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Effect of microstructures on deformation behaviour of high-strength low-alloy steel

S. K. Das · S. Chatterjee · S. Tarafder

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Abstract The role played by microstructural constituents of high-strength low-alloy (HSLA) steel in controlling the deformation processes has been studied. The steel was solution treated and water quenched followed by ageing at various temperatures. Microstructural characterization has been carried out by using scanning electron microscope and transmission electron microscope. Tensile tests were conducted as per ASTM standard at constant displacement rate. The conditions under which microvoid coalescence was suspended in spite of a constant resident population of void initiating carbide and carbo-nitride particles have been explained. The major role played by the coherency of Cu precipitates in controlling dislocations movement; and hence, plastic flow is thought to be responsible for the effects observed.

Introduction

Copper bearing high-strength low-alloy (HSLA) steels can provide various combinations of strength and toughness when the microstructure is engineered through thermal and thermomechanical processing such as rolling and ageing, normalizing and ageing and quenching and ageing [1–8]. Cu is used as an alloying element in these steels due to its precipitation hardening effect, and it contributes significantly increasing yield strengths without impairing weldability [9]. The machinability and corrosion resistance are improved in stainless steel by addition of Cu [10]. In microalloyed steel, increasing the amount of Cu increases the martensitic lath constituents and enhanced the strength [11]. A substantial amount of work has been carried out on the ductile fracture behaviour as well as hydrogen embrittlement of Cu-containing HSLA steels [12, 13]. However, the influence of microstructural constituents on the deformation behaviour of these steels has not yet been fully explored.

Microstructural variations in Cu-strengthened HSLA steel have been imparted through ageing at various temperatures after an initial quenching treatment. Descriptions of the various microstructures have been obtained through detailed scanning and transmission electron microscopy. These were used to correlate the tensile properties obtained through standard tests.

S. K. Das (✉) · S. Tarafder
MST Division, National Metallurgical Laboratory (Council of Scientific and Industrial Research), Jamshedpur 831007, India
e-mail: sapan@nmlindia.org

S. Chatterjee
Department of Metallurgical and Material Engineering, Bengal Engineering and Science University, Shibpur, Howrah 711103, India

Experimental

Material

The Cu-strengthened HSLA-100 steel used for this study was of a composition as given in Table 1. The material was available in the form of 50-mm thick plate from which blanks were machined with their axes oriented along the rolling direction for the fabrication of tensile specimens and metallographic samples.

Solution treatment and ageing

Specimens were solutionized at 910 °C for 1 h, and subsequently water quenched (WQ). The quenched specimens

Table 1 Chemical composition of HSLA steels in weight percent

Steel	C	Mn	P	S	N	Si	Cr	Mo	Ti	V	Nb	Ni	Cu	Fe
HSLA-100	0.04	0.90	0.01	0.005	0.015	0.25	0.6	0.60	0.02	0.03	0.03	3.50	1.60	Bal

were further aged at various temperatures from 350 to 750 °C in steps of 50 °C for 1 h, and then air-cooled to room temperature in order to produce microstructural variations.

Microstructural characterization

Metallographic samples were prepared by conventional grinding and polishing techniques for microstructural characterization. Polished specimens were etched with 4% nital for observation in a scanning electron microscope (SEM). Thin foils were prepared in a twin jet electro-polishing equipment for transmission electron microscopy (TEM). A SEM (JEOL JSM 840A with Noran Quest EDS system) and an analytical TEM (Philips CM 200 with TSL EDAX system) were used for microstructural characterization.

Tensile testing

The tensile properties of variously aged specimens were obtained by tensile tests conducted as per ASTM standard E-8M [14] using a 100 kN capacity servo-hydraulic testing machine (INSTRON 8562). Round specimens of 28 mm parallel length and 5 mm gauge diameter were tested at a constant displacement rate of 3×10^{-3} mm/s at room temperature (28 °C), employing a 25-mm gauge length extensometer for measurement of strain up to fracture.

