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Effective use of residual elements in steels produced from scrap

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Abstract

As the need for steel increases the benefits of using scrap for high quality steel products increases. In the production of steel by electric arc furnace (EAF) scrap is used as the main raw material. One of the problems of using scrap as a raw material for steel production is the accumulation of elements in the scrap that can not be removed in the steel making process. The restrictions on the levels of residual elements in low alloyed steel are often set from the point of view that “cleaner” steels are better. Although several of the elements considered harmful when recycling can be effective as alloying elements. There is a special concern for the elements that are nobler than iron, for example copper and nickel.

In this work the effects of copper and nickel on low alloyed (LA) steel intended for thin sheet applications are studied. The choice of material is based on a previous study done at KIMAB and a literature study in combination with company visits, which are reported in chapter 2. Batch annealing and continuous annealing treatments are used to compare the behaviour of the obtained sheets and the differences in the evolved microstructure. The mechanical properties such as yield strength and Lüders strain are measured. The microstructure is also investigated and correlated with mechanical properties. The microstructure is examined with conventional optical microscopy and electron microscopy, enabling texture measurements with Electron Back Scatter Diffraction (EBSD) and imaging of the fracture surface.

Results show that there is no major effect of copper on the tensile strength in neither batch- nor continuously annealed sheets. Copper in combination with the highest nickel content in this study show an increase in tensile strength in the batch annealed materials, but no significant effect in the continuously annealed material. The so-called gamma-fibre texture is more pronounced in the batch annealed material, and has a higher intensity with increasing copper and nickel content in both annealing types. The continuously annealed material shows a finer microstructure than the batch annealed, that has a more pronounced elongation of grains in the rolling direction.

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1. Literature study

1.1 Introduction.

This literature study aims to broadly summarise the experience and knowledge regarding residual elements in low alloyed steels and their effects on the processing and properties of the steel. The study is to be used as a basis for material choice and experimental trials regarding the ‘self alloying’ of steel.

Self-alloying refers to the possibility of using the existing elements in scrap as alloy elements in the finished steel product.

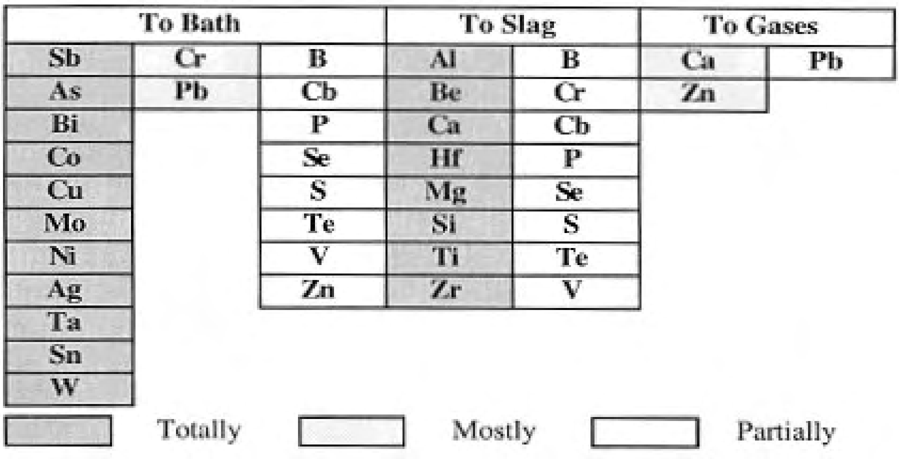
In any given type of steel there is always present a certain amount of elements besides iron and carbon, some of these have been added intentionally as alloy elements and others, so called residual elements, are leftovers from the processing or use of the material. These elements come from the scrap or ore used to produce the steel. A difficulty is trying to quantify the effect of any given element, especially when in combination with other elements, as has been reported by P. Hart [1]. Hart points out that the trend towards lower sulphur and lower oxygen level steel has been greatly beneficial but not exclusively so. This highlights the complexity of the dynamics behind the interactions of different elements and their effects on the properties of the steel in which they are present.

As copper has been found to cause problems both in hot-rolling and in welding as well as in cold working, studies [2, 3, 4, 5, 6, 7, 8, 9] have been made concentrating on the effects of copper in low alloyed steels. The effect of copper is to raise the yield strength, lower elongation and cause a refinement of the microstructure. Nickel has a similar effect on mechanical properties, but significantly lower compared to copper. Alloying with nickel in steels that contain copper can lessen the problem of surface cracking during hot rolling. Tin also shares the effects on mechanical properties with copper suggesting similar dynamics. Tin however has a larger impact on mechanical properties, which is attributed to its greater difference in atomic radius in relation to iron as compared to copper-iron. These studies outline both problems and benefits with copper and other residual elements.

With an increasing demand for steel world-wide, scrap metal has become a critical resource. In Electric Arc Furnace (EAF) melting of steel, 100% of scrap is generally used. Ovako steel in Sweden produces entirely from scrap with the EAF process and uses no virgin material other than alloy additives. In 2004 the total production of steel coming from EAF was 31% [10] of 6 million tonnes of total crude steel produced in Sweden. The need for steel is greater than the available amount of scrap being produced each year. This means that virgin material produced from ore is still needed, and virgin material is used to dilute the tramp elements to acceptable levels. With heightened demands on efficiency and ecological concerns the use of scrap material will increase. But the increased use of scrap metal in the steel making process results in rising amounts of some residual elements in the recycled materials. Among the materials that accumulate in recycled steel material, copper, tin and nickel are notable since they have less affinity for oxygen than iron (they are more “noble”). This means that they can not easily be reduced from the steel melt during normal processing. The understanding of the effects and dynamics of these elements in greater detail becomes increasingly important as the use of scrap in steelmaking will increase.

1.2 Residual elements in steel.

Elements are considered as impurities or alloy elements depending on where in the process the element is added or if it is added intentionally. The elements that have lower oxygen affinity than iron, such as Cu and Ni, are not reduced during the production process and mainly remain in the final alloy (**Figure 1**). To date there exists no economically viable way of removing these elements. However, there are studies concerning the possibility of evaporating Cu from steel by adding Si and/or C to the steel to manipulate the evaporation rate of copper [11]. The effects of residual elements are experienced in both production of the steel, and in the finished product during use. Most elements exist in both solid solution and in some stochiometric combination with another element such as manganese and sulphur forming MnS inclusions. In the case of high alloyed steels, such as stainless steels, the effect of residual elements is less pronounced than in low alloyed steels. Copper, like nickel, acts as an austenite stabiliser and has a retarding effect on corrosion and could at least partially replace nickel in stainless steel grades.



