: Electrodeposition of Alloys, 2, Academic Press, New York and London, (1963), 399.
11) M. Sarojamma and T. L. Rama Char: J. Surf. Finish. Soc. Jpn., 29 (1978), 627.
12) H. Fukushima and K. Higashi: J. Surf. Finish. Soc. Jpn., 30 (1979), 538.
13) H. Fukushima, T. Akiyama, S. Akagi and K. Higashi: Trans. JIM, 20 (1997), 538.
14) T. Akiyama, H. Fukushima and K. Higashi: Tetsu-to-Hagané, 72 (1986), 918.
15) T. Akiyama and H. Fukushima: ISIJ Int., 32 (1992), 787.
16) D. W. Ernst, R. F. Amlie and M. L. Holt: J. Electrochem. Soc., 102 (1970), 461.
17) S. Oue, H. Nakano, S. Kobayashi and H. Fukushima: J. Electrochem. Soc., 156 (2009), D17.
18) K. H. Hou, Y. F. Chang, S. M. Chang and C. H. Chang: Thin Solid Films, 518 (2010), 7535.
19) A. J. Detor and C. A. Schuh: J. Mater. Res., 22 (2007), 3233.
20) T. Omii, K. Nishido, H. Matsumoto and H. Yamamoto: J. Met. Finish. Soc. Jpn., 40 (1989), 1287.
21) S. Morioka, Y. Sawada and R. Shimada: J. Met. Finish. Soc. Jpn., 16 (1965), 512.
22) T. Omii, M. Umatake and H. Yamamoto: J. Surf. Finish. Soc. Jpn., 40 (1989), 311.
23) A. Kubota, Y. Tsutoba, H. Nakano, S. Oue, S. Kobayashi, T. Akiyama and H. Fukushima: Tetsu-to-Hagané, 85 (1999), 728.
24) T. Watanabe: J. Surf. Finish. Soc. Jpn., 40 (1989), 280.
25) M. Hansen: Constitution of Binary Alloys, McGraw-Hill Book Co., New York, (1958), 1057.
26) W. B. Pearson: A Handbook of Lattice Spacings and Structures of Metals and Alloys, Pergamon Press, Oxford, (1958), 778.