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Effect of prior cold work on mechanical properties, electrical conductivity and microstructure of aged Cu-Ti alloys

S. NAGARJUNA*, K. BALASUBRAMANIAN‡
**Defence Metallurgical Research Laboratory and ‡Non-Ferrous Materials Technology Development Centre, Kanchanbagh-PO, Hyderabad 500 058, India*
D. S. SARMA
Department of Metallurgical Engineering, Banaras Hindu University, Varanasi 221 005, India

The mechanical properties, electrical conductivity and microstructure of Cu-2.7wt%Ti and Cu-5.4wt%Ti alloys have been studied in different conditions employing hardness and resistivity measurements, tensile tests and optical, scanning and transmission electron microscopy. Ageing of undeformed as well as cold worked alloys raises their hardness, strength and electrical conductivity. The hardness increased from 120 VHN for solution treated Cu-2.7Ti to 455 VHN for ST + cold worked + peak aged Cu-5.4Ti alloy. While tensile stength increased from 430 to 1450 MPa, the ductility (elongation) decreased from 36 to 1.5%. A maximum conductivity of 25% International Annealed Copper Standard (IACS) for Cu-2.7Ti and 14.5% IACS for Cu-5.4Ti is obtained with the present treatments. Peak strength was obtained when the solution treated alloys are aged at 450 °C for 16 hours due to precipitation of ordered, metastable and coherent β', Cu_4Ti phase having body centred tetragonal (bct) structure. While mechanical properties of Cu-Ti alloys are comparable, electrical conductivity is less than that of commercial Cu-Be-Co alloys. © *1999 Kluwer Academic Publishers*

1. Introduction

The fact that binary Cu-Ti alloys can serve as a substitute for expensive and toxic Cu-Be-Co alloys is acknowledged by the world-wide research carried out in the past and currently in progress. The mechanisms of spinodal decomposition and precipitation strengthening in Cu-Ti alloys have been studied extensively by many researchers [1–12]. Earlier studies on Cu-Ti alloys have been limited to investigations on the mechanisms of spinodal decomposition and age hardening with hardness/yield strength vs. ageing time plots and tensile properties in peak aged condition. Further, investigations on the effect of prior cold deformation on ageing was confined to hardness measurements. The influence of ageing time, temperature and prior cold work on tensile properties and electrical conductivity have not been studied in detail. Moreover, the alloy compositions studied were also selective and not over a wide range. Therefore, we have taken up a project to investigate various aspects of Cu-Ti alloys of four compositions viz. Cu-1.5wt%Ti, Cu-2.7wt%Ti, Cu-4.5wt%Ti and Cu-5.4wt%Ti. Firstly, grain size strengthening, Hall-Petch relationship and equations for flow stress of the solution treated Cu-Ti alloys have been reported [13–16]. Secondly, the effects of prior cold work on hardness, tensile properties and electrical conductivity of Cu-4.5wt%Ti [17] and Cu-1.5wt%Ti [18] alloys have been studied. Further, High Cycle Fatigue (HCF) properties [19] and electrical resistivity dependence on Ti content [20] in the four Cu-Ti alloys have been reported very recently. However, detailed information on the effect of prior cold work on hardness, tensile properties, electrical conductivity and microstructure of Cu-2.7Ti and Cu-5.4Ti alloys is lacking. The present paper therefore, reports the results obtained on the effect of prior cold deformation on hardness, tensile properties, electrical conductivity and microstructure of aged Cu-2.7Ti and Cu-5.4Ti alloys and a comparison of properties of the four Cu-Ti alloys in solution treated (ST), ST + aged and ST + cold worked + aged conditions.

2. Experimental

A 30 kg melt each of Cu-Ti alloys with aimed Ti content of 3.0 and 5.5 wt % was made in a Stokes Vacuum Induction Melting (VIM) Furnace with Oxygen Free Electronic (OFE) copper and Cu-26wt%Ti master alloy as raw materials. The Cu-Ti master alloy was also made in VIM furnace first and the analysis showed the composition of Ti in the master alloy as 26 wt %. The raw materials were melted in a graphite crucible at a vacuum of 10^{-3} torr, held at 1350 °C for 1/2 hour, cooled to 1230 °C and the liquid alloy was poured into a graphite mould which was cooled in vacuum in the mould chamber of the VIM furnace. The ingots were homogenised at 850 °C for 24 hours and analysed for Ti content. The analysed composition of the ingots is 2.7wt%Ti and 5.4wt%Ti. The oxygen content in these alloys was found to be 6.0 and 6.4 ppm respectively. The homogenised ingots were hot forged and rolled at 850 °C into 10 mm thick flats and 12 mm diameter rods.

Specimens from the hot rolled flats were solution treated at 900 °C for 2 hours and quenched rapidly into water. The solution treated specimens were aged at different temperatues (400, 450, 500 and 550 °C) for different times and their hardness (VHN) was measured. Round tensile specimens having 4.0 mm gauge diameter and 25 mm gauge length were made and tested for tensile properties in solution treated as well as peak aged (450 °C/16 h) conditions. All the tensile tests were carried out at ambient temperature and at a nominal strain rate of 10^{-3} s^{-1} using INSTRON 1185 ball-screw driven universal testing system.

The solution treated specimens were cold worked (CW) by rolling, giving different amounts of cold deformation. Hardness of the cold rolled specimens was measured after ageing at 400, 450 and 500 °C for different times. Flat tensile samples of 25 mm gauge length, 6 mm width and 1 mm thickness were made from the cold rolled strips according to the ASTM specification E 8M-89b (sub-size specimen) [21], peak aged at 400 °C and tested for tensile properties.

The solution treated rods were cold drawn to 2 mm diameter wires with intermittent solution treatments. Electrical resistance of the wires of 300 mm length were measured using Kelvin's Bridge apparatus at room temperature in ST, ST + aged and ST + CW (drawn) + aged conditions. Electrical conductivity (EC) in %

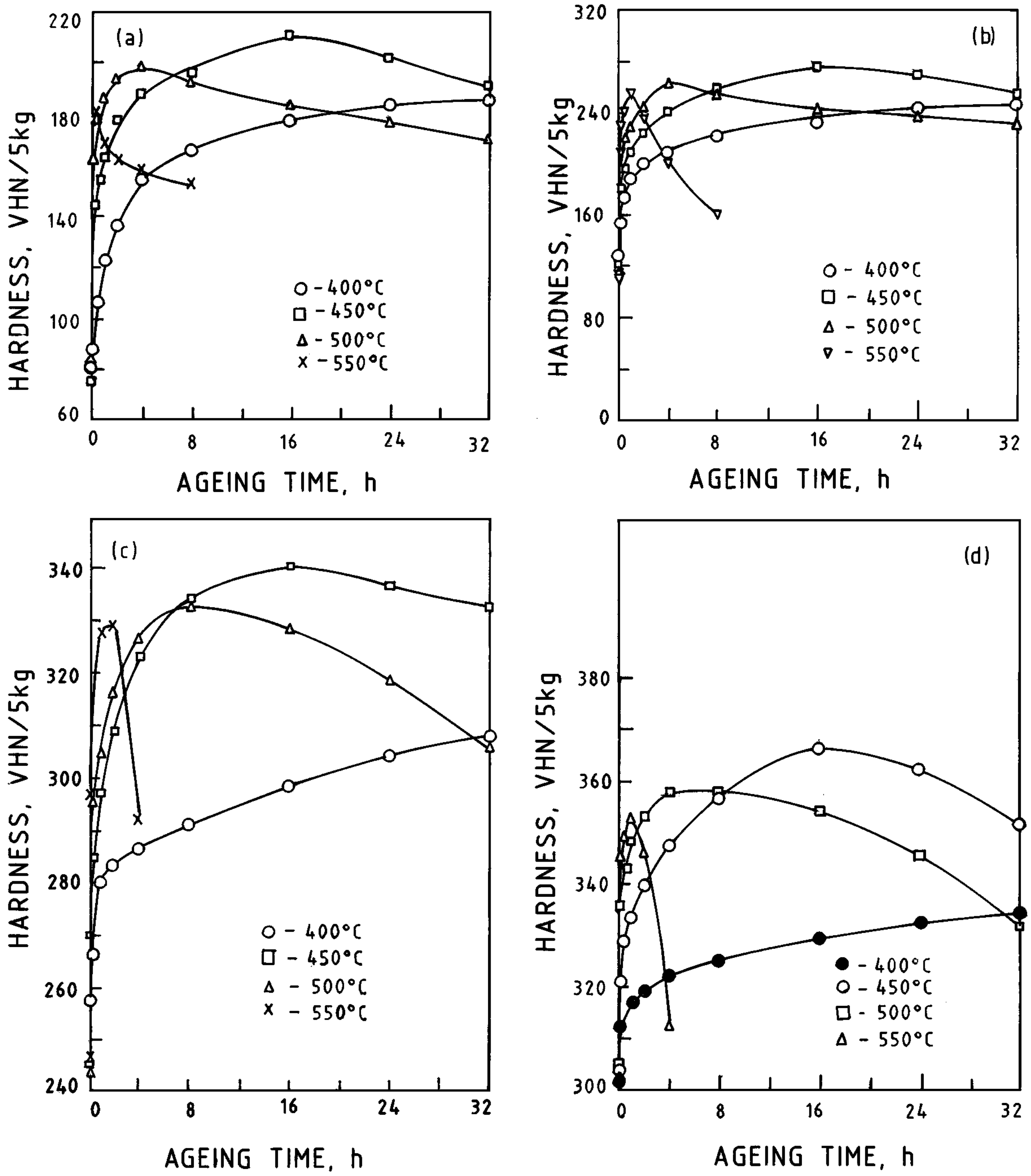

Figure 1 Age hardening in Cu-Ti alloys: (a) Cu-1.5Ti [18]; (b) Cu-2.7Ti; (c) Cu-4.5Ti [17]; and (d) Cu-5.4Ti.