To Bath			To Slag		To Gases	
Sb	Cr	B	Al	B	Ca	Pb
As	Pb	Cb	Be	Cr	Zn	
Bi		P	Ca	Cb		
Co		Se	Hf	P		
Cu		S	Mg	Se		
Mo		Te	Si	S		
Ni		V	Ti	Te		
Ag		Zn	Zr	V		
Ta						
Sn						
W						

▨ Totally ▒ Mostly ☐ Partially

Figure 1 *Some elements present in commercial grade steel and general behaviour under meltdown conditions in EAF operation.*

1.3 Hot-working

The effect of residual elements in steel, regarding the hot-working properties, has mostly been studied in relation to copper. This is because copper has been found to be the critical element in causing problems during hot working and has therefore been more extensively researched. The research done on other residual elements in regards to hot-working properties has mainly related the given elements behaviour with the presence of various copper contents in the steel, as seen in [2,12], or on varying levels of other residuals as in [6], in combination with copper. The discussion regarding copper as either an alloy- or residual element, as seen in [13], is dependent partly on the end product and partly on the steel production route. In several cases[6,8,9] it has been noted that copper has beneficial effects on the end products in regard to a finer grain structure and increased yield strength.

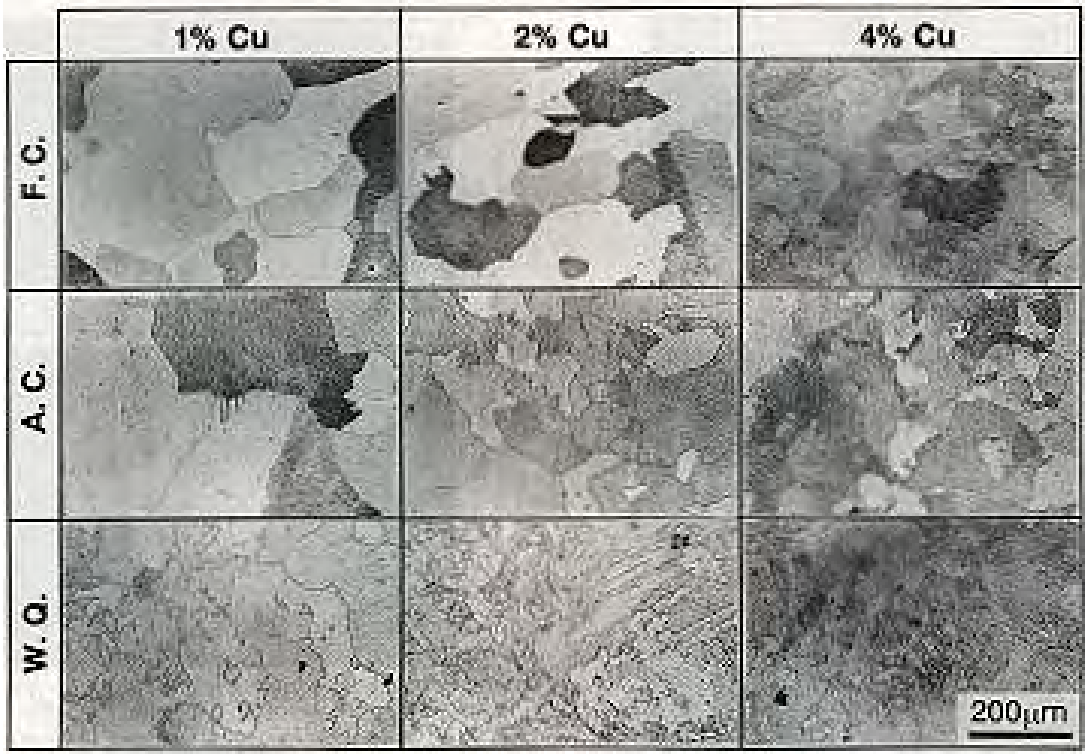


Figure 2 *: Differences in microstructure depending on cooling rate and copper content* *[14]*.

When the material is hot rolled the grain size of the finished material is largely dependent on the final temperature during rolling. Higher temperature results in coarser grains. Lowering the rolling temperature will result in a finer grain structure. Also a coarsening of precipitates is more pronounced at lower temperatures. The influence of cooling rates and copper content can be seen in **Figure 2**, where the steels shown were solution treated at 1523 K for one hour and were either furnace cooled (F.C.), air cooled (A.C.) or water quenched (W.Q.) [14]. The development of the microstructure will have a direct effect on the mechanical properties of the steel and its behaviour during cold working.

1.3.1 Copper – hot shortness

The increase of Cu in steel scrap can be mainly attributed to increased use of electrical and electronic devices in home appliances and cars. Using scrap from such products leads to steel that often must be diluted with virgin material to reduce the amount of Cu to acceptable levels. The main concern is that copper leads to surface cracking in hot-rolled steels due to the presence of a copper-rich liquid at the hot rolling temperatures of the steel. Due to preferential oxidation of Fe at the surface, Cu is enriched in the grain boundaries creating a liquid phase. The enriched phase becomes liquid over 1090 °C. The liquid can then infiltrate the austenite grain boundaries (**Figure 3**)causing severe cracking (**Figure 4**). However it has been observed that there exist processing windows to alleviate this problem[2].

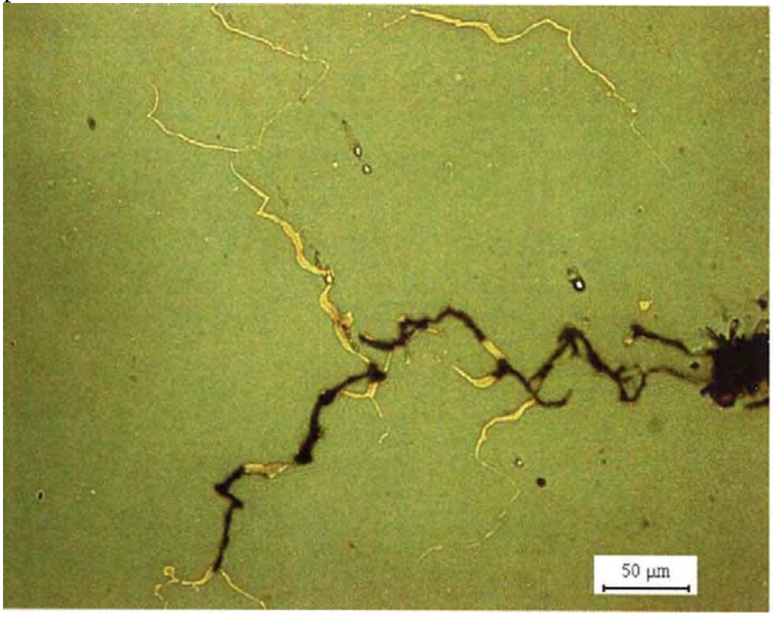


Figure 3 *Cracks following austenite grain boundaries that has been infiltrated by liquid copper in a steel alloyed with 2.5 w% Cu* [9].