Results and discussion

Evolution of microstructures in HSLA-100 steel due to ageing

Secondary electron images of microstructures at the variously aged conditions of the HSLA-100 steel under investigation are shown in Fig. 1a–d. In WQ condition, the microstructure of the steel is seen to consist of a mixture of acicular ferrite, bainite and lath martensite. Figure 2 is a bright field TEM image of the WQ sample. Though morphological appearance of bainite and acicular ferrite are similar, carbides are mainly distributed in bainitic regions. Figure 2b is a TEM micrograph of $M_{23}C_6$ carbides in WQ sample. The lath martensite formed on quenching was found to get tempered progressively on ageing at increasing temperatures. The entire range of ageing temperatures could be divided into three regimes based on the observations of microstructural changes following a definite pattern in each regime. TEM study revealed the process of

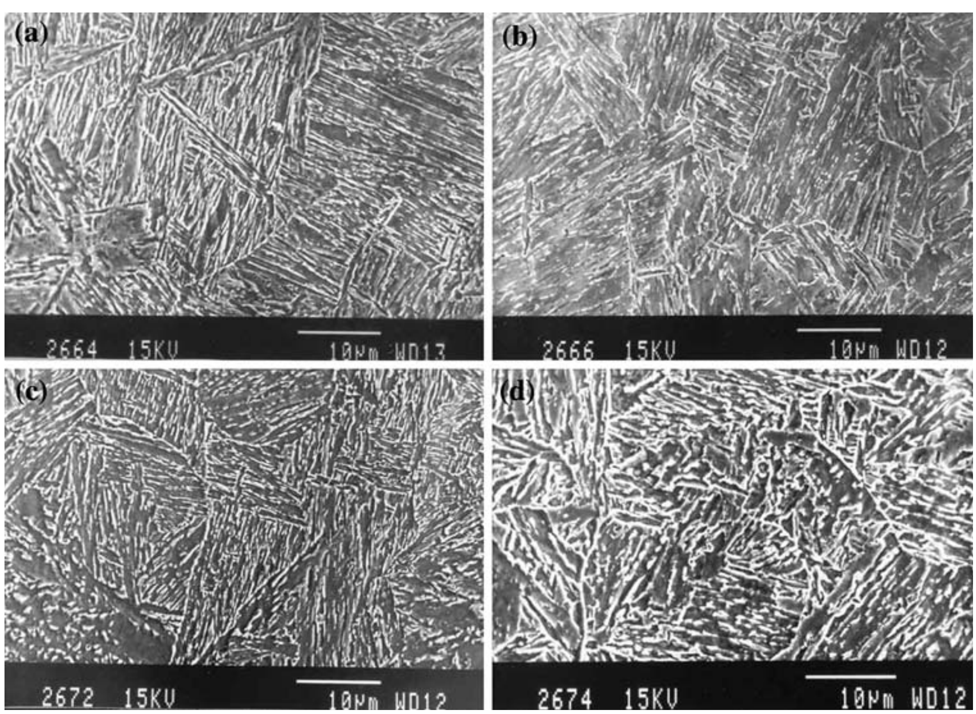

Fig. 1 SEM micrographs of various heat treated conditions of Cu-strengthened HSLA-100 steel. **a** WQ condition, **b** WQ and aged at 500 °C, **c** WQ and aged at 650 °C, and **d** WQ and aged at 700 °C