IACS, was determined according to the ASTM specification: B193-89 [22], using the following expressions:

$$\rho = (R \cdot A)/L \tag{1}$$

$$EC = 172.41/\rho \tag{2}$$

where ρ is resistivity ($\mu\Omega \cdot \text{cm}$) and R is resistance ($\mu\Omega$) of the wire of length L (cm) and area of cross-section, A (cm^2).

Optical metallography was carried out in ST, peak aged (PA), overaged, ST + CW and ST + CW + PA conditions. Transmission electron microscopy (TEM) was also carried out in the above conditions. Thin slices were cut from the heat treated samples using ISOMET low speed cutting machine and polished mechanically to about 20 μm. 3 mm ϕ discs were punched out from these thinned sections and electropolished in a twin-jet electropolisher using a solution of 30 vol % nitric acid and 70% methanol at −45 °C and at a voltage of 10 V. The thin foils were examined at 160 kV using JEOL 200 CX transmission electron microscope. The fractured tensile test specimens were examined using ISI-100A scanning electron microscope.

3. Results

3.1. Hardness

The effect of ageing time on hardness of the solution treated Cu-2.7Ti and Cu-5.4Ti alloys is shown in Fig. 1. The age hardening curves for Cu-1.5Ti (Fig. 1a) [18] and Cu-4.5Ti (Fig. 1c) [17] are also included here for comparison. While the alloys have not attained peak hardness at 400 °C even after ageing for 32 hours, a peak hardness of 210 VHN for Cu-1.5Ti, 275 VHN for Cu-2.7Ti, 340 VHN for Cu-4.5Ti and 366 VHN for Cu-5.4Ti alloy, was reached after ageing at 450 °C

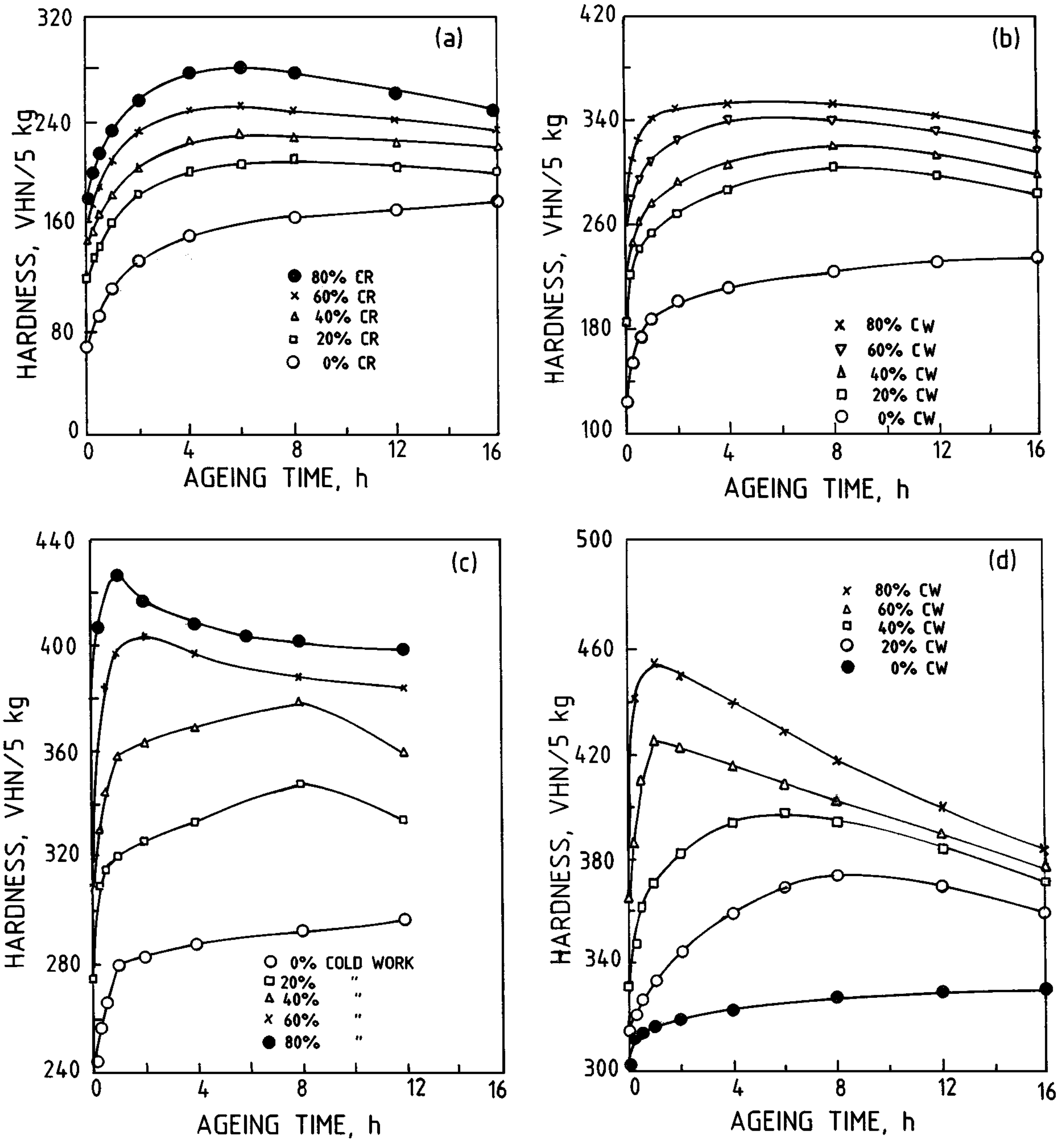

Figure 2 Influence of prior cold work on age hardening in Cu-Ti alloys aged at 400 °C: (a)Cu-1.5Ti [18]; (b) Cu-2.7Ti; (c) Cu-4.5Ti [17]; and (d) Cu-5.4Ti.

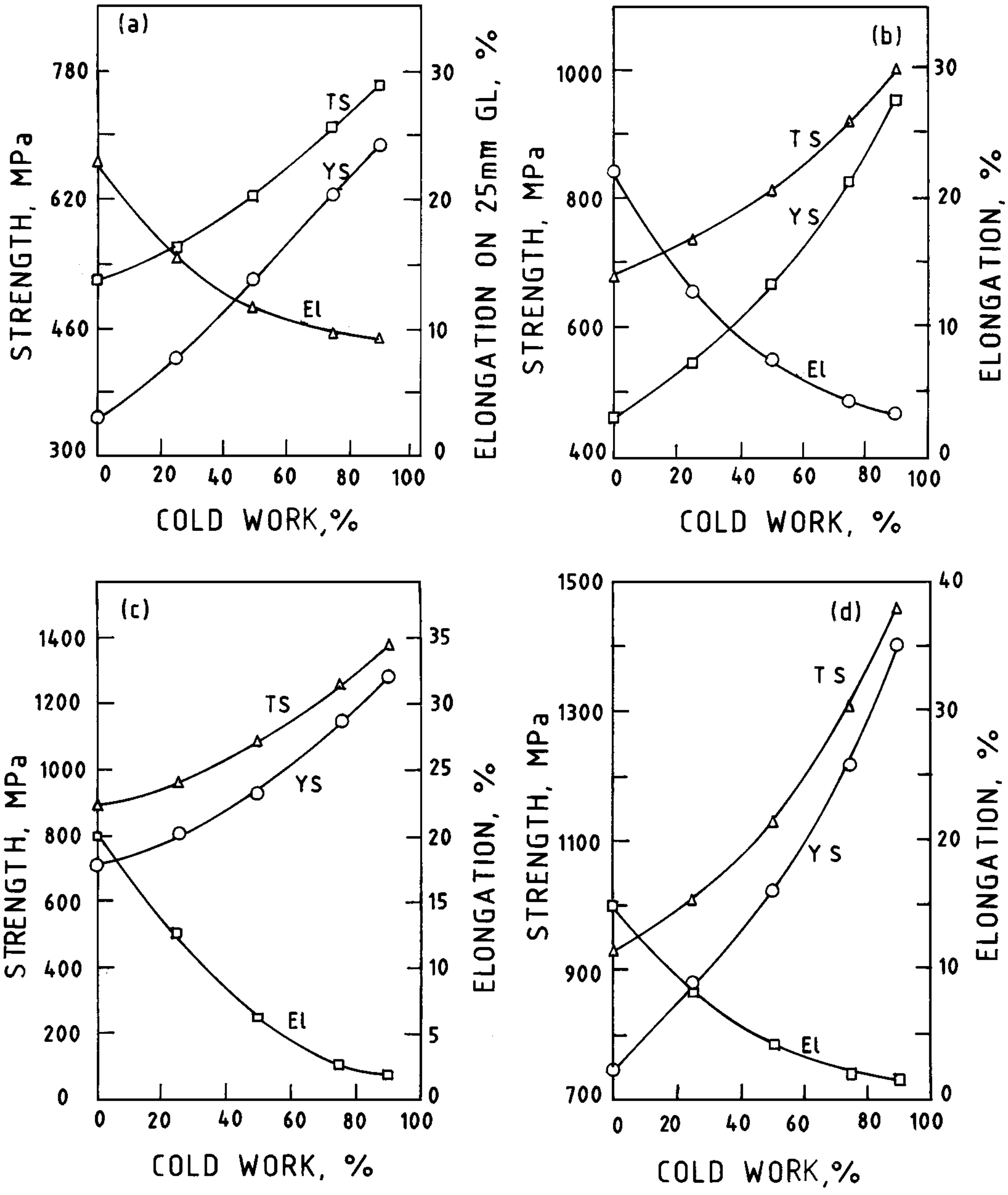

Figure 3 Variation of tensile properties with prior cold work in Cu-Ti alloys peak aged at 400 °C: (a) Cu-1.5Ti [18]; (b) Cu-2.7Ti; (c) Cu-4.5Ti [17]; and (d) Cu-5.4Ti.

for 16 hours, beyond which the hardness decreased with ageing time. Peak hardness was slightly lower and overaging was rapid on ageing at 500 °C. The overageing was drastic and peak hardness was further lowered when aged at 550 °C.