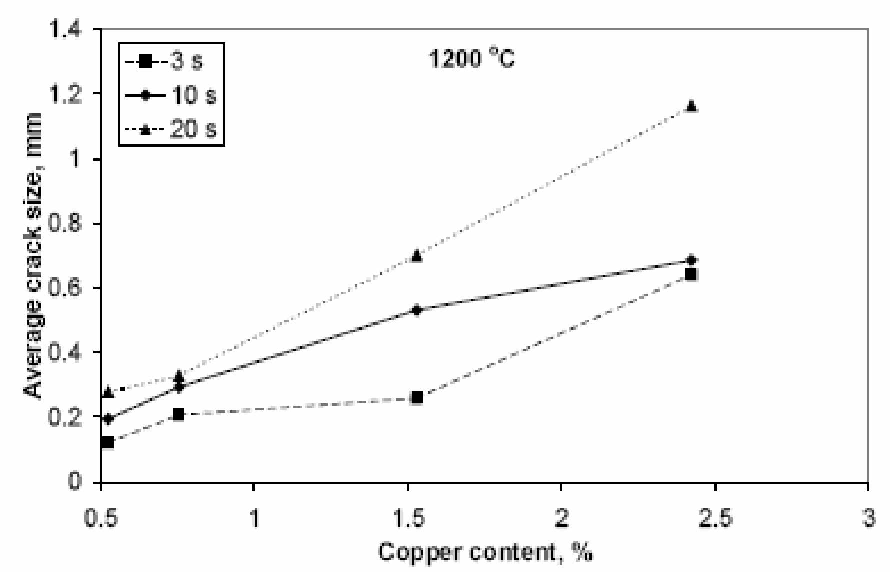


Figure 4 *: The effect of copper content on the average crack size in the hot ductility tests in [9] at varying oxidation times in air.*

The solubility of copper in iron increases significantly with temperature, so increasing the dispersion of copper into the grains. At 1193° C the solubility of copper in austenite is reported to be about 10 w% but only about 0.1 w% at 521° C [7]. However increasing the temperature will also increase the rate of oxidation of the surface. Formation of the copper rich phase occurs mainly in the struggle of two competing forces: copper enrichment in the grain boundaries due to preferential oxidation of iron at the surface and dispersion of copper in to the grains through diffusion [12]. As the temperature increases the dispersion of copper becomes dominant, at longer oxidation times a steady state is reached and the enrichment and dispersion rates of copper along the grain boundaries become constant. It has been found that there is a critical temperature where maximum cracking occurs that is dependent on copper content[3] (**Figure 5**).

However at temperatures above 1200° C up to 1300° C there is a marked reduction in surface cracking at Cu levels up to 0.39 w%[3] to the point where no cracking could be found. These results are consistent with observations of Cu-enrichment in metal/oxide interface of plain carbon steels [12].

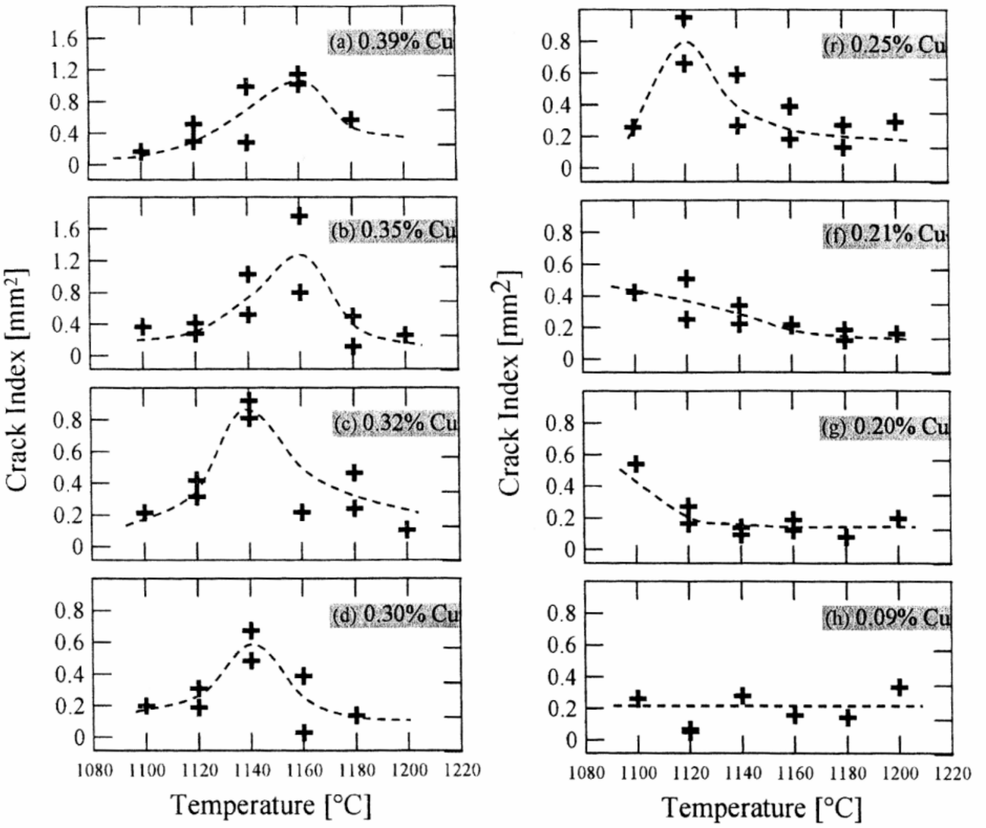


Figure 5 *Crack index measurement for steels containing differing levels of copper illustrating maximum cracking at a critical temperature. All samples were heated and oxidised for 10 min and deformed, cracking was quantified by average crack surface area[3].*

In order to counter the negative effects of mainly copper, new methods for near-net shape production have been studied [9]. By casting and rolling the steel directly from the melt and under a protective atmosphere of argon and carbon dioxide, many of the problems can be avoided. The so-called Direct Strip Casting (DSC), is a method that combines short oxidation times with rapid cooling and direct rolling. This means that there is very little oxidation of the surface compared to conventional hot-rolling. Avoiding oxidation also means that there is less material loss and the enrichment of copper in the surface is avoided. From a ladle the melt is fed on to a water cooled conveyor belt. The sides are contained by water cooled copper blocks that move along with the melt in a block-chain [15]. The Ar/CO_2 protective gas is used as a rake to ensure a uniform distribution of the melt across the casting belt. The rapid cooling from the combination of the conveyor, acting as a continuous casting form, and the Ar/CO_2 gas rake makes it possible to produce steel grades previously considered to be unfeasible. This method is currently only developed for flat products and has been tested for a width of 300 mm of the cast strip.