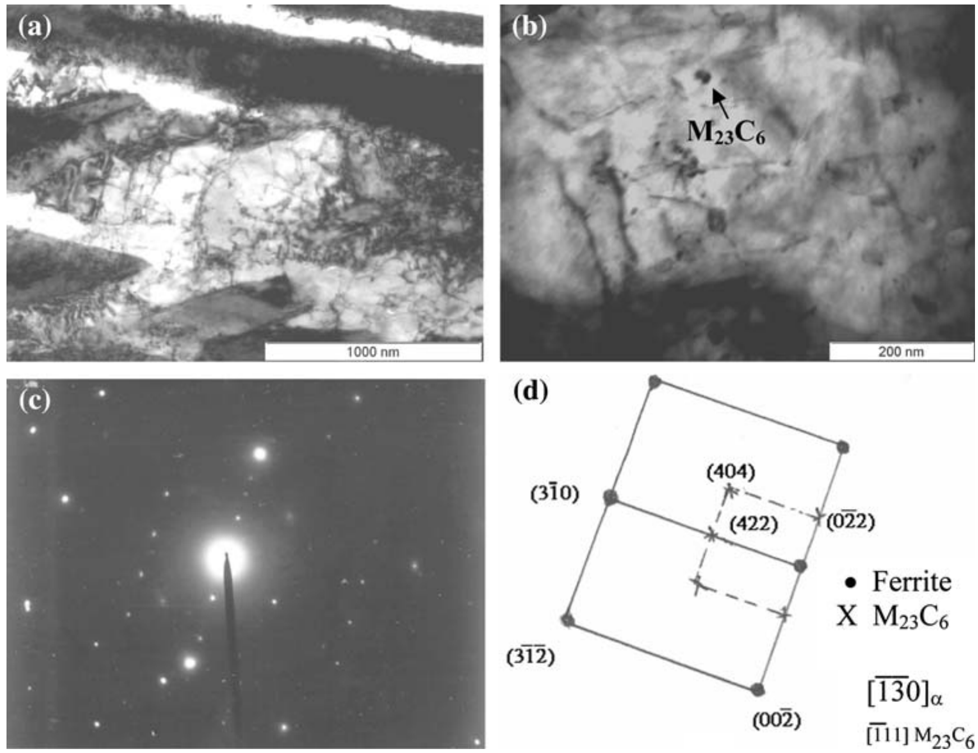

Fig. 2 TEM of WQ HSLA-100 steels showing: **a** lath martensite and acicular ferrite and **b** M23C6 carbides; **c** SADP of M23C6 carbides and **d** schematic representation of (**c**)

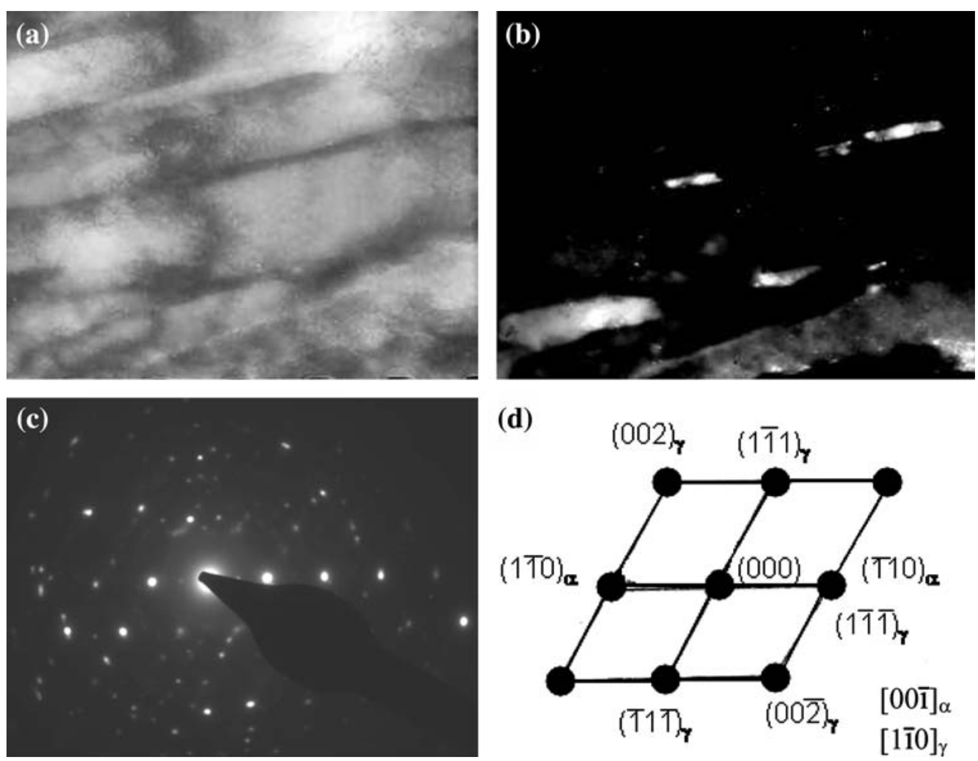

Fig. 3 TEM of WQ HSLA-100 steels showing retained austenite: **a** bright field and **b** dark field image; **c** SADP of retained austenite and **d** schematic representation of (**c**)