The influence of prior cold work on the hardness of Cu-2.7Ti and Cu-5.4Ti alloys aged at 400 °C is shown in Fig. 2. The age hardening curves for Cu-1.5Ti (Fig. 2a) [18] and Cu-4.5Ti (Fig. 2c) [17] are shown here for comparison. In all the alloys, increasing the amount of cold work (0–80%) caused an increase in the peak hardness. The highest hardness values measured were 280 VHN for Cu-1.5Ti, 355 VHN for Cu-2.7Ti, 425 VHN for Cu-4.5Ti and 455 VHN for Cu-5.4Ti alloy. Overageing of the cold worked alloys was marginal at 400 °C and drastic at higher ageing temperatures. For example, the hardness of the 80% cold worked alloys aged at 500 °C for 16 hours was lower than that of the undeformed and aged one (not shown here).

3.2. Tensile properties

The tensile properties of Cu-Ti alloys with prior cold work followed by peak ageing at 400 °C are shown in Fig. 3 (a, b, c and d for Cu-1.5Ti [18], Cu-2.7Ti, Cu-4.5Ti [17] and Cu-5.4Ti alloys respectively). Increasing the amount of cold work raised the yield strength from 350 to 680 MPa for Cu-1.5Ti, 460 to 950 MPa for Cu-2.7Ti, 700 to 1280 MPA for Cu-4.5Ti and 790 to 1400 MPa for Cu-5.4Ti and tensile strength, from 520 to 760 MPa, 680 to 1000 MPa, 890 to 1380 MPa and 930 to 1450 MPa respectively. The elongation however, decreased from 23 to 9% for Cu-1.5Ti, 22 to 3.5% for Cu-2.7Ti, 20 to 2% for Cu-4.5Ti and 15 to 1.5% for Cu-5.4Ti alloy.

3.3. Electrical conductivity (EC)

The influence of ageing at 400 and 450 °C on the electrical conductivity of Cu-2.7Ti and Cu-5.4Ti alloys in undeformed as well as cold worked conditions is shown

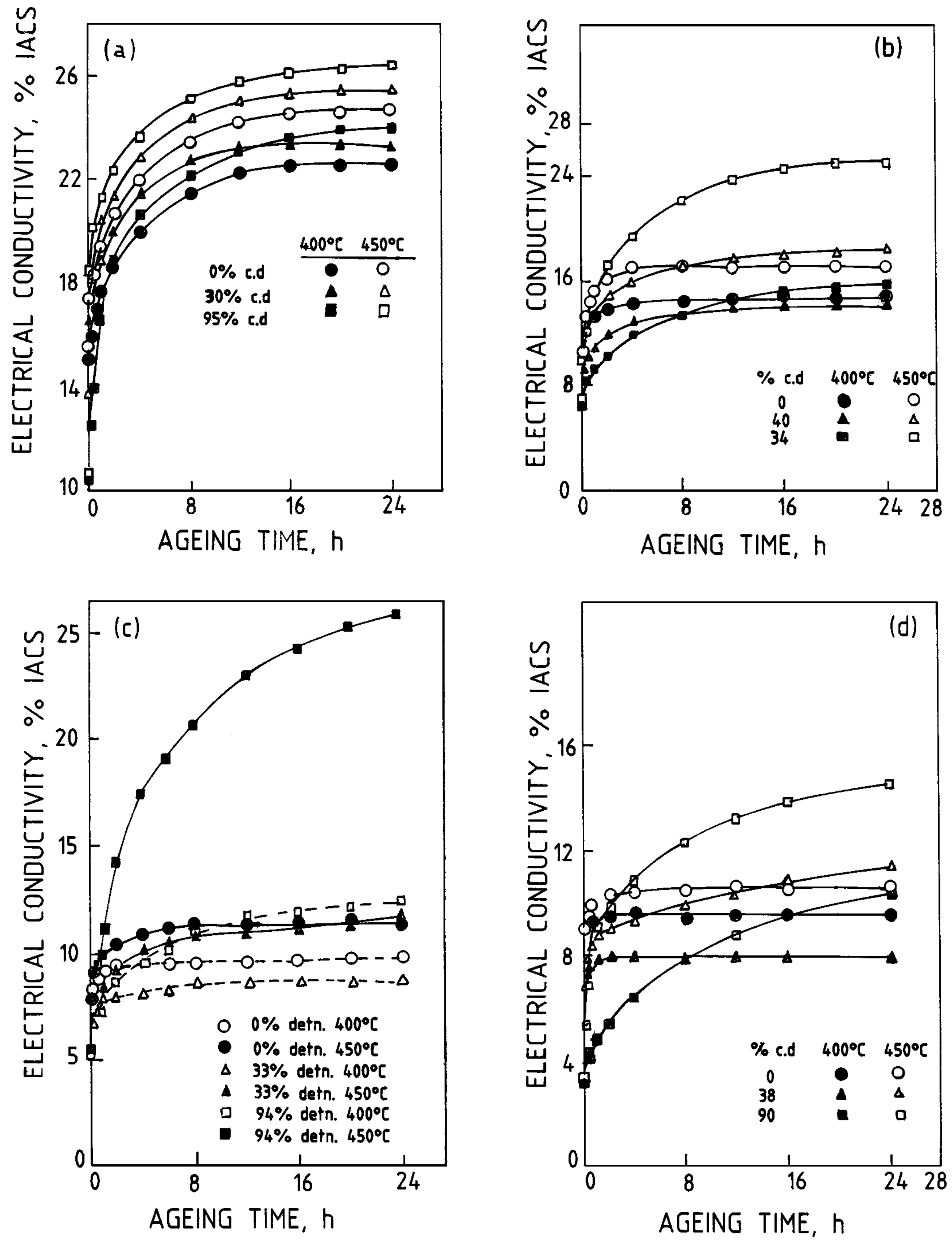

Figure 4 Electrical conductivity of Cu-Ti alloys: (a) Cu-1.5Ti [18]; (b) Cu-2.7Ti; (c) Cu-4.5Ti [17]; and (d) Cu-5.4Ti.

in Fig. 4. The EC curves for Cu-1.5Ti (Fig. 4a) [18] and Cu-4.5Ti (Fig. 4c) [17] are included here for comparison. The EC of the solution treated alloys was found to be 15.5, 10.0, 8.0 and 9.0% IACS (International Annealed Copper Standard) respectively for Cu-1.5Ti, Cu-2.7Ti, Cu-4.5Ti and Cu-5.4Ti alloys. On ageing at 450 °C for 16 hours, it increased to 24.5, 17.0, 11.3 and 10.8% IACS for Cu-1.5Ti, Cu-2.7Ti, Cu-4.5Ti and Cu-5.4Ti alloys respectively. These values remained constant even after 24 hours at 450 °C.

When the solution treated alloys are cold worked, EC reduced to 13.5, 8.5, 7.0 and 6% IACS (after 30% deformation) and 10.5, 7.0, 5.0 and 3% IACS (after 90% cold work). The 30% deformed alloys, on ageing, exhibited an increase in the EC and the maximum values for 450 °C/24 h treatment were found to be 25, 18.4, 11.5 and 11.4% IACS respectively for Cu-1.5Ti, Cu-2.7Ti, Cu-4.5Ti and Cu-5.4Ti alloys. Further, the alloys with 90% deformation also exhibited a similar trend on ageing, showing improvements in the EC of 26.5, 25.6, 25.6 and 14.5% IACS after 24 hours at 450 °C. A maximum EC of 25% IACS in Cu-1.5Ti, Cu-2.7Ti and Cu-4.5Ti alloys and 14.5% IACS in Cu-5.4Ti alloy was achieved with the present treatments.

3.4. Microstructure

The optical microstructure of Cu-2.7Ti and Cu-5.4Ti alloys in solution treated condition (900 °C/2 h/WQ) essentially consisted of equiaxed grains and annealing twins (not shown here). Peakageing at 450 °C for 16 hours has not changed the optical microstructure and it is similar to that of the solution treated alloys. The