A combination of methods allows the use of higher copper contents than is currently allowed in flat, hot-rolled steel products. Using lower rolling temperatures, shorter oxidation times and/or a protective atmosphere to reduce the oxidation rate, together with alloy elements in the steel, can decrease the susceptibility to hot-shortness.

1.3.2 Other elements combined with copper.

Tin in combination with Copper is highly detrimental for the susceptibility to hot-shortness in the steel. Tin lowers the melting point of the copper phase and decreases the solubility of copper in iron causing further enrichment[5]. However in low copper steels tin by itself does not seem to affect surface hot shortness and no significant enrichment seems to occur [12]. Tin affects the workability at high temperatures but does not show as much of an effect as in combination with copper.

Adding as much as 0.5 w% Ni (levels corresponding to roughly the same as the copper content) also decreases the susceptibility to hot shortness. In combination with an addition of 0.4 w% Si an addition of only 0.26 w% Ni is needed to reach the same reductions as with only 0.5 w% Ni by itself. Ni also increases the solubility of Cu in iron lessening the susceptibility to hot-shortness [5] further. Ni affects the solubility of Cu in austenite and raises the melting point of the copper enriched phase.

The exact effect of different elements on the hot-shortness dynamics is hard to quantify. Researches into specific elements have been made[4] regarding the interaction of the liquid Cu phase and the effect of the element. It has been found that in 0.1 w% C steels, boron in amounts about 7 ppm seems to reduce the susceptibility to hot-shortness by altering the ability of the Cu phase to wet the surface of the austenite grain boundaries[4]. One theory is that boron segregates to the grain boundaries and influences the surface boundary energy in the steel.

An addition of Si to the steel has been researched in order to suppress hot-shortness mainly by affecting the steel scale. The formation of a molten phase of iron and silicon oxide in the scale can work favourably to occlude liquid copper into the scale, not allowing it to penetrate the steel [5] (**Figure 6**). Si in combination with P and C in solid solution also reduce the susceptibility to hot-shortness through restraint of the level of penetration of the Cu-rich phase into the grain boundaries. The influence of silicon on the evaporation rate of copper has also been studied [11]. The affinity of silicon to iron in the melt affects the activity of copper in the bath. Evaporation of copper from the bath was found to be promoted by the decarbonization of the melt that caused turbulence in the gas/metal interface.

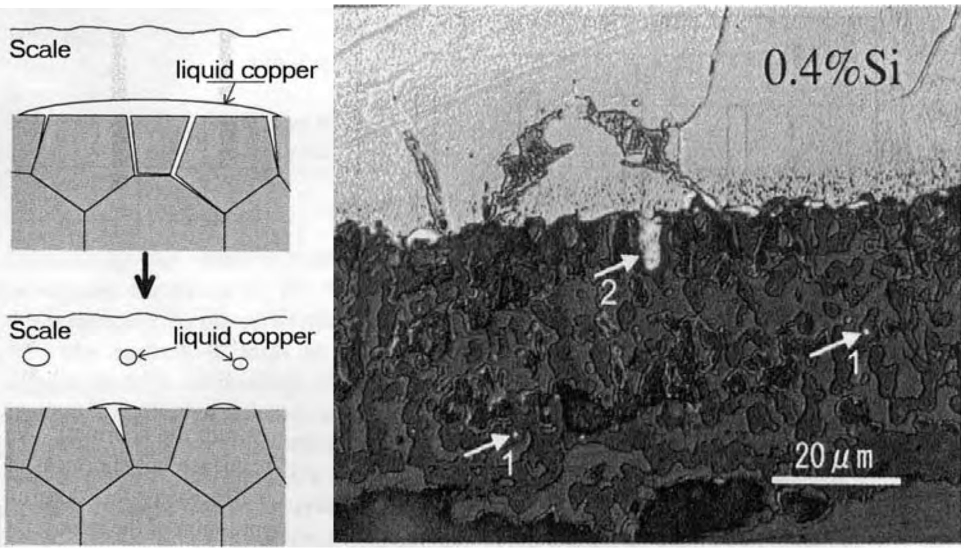


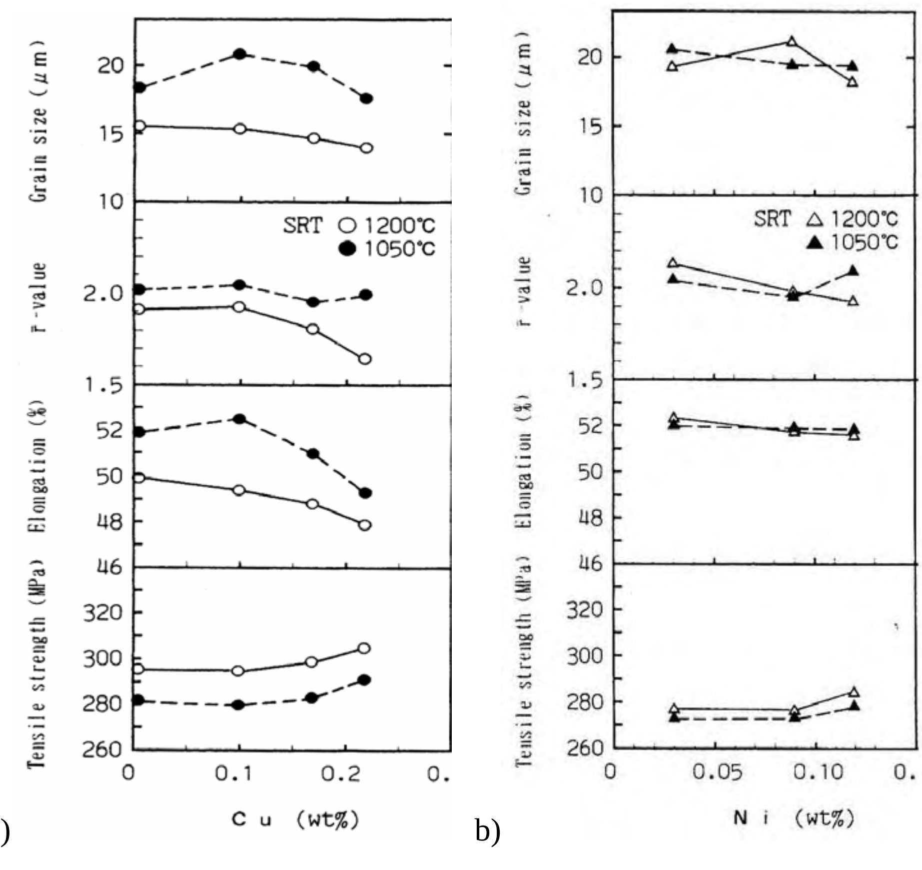
Figure 6[2] *Left: principle of Cu-enriched phase occlusion . Right: actual micrograph of occluded copper indicated by arrows. 1 shows fully occluded Cu-enriched phase, 2 indicates partially occluded Cu-enriched phase.*