evolution of various microstructural constituents in the steel on increasing the intensity of ageing by increasing the ageing temperature. Lath martensite and acicular ferrite were found to be stable up to 500 °C. Beyond this temperature, martensite was tempered, and recovery of ferrite occurred. A small amount of austenite was retained in the quenched condition (WQ), which continued to persist throughout the temperature range of ageing, as shown in Fig. 3. Above 675 °C, as the ageing temperature exceeds the A_{C1} temperature, ferritic constituents in small local

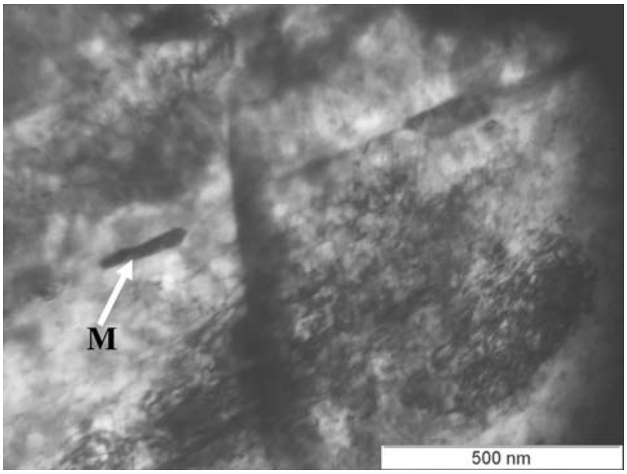

Fig. 4 TEM image of sample aged at 700 °C showing newly formed martensite

regions were transformed to reverted austenite, and, on subsequent cooling, a part of the reverted austenite was converted to small martensite islands, as shown in Fig. 4. Fine carbides and carbonitrides remained unchanged during ageing.

Upon quenching after solution treatment of the HSLA steel at 910 °C, low carbon lath martensite along with acicular ferrite and bainite were formed easily. The maximum solubility of Cu in iron at 850 °C is 2.11 wt% [15]. A super saturated solid solution of Cu in the steel was formed on quenching of the steel. A higher Cu content (1.6 wt%), together with Mn, reduces the A_{C1} transformation temperature [16], and retained austenite is observed in the TEM in between martensite laths. Such retained austenite is very stable due to the presence of γ-iron stabilisers like Cu, Ni and Mn in the system. Carbides and carbo-nitrides, like $M_{23}C_6$, Mo_2C and Nb(CN), were also observed in the WQ microstructure. These stable carbides and carbo-nitrides are formed during solidification and thermo-mechanical processing of the steel, at temperatures above 1,000 °C [17]. Ageing for 1 h is not sufficient to change the morphology and composition of these stable particles.

During ageing, Cu precipitates out from the super-saturated solution and increases the hardness and strength properties through precipitation hardening mechanism. Cu is in the solid solution as a super saturated solute in the quenched state and nano-scale coherent precipitates are formed on ageing up to 500 °C; subsequently, the precipitates coarsen and lose coherency at higher temperatures of ageing, as shown in Fig. 5. Very fine Cu precipitates accompanied by strain contrasts and cluster of copper-rich particles were observed in the specimens aged between