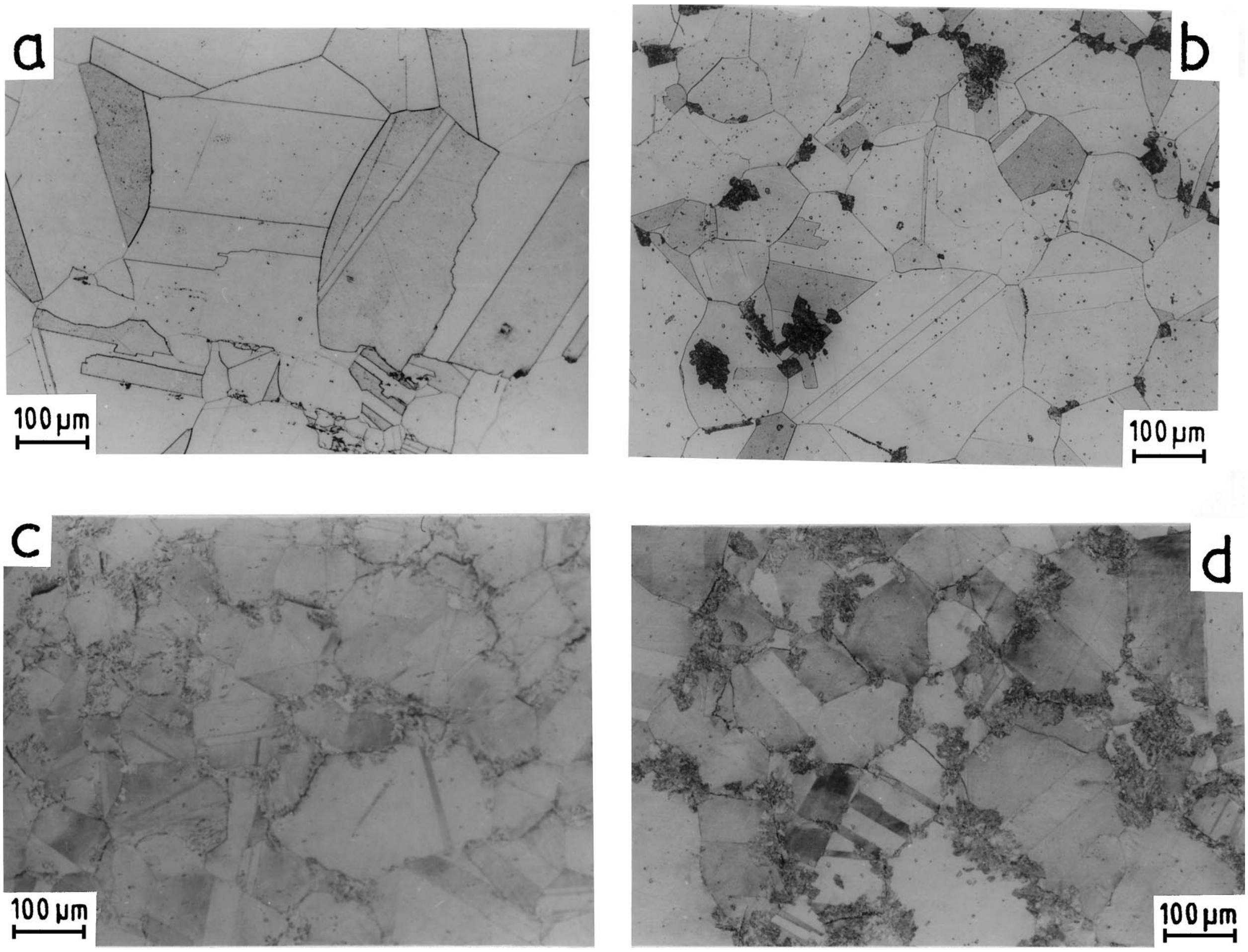

Figure 5 Optical micrographs of Cu-Ti alloys in overaged (500 °C/32 h) condition: (a) Cu-1.5Ti; (b) Cu-2.7Ti; (c) Cu-4.5Ti; and (d) Cu-5.4Ti.

optical micrographs of Cu-Ti alloys overaged at 500 °C for 32 hours are shown in Fig. 5. Overageing has resulted in the formation of discontinuous (lamellar) precipitation at grain boundaries in Cu-2.7Ti, Cu-4.5Ti and Cu-5.4Ti alloys. Such discontinuous precipitation has not been observed in Cu-1.5Ti alloy even after ageing at 500 °C for 32 hours (Fig. 5a). Fig. 6 shows optical microstructures of the Cu-Ti alloys in solution treated and cold worked (90%) condition. The equiaxed grains have been deformed and elongated in the rolling direction. Peak ageing of as-quenched and cold deformed alloys at 400 °C did not lead to any noticeable changes in the optical micrographs. Further, overageing at 400 °C for 16 hours also did not produce any changes in the optical micrographs of the deformed alloys.

The TEM micrographs of solution treated Cu-Ti are shown in Fig. 7. TEM micrographs of Cu-1.5Ti [18], Cu-4.5Ti [17] and Cu-5.4Ti [15] alloys are included here for comparison. Grain boundaries and dislocations are seen in the bright field (BF) images of Cu-1.5Ti and Cu-2.7Ti alloys in Fig. 7a and b respectively. Further, no signs of any precipitation was observed in these alloys. Fine scale precipitation in the form of modulated structure is found in Cu-4.5Ti alloy in Fig. 7c. Clear precipitation can be seen in the BF image in Fig. 7d and dark field (DF) image in Fig. 7e in Cu-5.4Ti alloy. From the Selected Area Diffraction (SAD) pattern in Fig. 7f and its schematic in Fig. 7g, the precipitate has been identified as ordered, metastable and coherent β', Cu_4Ti phase having body centred tetragonal (bct) crystal structure with lattice parameters of $a = 0.584$ nm and $c = 0.362$ nm.

The TEM micrographs of Cu-2.7Ti alloy peak aged at 450 °C for 16 hours are shown in Fig. 8. The BF image in Fig. 8a reveals intertwined microstructure of matrix and precipitate. From the DF image, SAD pattern and its schematic shown in Fig. 8b, c and d respectively, the precipitate has been identified as β', Cu_4Ti phase having bct structure. Similar DF images and SAD patterns have also been obtained for Cu-5.4Ti alloy (not shown here) which indicate that the precipitate in peak aged (450 °C/16 h) condition is β' Cu_4Ti only. Fig. 9 shows TEMs of overaged (450 °C/50 h) Cu-5.4Ti alloy. Fig. 9a and b show BF and DF images respectively of the equilibrium phase while SAD in Fig. 9c and its schematic in Fig. 9d identify the precipitate as β, Cu_3Ti with orthorhombic structure having lattice parameters of $a = 0.5162$ nm, $b = 0.4347$ nm and $c = 0.453$ nm. The TEM micrographs of overaged (450 °C/50 h) Cu-2.7Ti alloy (not shown here) also reveal the presence of equilibrium precipitate β, Cu_3Ti with orthorhombic structure having the above lattice parameters.

Fig. 10 shows TEM micrographs of solution treated and cold worked (25%) Cu-2.7Ti and Cu-5.4Ti alloys compared with those of Cu-1.5Ti [18] and Cu-4.5Ti [17] alloys. Dislocation cell structure is seen in BF images in Cu-1.5Ti and Cu-2.7Ti alloys in Fig. 10a and b whereas deformation twins have been observed in Cu-4.5Ti and Cu-5.4Ti alloys in Fig. 10c and d. The SAD

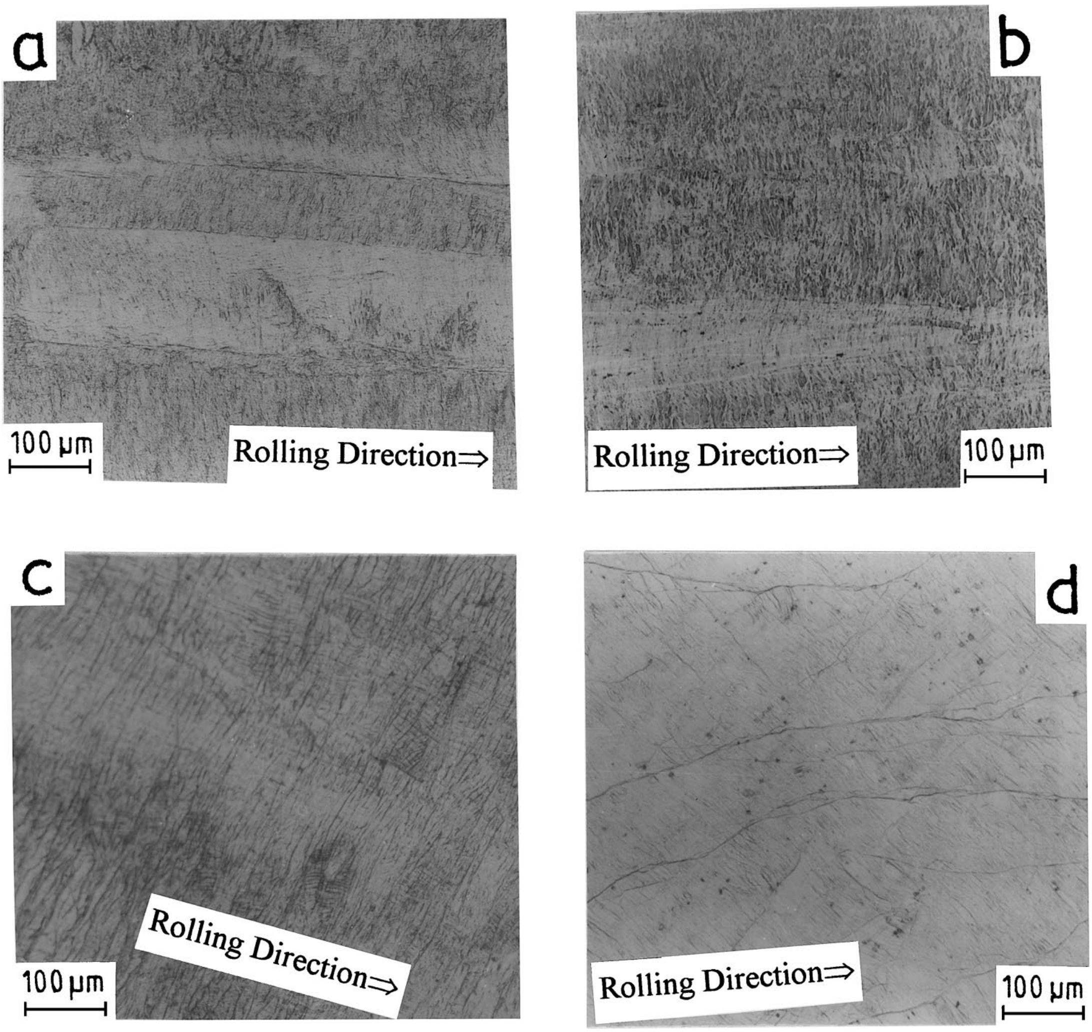

Figure 6 Optical micrographs of solution treated and cold worked (90%) Cu-Ti alloys: (a) Cu-1.5Ti [18]; (b) Cu-2.7Ti; (c) Cu-4.5Ti [17]; and (d) Cu-5.4Ti.

patterns for twins are seen in the inset of Fig. 10c and d and their schematics in Fig. 10e and f. Similar microstructures were observed in the alloys with 90% cold deformation also. TEM micrographs of solution treated and cold worked (90%) + peak aged (400 °C) Cu-Ti alloys are shown in Fig. 11. The BF images in Fig. 11a and b show fine scale precipitation of β', Cu_4Ti phase in Cu-1.5Ti and Cu-2.7Ti alloys whereas those in Fig. 11c and d reveal β', Cu_4Ti precipitate as well as deformation twins in Cu-4.5Ti and Cu-5.4Ti alloys respectively. The inset of Fig. 11c and d shows SAD patterns for twins and their schematics in Fig. 11e and f.