1.4 Cold Working: Drawing, rolling and machining.

Residual elements in solid solution and in fine precipitates mostly contribute to hardening of the steel. The harder the material the more likely it is to exhibit brittle behaviour when deformed. For materials intended for deep drawing or good formability, such as press sheets for car bodies etc, a low yield strength and long elongation are desirable.
The drawability of the material is quantified in part by the so-called r-value. The r-value is the quota between the contraction of the material in the width direction during strain and the contraction of the material in the thickness direction. The microstructure and texture of the material controls the elongation, tensile strength and the r-value. The microstructure in turn is a result of the hot and cold working, alloying elements and heat treatments of the material.

1.4.1 Copper and Nickel

Copper specifically lowers the elongation of the material and increases the yield strength (**Figure 7**) while contributing to a refinement of the microstructure [6]. Copper by itself though, up to levels of 0.5 w%, does not seem to have any significant influence on cold formability in regard to surface cracking; the presence of inclusions, such as MnS, seems to be the main factor contributing to the formation of surface cracks [8].



a) b)

Figure 7 *a) Mechanical properties as a function of copper content in titanium-bearing low-carbon steel sheets[6]. b) Mechanical properties as a function of nickel content in titanium-bearing low-carbon steel sheets[6]*

The refinement of the microstructure by precipitation of small copper particles (**Figure 8**) contributes to the hardening and strength by resisting the movement of dislocations. Yield/tensile strength and elongation have an approximately inverse relation to each other. In a short study made on the annealing effects on copper in thin sheet products, it was found that a moderate amount (0,25 w%) of copper in the material seemed to eliminate the Lüders strain.

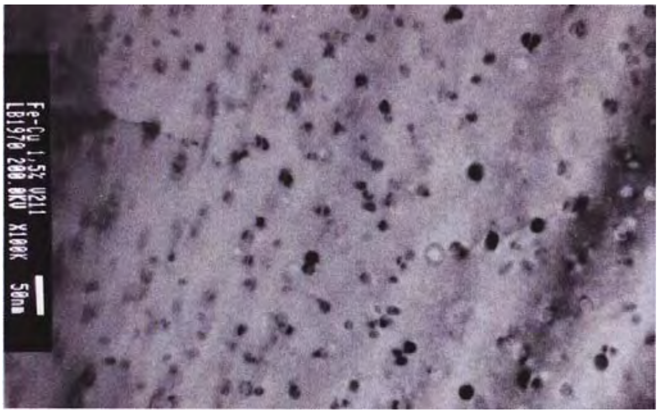


Figure 8 *TEM picture of a 1.53 w% Cu alloyed steel after simulated hot rolling and coiling at 873 K*[9]*. The dark spots are precipitates of ε-Cu 8-16 nm in diameter.*

This was only seen in material that had been batch annealed (**Figure 9**a). The effect was not seen in the continuously annealed specimen with the same copper content (**Figure 9**b). The elimination of the Lüders strain is of interest for press sheets and steels that are intended for deep drawing and cold forming.

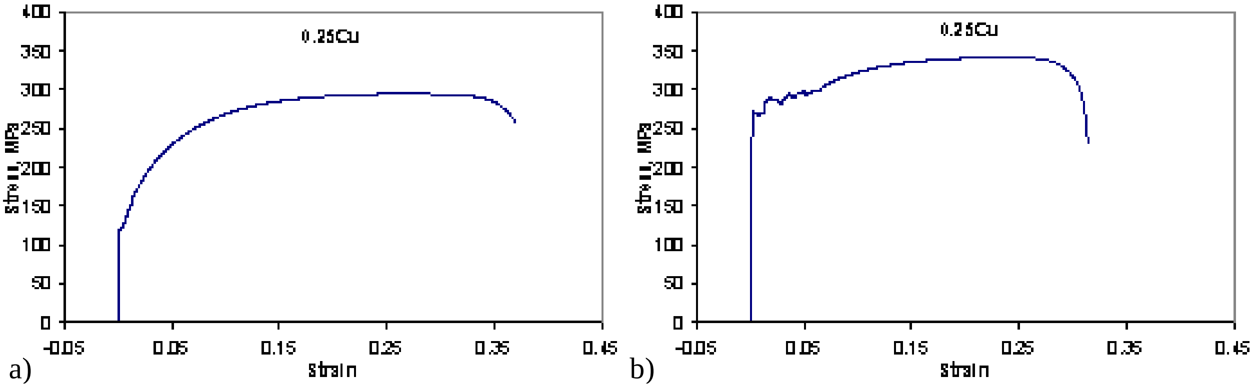


Figure 9 *a) Stress-strain curve for a batch-annealed steel, containing 0.25 w% Cu, showing no Lüders strain. b) Stress-strain curve of same material as in a) but continuously annealed and showing a clear Lüders strain.*

Nickel, like copper, is another element that remains in the steel during the melting process, being nobler than iron under the conditions experienced during steel making. The influence of nickel is of about the same magnitude as copper in respect to tensile strength increase per 0.1 w%. It should be noted that these effects were observed in low carbon steel sheets containing titanium [6]. Only minor influences on elongation and r-value were observed at varying nickel content and at different heating temperatures, though the small changes are more probably related to the low carbon content in combination with the presence of titanium. In total the influence of nickel contents up to 0.12 w% have little effect on the mechanical properties of low-carbon titanium containing steel sheet.