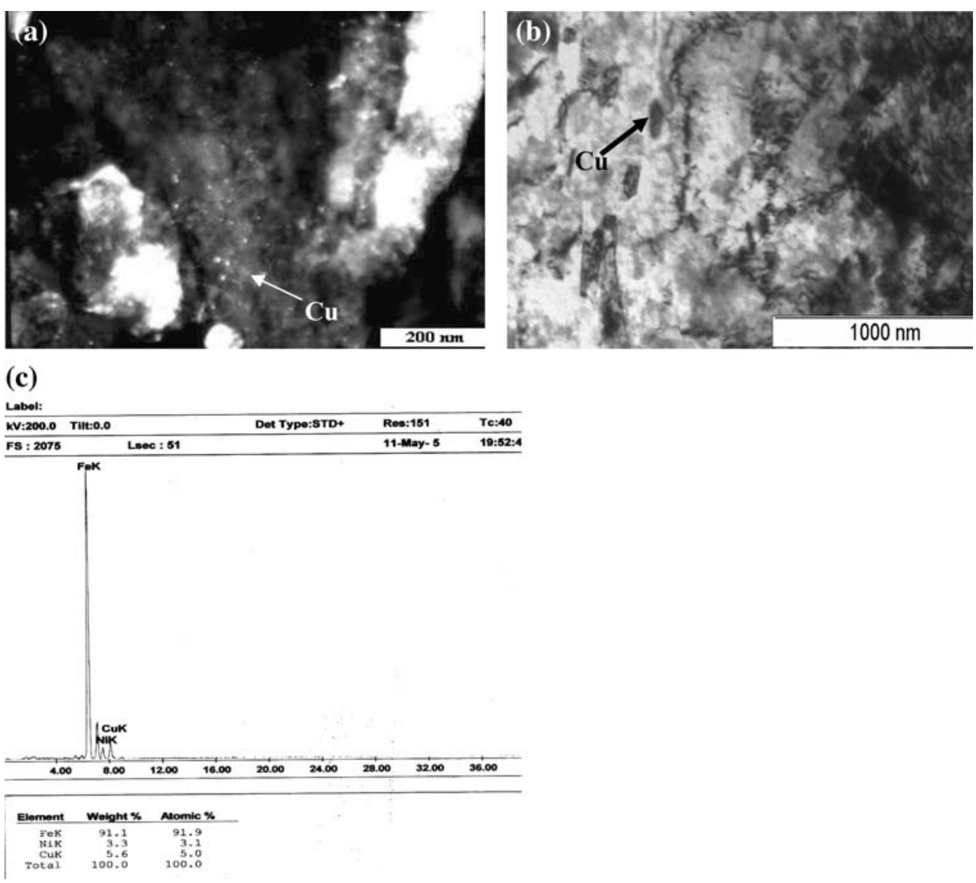

Fig. 5 TEM images showing Cu precipitates at different ageing conditions: **a** aged at 500 °C; **b** aged at 700°C; **c** EDX analysis of the (**b**)

400–500 °C (Fig. 5a). Size of these Cu precipitates was estimated to be less than 20 nm. Strain field surround such precipitates, and the resultant fringes in TEM micrograph indicate coherency with the matrix. Energy dispersive X-ray microanalysis (EDX) of precipitate regions showed an increase in Cu concentration. On ageing above 500 °C, Cu precipitates lost coherency as they grew, and spherical copper particles were transformed to larger rod shaped fcc Cu precipitates at higher ageing temperature. The above observations have been reported by the author and other researchers also [18–23]. Hornbogen and Glenn [22] have mentioned that bcc Cu-rich clusters readily transform to fcc Cu at a small size, and the strain contrast of coherent precipitates is difficult to observe. The tempering of martensite laths, recovery of acicular ferrite to polygonal ferrite and coarsening of Cu precipitates occurred at ageing temperatures above 600 °C. The Ac_1 transformation temperature of this steel is around 675 °C. A small amount of freshly reverted austenite is formed locally in certain regions on ageing above 675 °C that is not as stable as the original retained austenite due to lesser amount of austenite stabilizers in such local regions. This newly formed reverted austenite readily transforms to small martensite islands during cooling, after the ageing treatment. Granular bainite, recovered martensitic laths and acicular ferrite, coarsened Cu precipitates and newly formed martensitic islands were observed in the TEM from samples aged above 700 °C.

Deformation behaviour of HSLA-100 steel on ageing

The variation in yield strength (YS) and ultimate tensile strength (UTS) with change in ageing temperature is shown in Fig. 6. During the first stage of ageing, YS and UTS increased with rise in ageing temperature, up to 500 °C. Beyond this, YS and UTS gradually decreased in the second stage of ageing above 500 °C, with the sample aged at 675 °C showing minimum strength. The third stage of ageing was associated with a rise in strength, and a second peak was observed at 700 °C. The steel softened and strength decreased on ageing above 700 °C due to the combined effect of granular bainite formation, martensite and acicular ferrite recovery, and coarsening of Cu precipitates simultaneously.