Fig. 12 shows scanning electron microscopy (SEM) fractographs of the tensile tested Cu-Ti alloys in solution treated + cold worked (90%) + peak aged (400 °C) condition. Ductile mode of fracture showing dimples are seen in these alloys even after 90% cold deformation and ageing at 400 °C.

4. Discussion

4.1. Mechanical properties and electrical conductivity

The present study confirms that Cu-Ti alloys are age-hardenable with substantial improvements in hardness and strength. In the as-quenched and peak aged (450 °C/16 h) condition, the hardness of Cu-Ti alloys has significantly increased (Fig. 1) due to precipitation of ordered, metastable and coherent β', Cu_4Ti phase (Fig. 8) which is in agreement with the reported results by Hakkarainen [5], Laughlin and Cahn [7], Datta and Soffa [8], Wagner [9] and Nagarjuna *et al.* [17, 18]. The hardness of Cu-Ti alloys has further increased with cold work followed by peak ageing at 400 °C (Fig. 2). A similar behaviour was reported by Dutkiewicz [11] in Cu-2.4wt%Ti and Cu-4.29wt%Ti alloys, Saji and Hornbogen [12] in a Cu-4.0wt%Ti alloy and Nagarjuna *et al.* in Cu-4.5wt%Ti [17] and Cu-1.5wt%Ti [18] alloys. Yield and tensile strengths have followed the trend of hardness exhibiting an increase on ageing of the undeformed as well as prior cold worked alloys. The ductility however, decreased. The strengthening mechanism dominating in the peak aged Cu-Ti alloys is precipitation hardening. Coherency strains are developed due to the precipitation of ordered, metastable and coherent β' Cu_4Ti phase. When a solution treated Cu-Ti alloy is cold worked prior to ageing, dislocation density increases which enhances the strength (due to work hardening). Further, on ageing the deformed alloy at a low temperature, a fine dispersion of β' Cu_4Ti precipitate with increased volume fraction forms, thereby resulting in increased strengthening of the alloy. Prolonged ageing or ageing at higher temperatures results

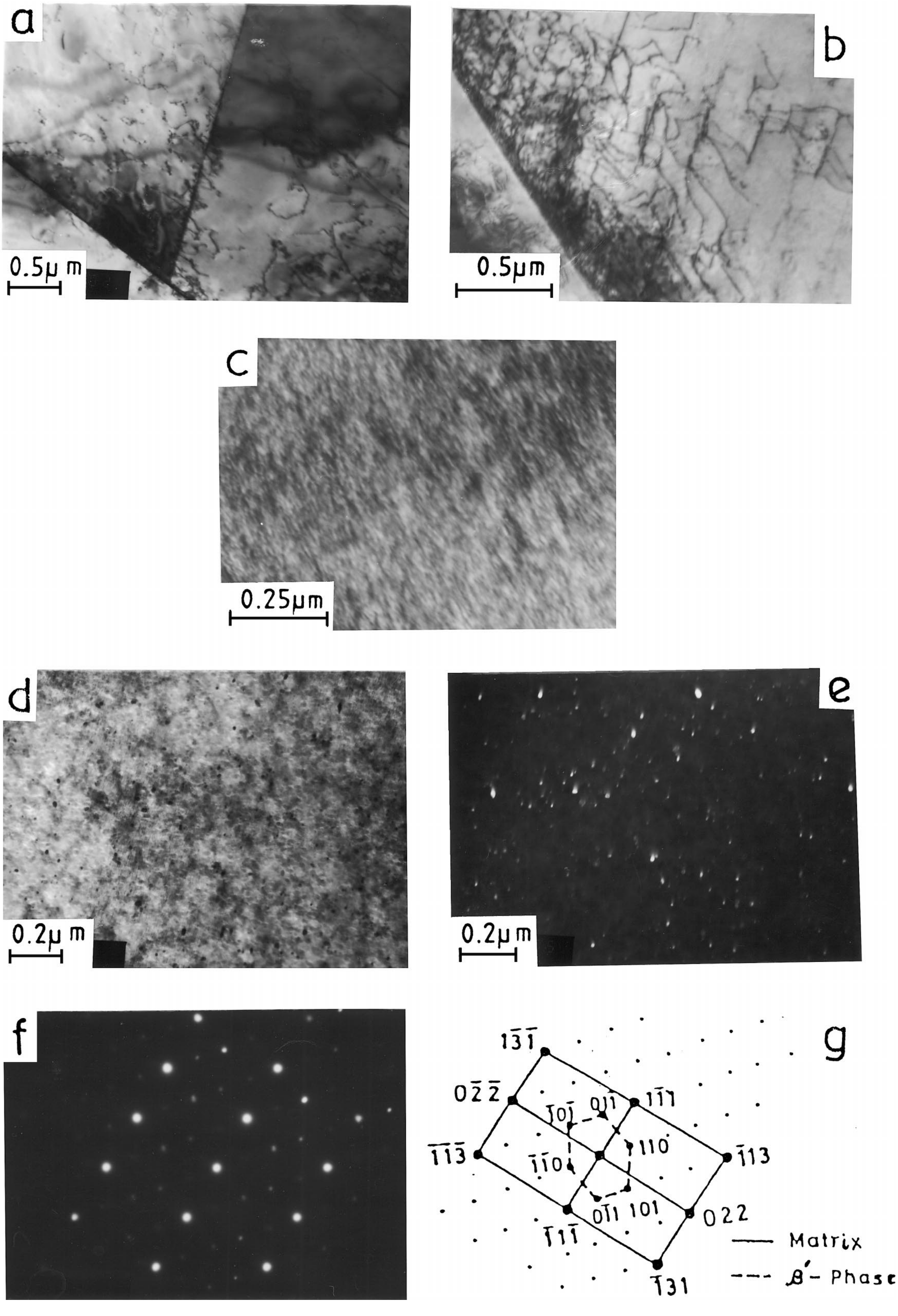

Figure 7 TEM micrographs of solution treated Cu-Ti alloys: (a) Cu-1.5Ti [18]; (b) Cu-2.7Ti; (c) Cu-4.5Ti [17]; (d) BF image; (e) DF image; (f) SAD pattern; and (g) Schematic of the SAD pattern of Cu-5.4Ti [15].