1.4.2 Other elements.

The effect of sulphur on machining, as studied in [16], is observed to be mainly by the forming of sulphides with other elements such as Mn. MnS can act as a nucleation site for microcracks, which promote shear instability and so improve machinability.

Tin increases the tensile strength of steel and lowers elongation and r-value [6] (**Figure 10**). A refinement of the grain size was observed with increasing tin content. The change in tensile strength was reported to be approximately 45 MPa /0.1 w% tin. Tin has a larger difference in atomic radius versus iron compared to copper, nickel and chromium. This causes tin to have a larger effect in both solid solution and as a precipitate.

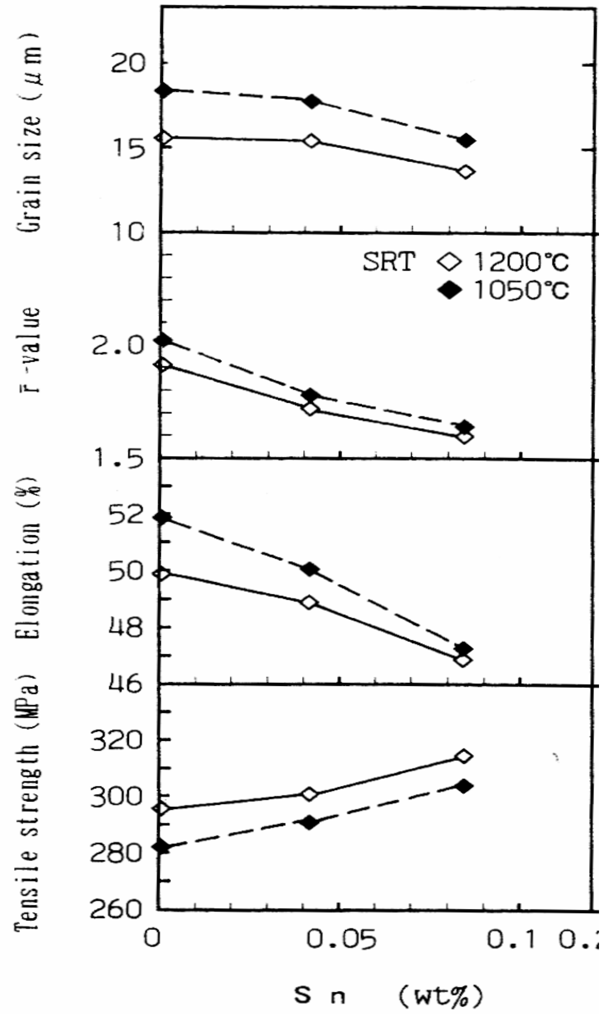


Figure 10 *Influence of tin on mechanical properties in titanium-bearing low-carbon steel sheets*[6].

As an alloy element in steel titanium is used both as a dispersion hardener through the forming of titanium carbide particles, and as a sulphur binder to form hard sulphides and thereby improve cold working abilities. Titanium can also form precipitates that can act as nucleation sites for grains, contributing to a finer grain size.

1.5 Welding.

The effects of residual elements on the welding properties of low alloyed steels are widely varied. Most residual elements contribute to the hardenability in the HAZ of a welded material. This can be a disadvantage in the weld as it reduces the toughness.
The weldability of a material can be judged by its properties in relation to[7]:

1. Hot cracks during solidification.
2. Lack of fusion, blowholes and spatter.
3. Cold cracking, hydrogen embrittlement.
4. Lamellar tearing.
5. Weldment strength.
6. HAZ toughness.
7. Weld metal toughness.
8. Loss of toughness during stress relief treatment.
9. Reheat cracking
10. Weldment fatigue strength.
11. Preferred corrosion for example due to formation of chromium carbides in the grain boundaries, causing chrome or other carbide formers to be leached from the grain boundaries.

Hot cracking, also called solidification cracking, occurs in the boundary between the weld metal and the base material. When the melted material cools down, grains grow from the material in towards the thermal centre of the weld. The composition of the material in this zone is dependent on both the base- and the weld material, as the part of the base metal is melted along with the weld metal. As the melt solidifies, segregation and solidification shrinking occurs. This can cause low melting phases to be present as the material solidifies and it is this that can cause tearing and cracking in the weld.
In different models developed to calculate the so called ‘carbon equivalent’, used to judge the susceptibility to cold cracking among other things, there is a great variation in coefficients for a specific element in the different formulae[5]. This shows that it is difficult to absolutely estimate the effect of any given element.

1.5.1 Copper

The effect of copper on the weldability of steel is quantified through its contribution towards the material tendency for hot and cold cracking. Copper has a tendency to cause hot cracking. Usually the hot cracks associated with copper are small but can serve as initiation points for future cracking and stress corrosion cracking. Hot cracking occurs when a critical strain is applied at elevated temperatures. The problem of hot cracking can generally be avoided as long as the applied strain at high temperature remains below the critical value for the given material.

The solubility of copper in austenite is higher than in ferrite; this causes the matrix to become supersaturated with copper as austenite transforms to ferrite during sufficiently high cooling rates. If the cooling rate is slow the copper will form precipitates in the bulk matrix of the steel. The hardening of the material by dispersed fine grains of precipitated copper is one of the reasons for alloying steels with copper. Hardenability of the HAZ is inversely related to the weldability: higher hardenability leads to lower weldability. The finely dispersed precipitates can also contribute to a refinement of the grain size in the HAZ. Refinement of the microstructure contributes to increasing the toughness in the weld. It is not only the copper content of the bulk material that has to be considered though, but also the copper content of the welding rod that together with the bulk material make up the total copper content in the weld.

Copper by itself seems to have a limited effect on the weldability of C-Mn steels, but in combination with Sn, Sb and As copper has a strongly negative effect. This can partly be attributed to the lowered solubility of copper in austenite when the austenite contains Sn, Sb and As. It is suspected, but not extensively researched, that the problem is the same as with hot shortness, i.e. presence of a liquid copper phase, is responsible for the poor welding qualities in this case[5]. It has been observed according to [17] that the addition of copper by itself, from scrap in steels produced through EAF route, does not seem to have any debilitating effects on either the base material or on the welded metal in the HAZ concerning levels up to 1 w% Cu [7].

Nickel in combination with manganese and chromium act to restrict occurrence of a copper liquid by increasing the melting point of the copper phase and increasing solubility of copper in the iron matrix.