As the plastic flow of the material is restricted in the initial stage of ageing, strength of the material increased. A reversed trend was observed with respect to percentage reduction in area (%RA) and percentage elongation (%EL), as shown in Fig. 7. There is a decreasing trend of %RA in the samples on ageing up to 500 °C, and again there is an increasing trend on ageing beyond 500 °C. The specimen aged at 500 °C showed the minimum elongation (5.9%) and maximum elongation (19.84%) was noted against the specimen aged at 675 °C. Figure 8a and b are SEM fractographs of the failed tensile specimens aged at 500 and 675 °C, respectively. Shallow and small microvoids with quasi-cleavage are visible in the fractograph of specimen aged at 500 °C (Fig. 8a), whereas bigger microvoids are visible in the fractograph of specimen aged at 675 °C (Fig. 8b).

Tensile properties of steels are generally dependent on the carbon content of the system. As the amount of C in this system is very low (0.04%), precipitation of very fine coherent Cu particles during initial stage of ageing cause an increase in strength. Strength is decreased during ageing the steel above 500 °C due to coarsening of Cu precipitates. The increase in UTS and YS is due to resistance of the material to plastic flow, as the mobility of dislocations is constrained and higher stress is required for continued deformation. Coherent Cu precipitates and the strain fields associated with them restrict the movement of dislocations. Plastic flow of the material increases and strength