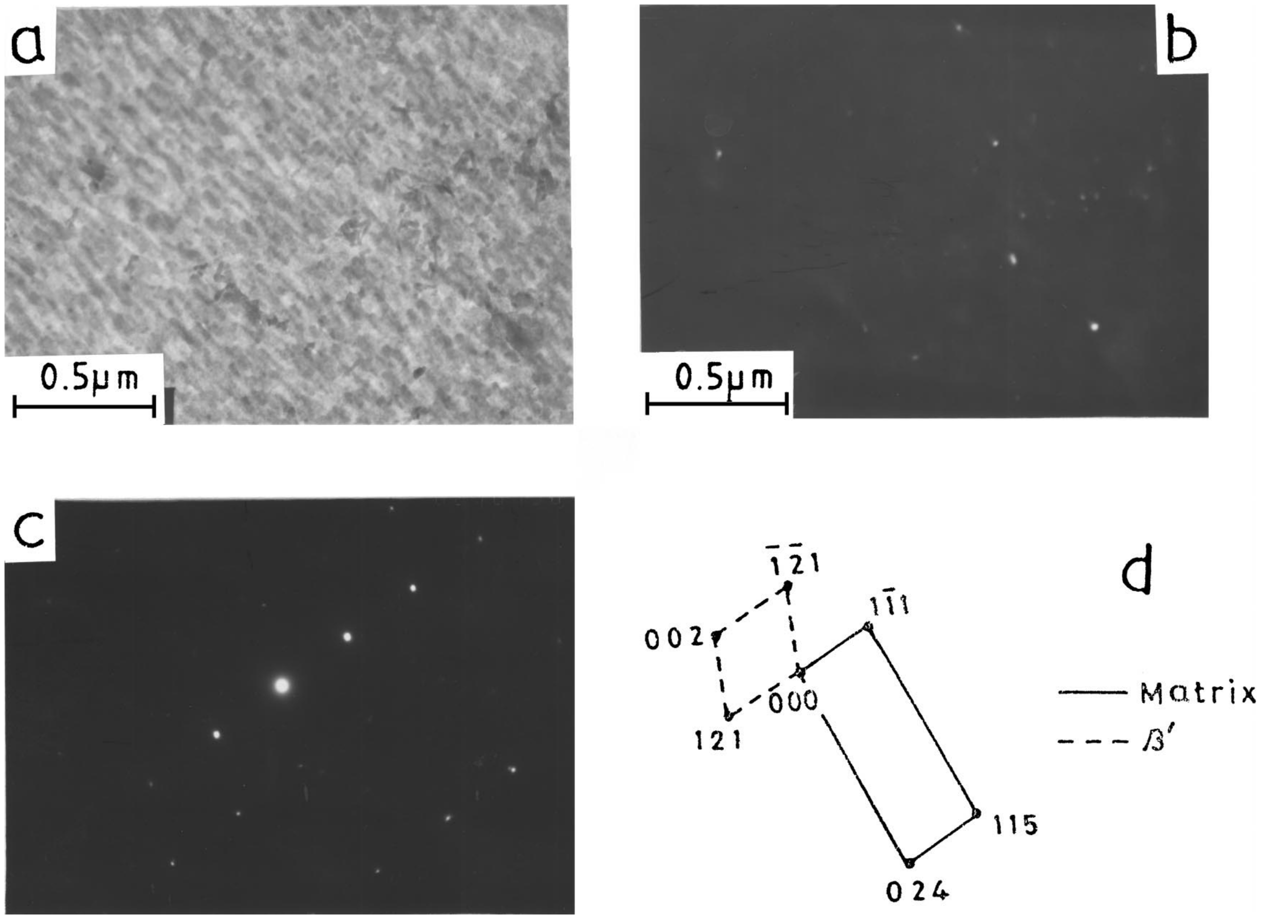

Figure 8 TEM micrographs of peak aged (450 °C/16 h) Cu-2.7Ti alloy: (a) BF image; (b) DF image; (c) SAD pattern; and (d) Schematic of the SAD pattern.

in the formation of equilibrium precipitate β, leading to softening of the alloy.

The mechanical properties and electrical conductivity of Cu-5.4Ti alloy are compared with those reported for Cu-4.3Ti alloy by Saarivirta [1] and Cu-4.5Ti by Nagarjuna *et al*. [17] in Table I. Further, the properties of Cu-2.7Ti alloy are compared with those of Cu-1.5Ti [18] and Cu-0.5Be-2.5Co(17500) [23] in Table II and those of Cu-5.4Ti alloy with Cu-4.5Ti [17] and Cu-2.0Be-0.5Co (C17200 or 17300) [23] in Table III. The mechanical properties obtained in the present study compare well with the reported results. The electrical conductivity of Cu-2.7Ti alloy is less than that of both Cu-1.5Ti and Cu-0.5Be-2.5Co alloy (Table II) and therefore, needs further improvement in order to compare favourably with the commercial Cu-Be-Co alloy for high conductivity applications. However, the EC of Cu-4.5Ti alloy in 90% cold worked and aged

TABLE I Comparison of properties of Cu-5.4Ti alloy with those reported for Cu-4.3Ti [1] and Cu-4.5Ti [17] alloys

		Cu-5.4Ti			Cu-4.5Ti			Cu-4.3Ti		
Sl. no.	Property	ST	ST + PA[c]	ST + 90 CW + PA[d]	ST	ST + PA[c]	ST + 90 CW + PA[d]	ST	ST + aged[e]	ST + 85cd + aged[f]
1.	Yield strength[a] (MPa)	590	790	1400	440	700	1280	NA	NA	NA
2.	Tensile strength (MPa)	780	930	1450	680	890	1380	480	893	1442
3.	Elongation[b] (%)	23	15	1.5	29	20	2	33	16	NIL
4.	Hardness (VHN)	310	366	455	245	340	425	130	280	390[g]
5.	Electrical conductivity (%IACS)	9	10.8	4.8 14.5[h]	8	11.3	8 25.6[h]	3	18	10[g]

[a] 0.2% offset.
[b] GL: 25 mm for Cu-4.5 and 5.4Ti and 50 mm for Cu-4.3Ti.
[c] Aged at 450 °C for 16 hours.
[d] Aged at 400 °C for 1 hour.
[e] Aged at 450 °C for 3 hours.
[f] Aged at 400 °C for 2 hours.
[g] ST 56% c.d. + aged at 400 °C for 4 hours.
[h] aged at 450 °C for 24 hours.
ST: Solution treated; PA: Peak aged; CW: Cold worked by rolling (%); cd: Cold drawn (%); NA: Not available.

TABLE II Comparison of properties of Cu-2.7Ti alloy with those reported for Cu-1.5Ti [18] and Cu-0.5Be-2.5Co [23] alloys

		Cu-2.7Ti			Cu-1.5Ti			Cu-0.5Be-2.5Co		
Sl. no.	Property	ST	ST + PA	ST + 90 CW + PA	ST	ST + PA	ST + 90 CW + PA	ST	ST + PA	ST + full hard + aged
1.	Yield strength[a] (MPa)	192	460	950	112	350	670	140–205	550–690	690–825
2.	Tensile strength (MPA)	430	680	1000	292	520	760	240–380	690–825	760–895
3.	Elongation[b] (%)	36	22	3.5	41	23	9	20–35	10–20	8–15
4.	Hardness (VHN)	120	275	355	75	210	280	72–92	188–215	206–225
5.	Electrical conductivity	10	17	12 25[c]	15.5	24.5	21.5 26.5[c]	20–30	45–60	50–60

[a] 0.2% offset for Cu-1.5Ti and Cu-2.7Ti.

[b] GL 25 mm for Cu-Ti alloys and 50 mm for Cu-Be alloys.

[c] Values for Cu-1.5Ti and Cu-.27Ti deformed alloys aged at 450 °C for 24 hours.

ST: Solution treated; PA: Peak aged; CW: Cold worked by rolling (%).

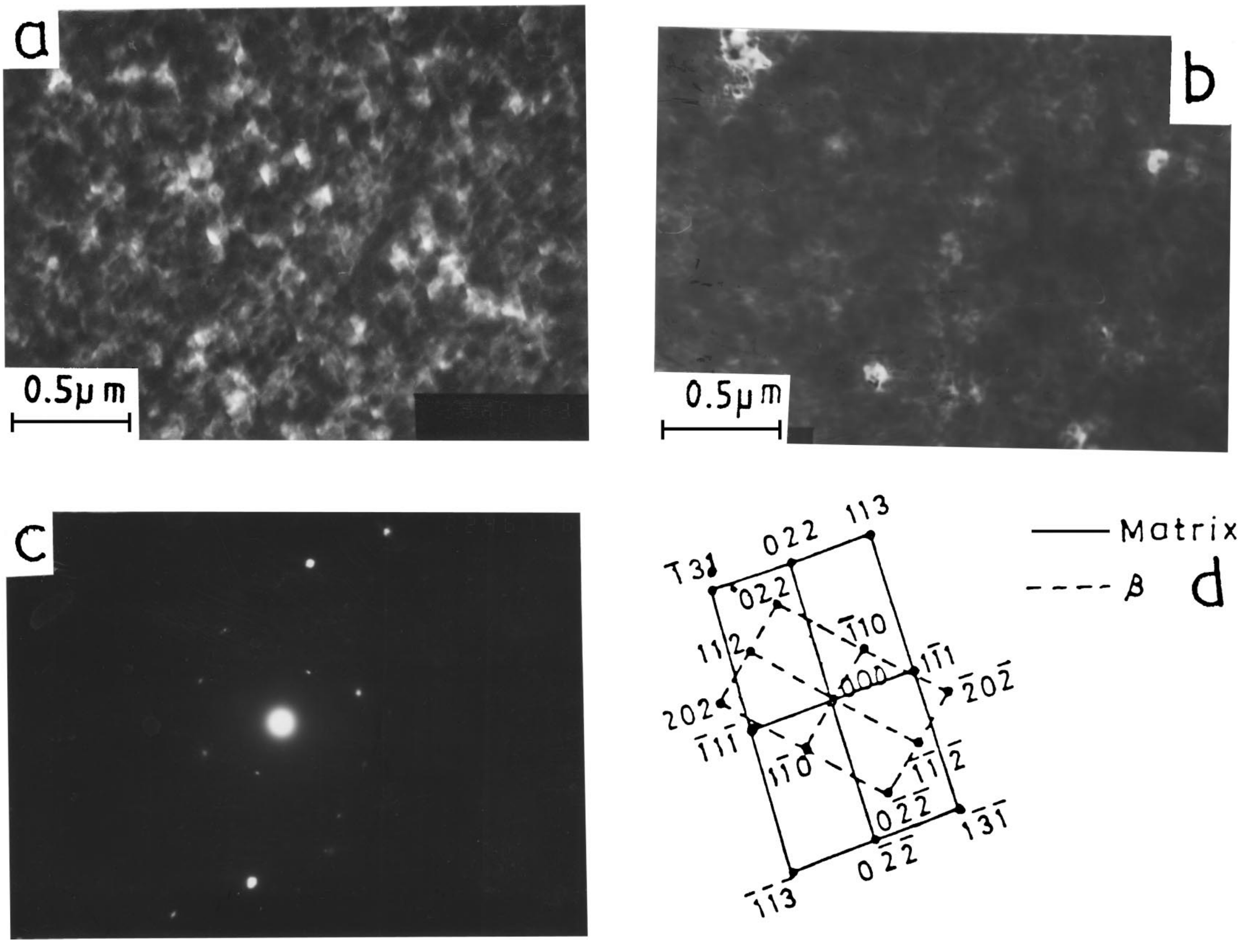

Figure 9 TEM micrographs of overaged (450 °C/50 h) Cu-5.4Ti alloy: (a) BF image; (b) DF image; (c) SAD pattern; and (d) Schematic of the SAD pattern.

(450 °C/24 h) condition is comparable with that of Cu-2.0Be-0.5Co alloy (Table III). Though, the EC is comparable, strength may be reduced due to overageing and hence, a compromise has to be reached between the strength and EC, depending on the application. Further, Cu-5.4Ti alloy should be used only in those applications which require high strength and low conductivity, as its EC is one fifth of that of the Cu-2.0Be-0.5Co alloy.