1.5.2 Sulphur

The presence of sulphur is known to have adverse effects on cracking in the heat affected zone (HAZ) of welded materials. However, when the levels of sulphur are too low, problems with hydrogen cracking in C-Mn steels, or improper penetration in specific Cr-Mo-V steels have been reported [18]. The hardenability of the HAZ is increased with lower oxygen/lower sulphur steels and this is considered to be the cause of the increased risk of hydrogen cracking. Also it can be noted that sulphur act as a surface reactant and decreases the rate at which nitrogen is picked up[19]. Therefore higher nitrogen levels can be found in low sulphur steels. Sulphur has an affinity towards manganese but can form FeS in low manganese steels. FeS melts at 988 ^{o}C and act much like copper in that it forms a film in the grain boundaries. This behaviour cause hot-tearing which is seen during solidification of the weld material. The sulphide film causes material separation between the grains as they are cooled and undergo solidification strain.

1.5.3 Other elements.

In welded material tin, together with antimony and arsenic, acts to change the surface energy of the grains, since these elements have a tendency towards grain boundary segregation. Tin like boron can affect the wetting angle of a copper liquid and lessen the penetration in to the grain boundaries. This decreases the susceptibility to hot-tearing and hot-shortness. However, tin also reduces the solubility of copper in austenite and lowers the melting point, so can thereby be severely detrimental in the case of hot tearing.

The decrease in sulphur and oxygen contents in steel has not been only beneficial. Boron is often present as a residual element in steel but may not exist in an active form in steel that was not intentionally alloyed with boron. Often boron was combined with oxygen and nitrogen to form an inactive state. In combination with more aluminium and titanium in modern steels as well as lower nitrogen and oxygen contents boron can exist in an active form. Even levels as low as 3-5 ppm can have a significant effect on the hardenability of a material. In relation to the amount for a given hardening effect, boron has a large effect [1].

1.6 Conclusions

The following trends and features could be found in the studies included in this literature survey:

(1) The levels of copper found in steel, resulting from using scrap as raw material, can today be managed with metallurgical processing techniques.
(2) The current restrictions on levels of residual elements have often been set from a cautious viewpoint. With more research producing a deeper understanding of the dynamics behind the interactions of different elements, it is reasonable to believe that the level of residual elements can be increased in high quality steel products.
(3) Better scrap analysis is needed to make effective use of residual elements in scrap for micro alloying of low alloyed steels.

(4) The benefits of copper in steel makes it viable to use scrap containing higher levels of copper as raw material.
(5) A combination of methods to avoid hot-shortness allows production of steel grades with improved mechanical properties compared to currently produced commercial grade low-alloy steel.
(6) There exists a critical temperature interval, depending on the level of copper, at which surface cracking has a maximum. If this temperature is avoided, hot-shortness can be significantly lessened.

(7) Nickel levels of about half the amount of copper in w% can offset the problem of hot shortness caused by copper.
(8) The combination of tin and copper is a problem both in hot rolling and welding, causing similar problems. The presence of a copper phase that has a low melting point compared to the base material can cause surface cracking during hot rolling and hot tearing during welding as the material solidifies.
(9) Copper levels up to 0.85 w% do not cause problems with weld toughness and can be welded safely with some additional weld heat treatment.

2. Company visits

2.1 Introduction

Two companies, which are industrial partners in this project, were visited to gather feedback, the Ovako steel mill in Hofors and Sandvik Steel in Sandviken. Before the visits a questionnaire was sent to the companies to have a basis for discussion during the visits. The questions were regarding which residual elements were being retained in the production line at present time and which ones would be interesting to maintain in the products if there was a hypothetical increase in cost of certain alloy elements. At Ovako the discussions focused on production and raw material, at Sandvik the problems/benefits from residual elements in finished products were discussed. The interests of the specific industries were also discussed to determine what technical trials may be interesting to perform.

2.2 Ovako

Ovako produces steel and steel products through a scrap based electric arc furnace route. Products delivered by Ovako includes bars, rods, wire and rings (**Figure 11**) for bearings and for uses in the automotive industry for example. The Ovako steel mill in Hofors, Sweden both produces semi-finished products as well as rings, tubes and bars. The production of tubes and rings in Hofors 2005 was about 125 kton.

***Figure 11** Bars, tubes and rings from Ovako.*

2.2.1 Steel mill production.

The steel mill based in Hofors produces steel entirely from scrap. The raw materials are purchased from varying scrap sources and about 20% are imported, mostly from Russia and the Baltic countries. 27% of the material used is internal scrap from production including ingot ends, machining chips and other scrap from product manufacturing. Classification and composition analysis is mainly done by the supplier of the scrap and to a lesser degree at Ovako. The scrap is classified according to composition and size. The scrap is melted in an electric arc furnace (EAF) treated in a ladle furnace and cast into ingots by uphill teeming (**Figure 12**). Elements that are wanted in the steel are added as alloy materials in the ladle furnace, where the steel is given its chemical composition. The ingots are then put into a heating pit to reduce segregation in the material before rolling and oxygen scarfing. Rolling and reduction of the material help to control inclusions in the steel. The resulting billets are then used in production of the finished products.