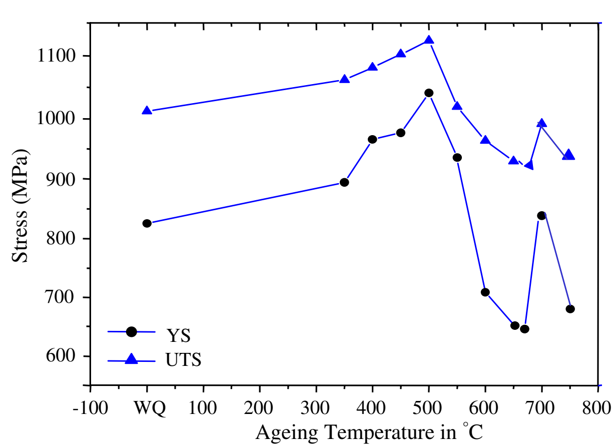

Fig. 6 YS and UTS as a function of ageing temperatures

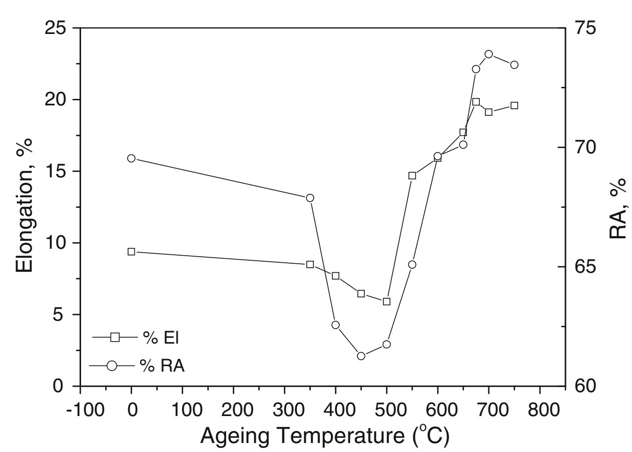

Fig. 7 %RA and %EL as a function of ageing temperatures

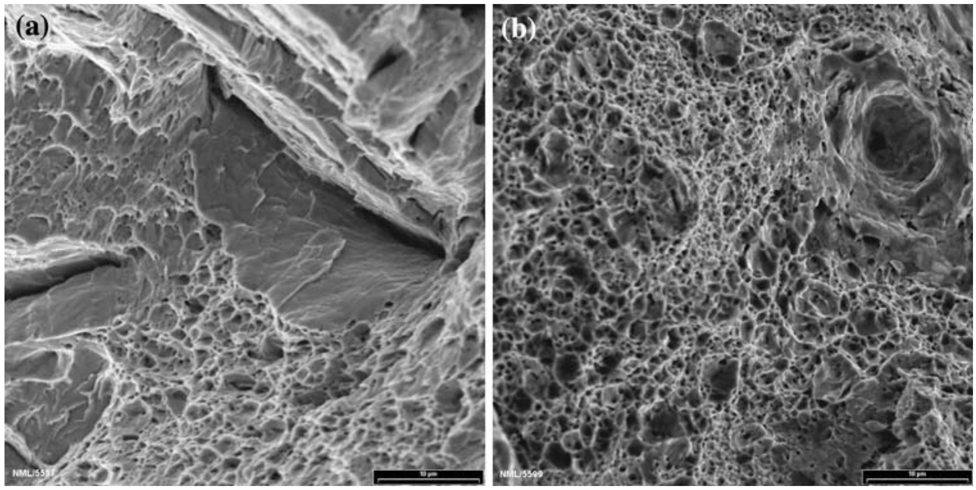

Fig. 8 SEM fractographs of tensile tested specimens: **a** shallow small microvoids and quasicleavage are prominent in specimen aged at 500 °C and **b** only microvoids are prominent in specimen age at 675 °C

decreases in the second stage of ageing above 500 °C. Cu precipitates are coarsened and transformed to incoherent rod shaped precipitates in this stage of ageing, as has been reported by the author elsewhere [23].

In the third stage of ageing (above 675 °C), strength is again increased due to formation of new martensite islands from reverted austenite. Reverted austenite and new martensite islands may have formed below 675 °C, but it was not observed in the TEM. The increase in strength above 675 °C is the net effect of all microstructural changes, i.e. the effect of formation of new martensite islands overbalancing the negative influence of matrix softening and coarsening of Cu precipitates. Reverse trend in %RA and %El can be co-related to the decrease and increase in the plastic hardening of the material in the different stages of ageing. Strength of the steel is decreased above 700 °C due to softening of the matrix and that has also been reported by other researchers [3].

Ductile fracture involves nucleation, growth and coalescence of microscopic voids that initiate at second phase particles and inclusions, and cleavage fracture results from separation along specific crystallographic planes. Restrictions in plastic flow cause cleavage. Fractographic study of the tensile fracture revealed that there are quasi-cleavage features along with small shallow dimples on fracture surfaces of specimens aged between 400 and 500 °C. Ductile fracture features are predominant in specimens aged above 500 °C. The carbides, nitrides and stable second phase particles, which are responsible for initiation of microvoids during ductile fracture, remain unchanged throughout the ageing process [24]. However, precipitation of nanosize coherent Cu particles occurred, which gradually coarsened and lost coherency on ageing at higher temperatures. While it is true that in many cases incoherent precipitates may lead to low fracture toughness and ductility, such observation is usually made for precipitates in which the interface has lost almost all load bearing capacity, and when the precipitates are of some what larger length scale or where they appear in profusion, as for bigger inclusions in steel. In the present case, the small size of precipitates (<1 μm) even for overaged conditions and their low number density lead to variants of particle-strengthening behaviour. Thus, when precipitates are coherent and as their number density increases, strength increases, while on coarsening of precipitates, with loss of coherency, strength decreases. Fracture toughness, being usually inversely related to strength, changes accordingly as observed. Thus, quasi-cleavage features appeared at 500 °C when the material showed higher strength and the Cu precipitates were in a coherent state. The material also showed low ductility at 500 °C. On the other hand, incoherent copper precipitates facilitated microvoid initiation and growth, and hence, increased ductility. The presence of larger microvoids and disappearance of quasi-cleavages in later stages of ageing support this argument.

Conclusions

The plastic flow of the material is restricted in the initial stage of ageing due to precipitation of very fine coherent Cu-rich particles leading to an increase in brittleness and strength at the expense of the ductility of the material. At a later stage of ageing (above 500 °C), Cu precipitates coarsened and lose their coherency which in turn increases ductility and flow of the material.

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