4.2. Microstructure

In solution treated condition, Cu-Ti alloys exhibit different kinds of microstructure (Fig. 7) depending on the Ti content. While alloys with low Ti content (Cu-1.5Ti and Cu-2.7Ti) show no signs of any precipitation, high Ti alloys i.e., Cu-4.5Ti and Cu-5.4Ti, reveal modulated structure and β', Cu_4Ti precipitate respectively (Fig. 7a–g), formed already during quenching. This observation i.e., formation of fine scale precipitation in high Ti alloys during quenching is in agreement with the reported behaviour by Cornie *et al.* [6], Laughlin and Cahn [7], Datta and Soffa [8], Biehl and Wagner [10] and Nagarjuna *et al.* [14, 15, 17, 19, 20]. In the peak aged condition (450 °C/16 h), maximum strengthening corresponds to the formation of ordered, metastable and coherent Cu_4Ti, β' precipitate having bct structure with lattice parametres of $a = 0.584$ nm and $c = 0.362$ nm

TABLE III Comparison of properties of Cu-5.4Ti alloy with those reported for Cu-4.5Ti [17] and Cu-2.0Be-0.5Co [23] alloys

		Cu-5.4Ti			Cu-4.5Ti			Cu-2.0Be-0.5Co		
Sl. no.	Property	ST	ST + PA	ST + 90 CW + PA	ST	ST + PA	ST + 90 CW PA	ST	ST + aged	ST + full hard + aged
1.	Yield strength[a] (MPa)	590	790	1400	440	700	1280	195–380	965–1205	1140–1415
2.	Tensile strength (MPa)	780	930	1450	680	890	1380	415–540	1140–1310	1310–1480
3.	Elongation[b] (%)	23	15	1.5	29	20	2	35–60	4–10	1–4
4.	Hardness (VHN)	310	366	455	245	340	425	60–90	343	385
5.	Electrical conductivity (%IACS)	9	10.8	4.8 14.5[c]	8	11.3	8 25.6[c]	19	25	25

[a] 0.2% offset.

[b] GL: 25 mm for Cu-Ti alloys and 50 mm for Cu-Be alloys.

[c] Values reported for Cu-4.5Ti and Cu-5.4 Ti deformed alloys aged at 450 °C for 24 hours.

ST: Solution treated; PA: Peak aged; CW: Cold worked by rolling (%).

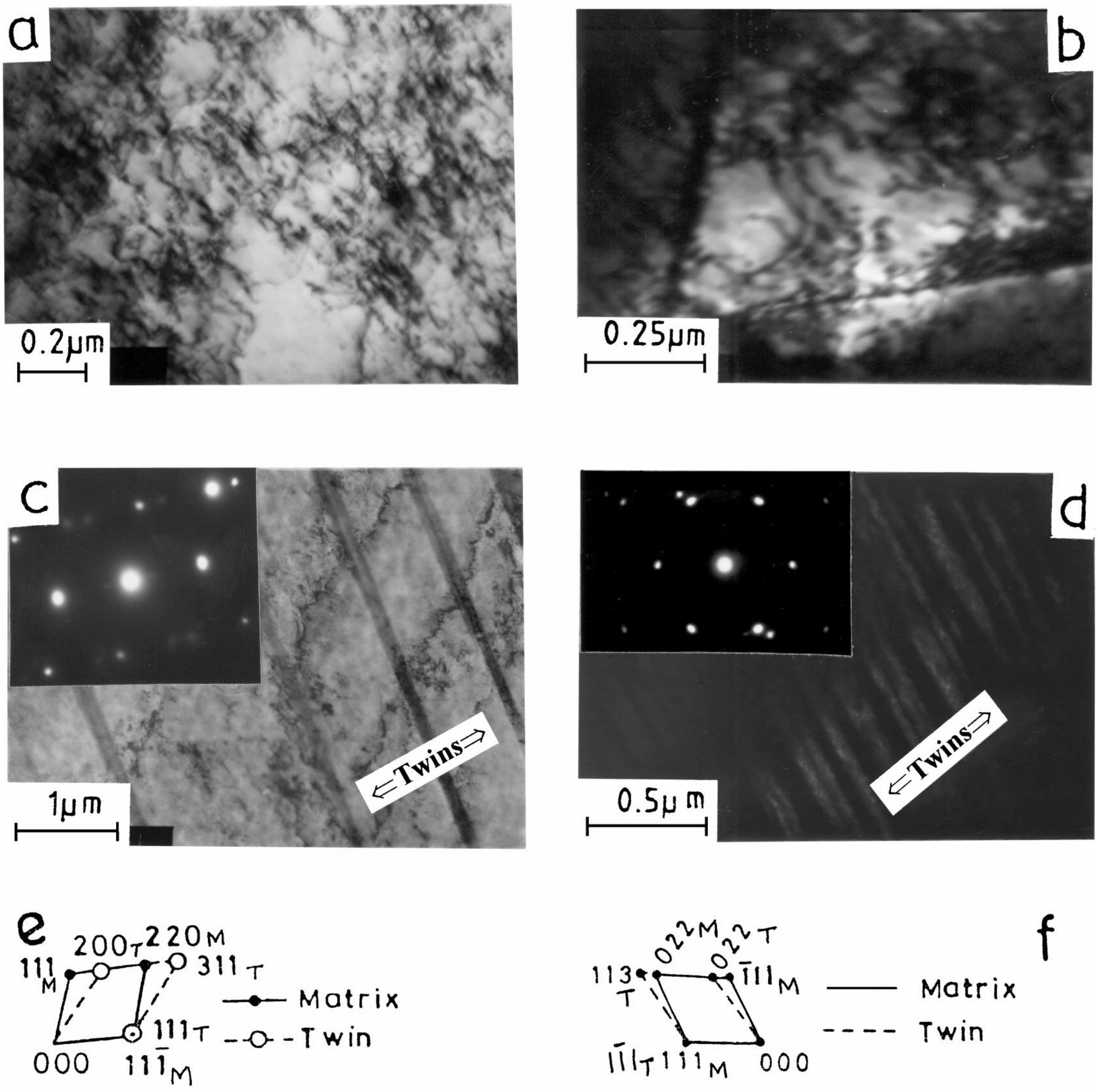

Figure 10 TEM micrographs of solution treated and cold worked (25%) Cu-Ti alloys: (a) Cu-1.5Ti [18]; (b) Cu-2.7Ti; (c) Cu-4.5Ti [17], c inset: SAD pattern for twins; (d) Cu-5.4Ti, d inset: SAD pattern for twins; (e) Schematic of the SAD pattern in c inset; and (f) Schematic of the SAD pattern in d inset.

(Fig. 8), which is in agreement with Hakkarainen [5], Laughlin and Cahn [7], Dutta and Soffa [8], Wagner [9] and Nagarjuna *et al.* [17, 18]. Overageing at 450 °C for 50 hours resulted in the formation of metastable and coherent β', Cu_4Ti phase in Cu-1.5Ti alloy [18] whereas equilibrium phase β, Cu_3Ti having orthorhombic structure with lattice parameters of $a = 0.5162$ nm, $b = 0.4347$ nm and $c = 0.453$ nm was observed in

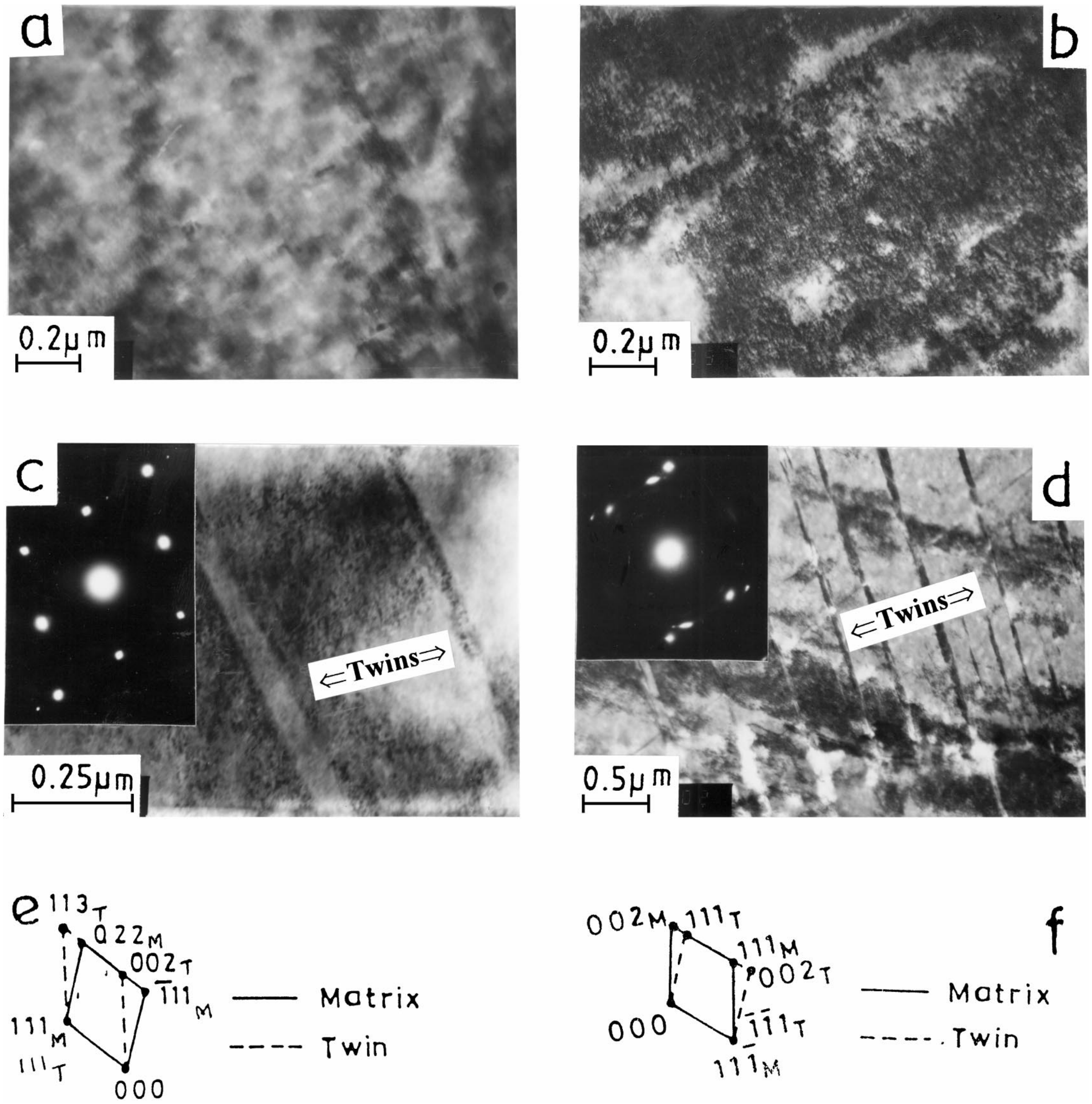

Figure 11 TEM micrographs of solution treated + cold worked (90%) + peak aged (400 °C) Cu-Ti alloys: (a) Cu-1.5Ti [18]; (b) Cu-2.7Ti; (c) Cu-4.5Ti [17], c inset: SAD pattern for twins; (d) Cu-5.4Ti, d inset: SAD pattern for twins; (e) Schematic of the SAD pattern in c inset; and (f) Schematic of the SAD pattern in d inset.

Cu-2.7Ti, Cu-4.5Ti [17] and Cu-5.4Ti (Fig. 9) alloys. In Cu-1.5Ti alloy however, the equilibrium phase β is found to form at higher ageing temperatures, but in shorter times (500 °C/16 h) [18]. Similar structure and lattice parameters for the equilibrium phase β, have been reported by Nagarjuna *et al.* [17, 18], Karlsson [24] and Knights and Wilkes [25]. However, Ecob *et al.* [26, 27] have shown that the equilibrium phase β is also orthorhombic but with different lattice parameters ($a = 0.453$, $b = 0.434$ and $c = 1.293$ nm) and composition (Cu_4Ti). Earlier workers [6, 25] attributed overageing to discontinuous or cellular precipitation leading to lamellar distribution of α and β phases. Such discontinuous precipitation has not been observed in the alloys overaged at 500 °C for 32 hours (Cu-1.5Ti) or at 450 °C for 50 hours (Cu-2.7Ti, Cu-4.5Ti and Cu-5.4Ti). However, the discontinuous precipitation leading to lamellar distribution of α and β phases has been found to occur in Cu-2.7Ti, Cu-4.5Ti and Cu-5.4Ti alloys overaged at 500 °C for 32 hours (Fig. 5). Ecob *et al.* [27] reported that cellular precipitation starts during ageing at 500 °C for 333 hours in a Cu-1wt%Ti alloy and 700 °C for 10 minutes in a Cu-3.0wt%Ti alloy. In solution treated and cold worked condition, low Ti alloys show dislocations whereas high Ti alloys exhibit deformation twins (Fig. 10). Saito [28] and Michels *et al.* [29] also reported deformation twins in Cu-Ti alloys. When the as-quenched and cold worked alloys are aged at 400 °C, metastable and coherent Cu_4Ti, β' precipitate forms (Fig. 11), leading to extensive strengthening of the alloy. Prolonged ageing results in the formation of equilibrium precipitate β, which reduces the strength.

Regarding the fracture behaviour, Cu-2.7Ti and Cu-5.4Ti alloys deformed by shear mode of fracture (microvoid coalescence) in the ST + peak

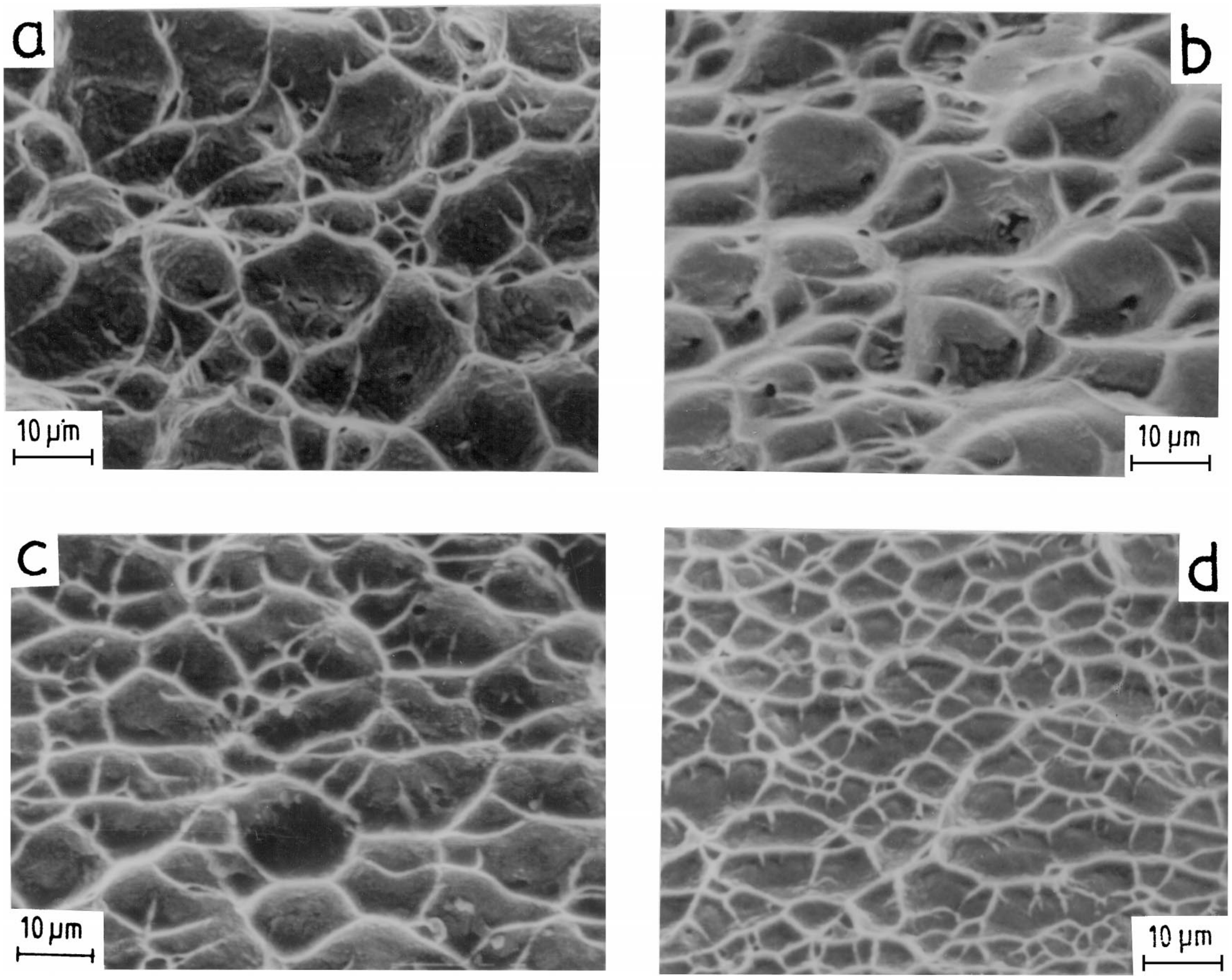

Figure 12 SEM fractographs of the tensile tested Cu-Ti alloys in solution treated + cold worked (90%) + peak aged (400 °C) condition: (a) Cu-1.5Ti [18]; (b) Cu-2.7Ti; (c) Cu-4.5Ti [17]; and (d) Cu-5.4Ti.

aged (not shown here) as well as ST + CW + peak aged conditions (Fig. 12). The observation of nearly equiaxed dimples even in the 90%CW + peak aged condition confirms the ductile mode of fracture in these alloys. Ductile mode of fracture has been reported by us in Cu-4.5Ti [17] and Cu-1.5Ti [18] alloys in CW(90%) + peak aged condition.

5. Conclusions

(1) Peak ageing (450 °C/16 h) of as-quenched Cu-2.7Ti and Cu-5.4Ti alloys increases their hardness and strength considerably; hardness from 120 and 310 VHN to 275 and 366 VHN and tensile strength from 430 and 780 MPa to 680 and 930 MPa respectively.

(2) Cold work (90%) + peak ageing (400 °C) further enhances their hardness and strength; hardness to 355 and 455 VHN and tensile strength to 1000 and 1450 MPa respectively for Cu-2.7Ti and Cu-5.4Ti alloys.

(3) Electrical conductivity also increases on ageing the undeformed as well as prior cold worked alloys. A maximum electrical conductivity of 25% IACS for Cu-2.7Ti and 14.5% IACS for Cu-5.4Ti alloy is obtained with the present treatments.

(4) Precipitation of β', Cu_4Ti phase with bct structure in peak aged condition is responsible for maximum strength in Cu-2.7Ti and Cu-5.4Ti alloys. Overageing results in the decrease of hardness and strength due to precipitation of equilibrium phase β, Cu_3Ti.

(5) The mechanical properties of Cu-Ti alloys are comparable and electrical conductivity is less than that of commercial Cu-Be-Co alloys. Hence, Cu-Ti alloys can be used in those applications where high strength and low electrical conductivity are the most desirable properties.

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