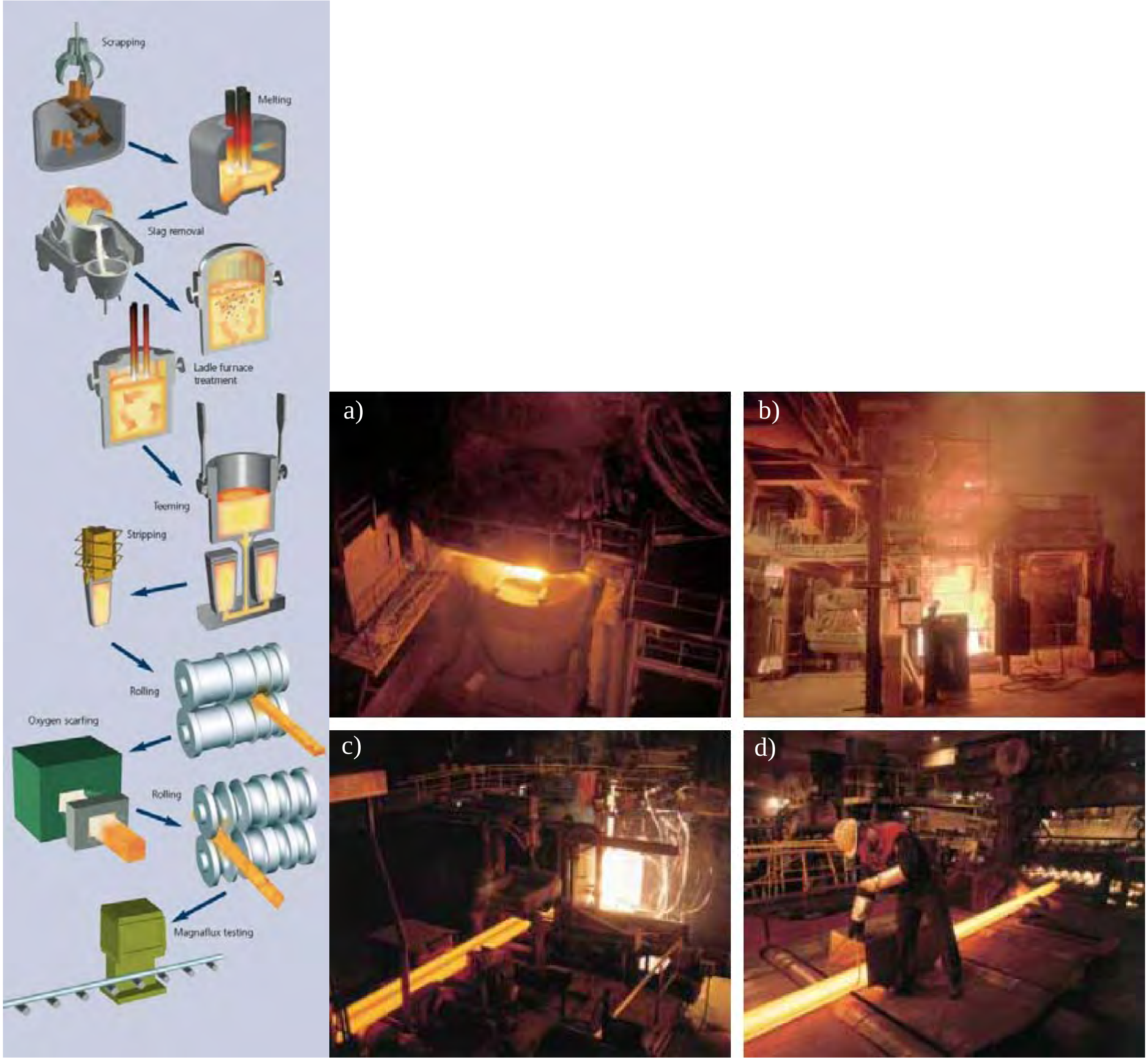


Figure 12 *Production process in the Ovako steel mill in Hofors.a) Steel melting, b) Ladle treatment, c)Billet rolling, d) Oxygen scarfing.*

2.2.2 Product demands.

Producing steels with low oxygen and sulphur content and controlling the composition of the steel in each production step, allows Ovako to have good control over the finished product. The chemical compositions of the steels are specified by the product and the need for further machining, through hardening and final product application etc. The levels of copper, nickel, and other alloy and residual elements, are controlled through choosing the right scrap for raw material and through direct addition of elements when needed. The hardening behaviour of the material is in part dependent on the levels of alloy elements and also on residual elements. Steels intended for through hardening, used extensively in roller- and ball bearings, cannot have too low levels of alloy and residual elements or else the material will not harden properly. Part of the maximum and minimum limits for residual- and alloy elements are set to get as small differences in hardenability as possible.

2.2.3 Response from Ovako.

Scrap with more residual elements will have a lower price than high-grade ‘clean’ materials. Since the production in the Hofors steel mill is based entirely on scrap, the benefits of being able to use lower priced scrap are greater than in other steel production that uses mostly ore based raw material. A possible future development for Ovako is to produce through near net-shape casting, as it would reduce machining and handling costs, bloom casting would also be a possibility.

2.3 Sandvik

The Sandvik business group produces cemented-carbide and high speed steels for metal working, rock excavation machinery and stainless and high-alloyed steels. The Sandvik steel mill in Sandviken produces stainless steels and products such as seamless tubes and different grades and types of steel wire (**Figure 13**).

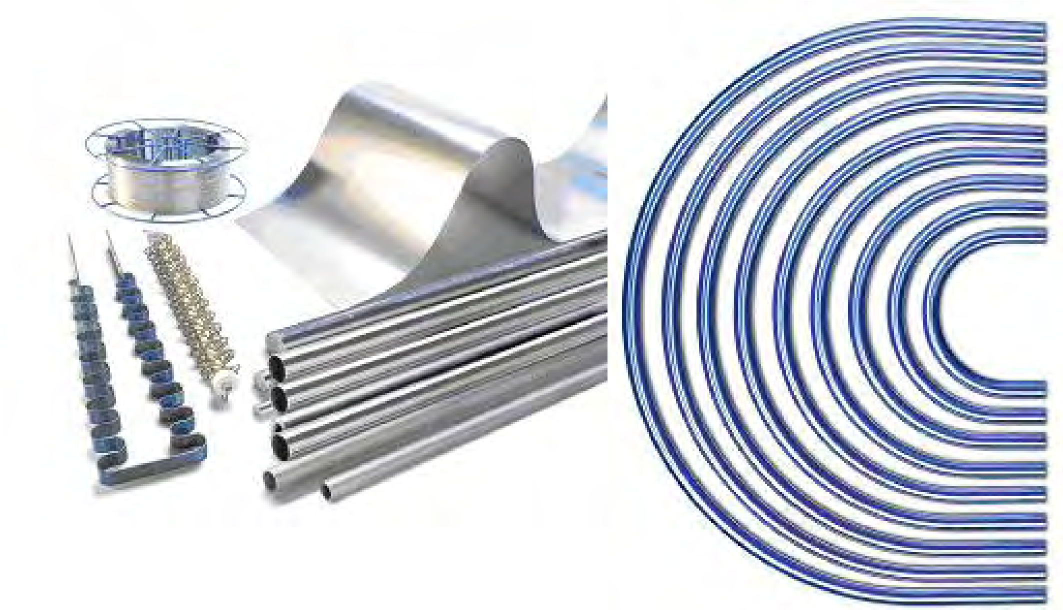


Figure 13 *Products made by Sandvik Steel. To the left is speciality steels in wire, sheet and other forms, on the right is seamless stainless steel tubes.*

2.3.1 Response from Sandvik.

Sandvik have seen a trend of rising copper levels in scrap used for raw material in their production. However there are fewer problems with copper, and other rest elements, in high-alloyed products such as stainless steels. Although there may not be any technical problems in production or in the end product with higher levels of rest elements, levels of residual elements are certified in different steels grades. The established standards would require too much work to change without significant gains.

3. Effects of Copper and Nickel in Low-alloyed Sheet Steel.

Introduction.

This study is based on a previous study done at KIMAB concerning copper and tin in direct cast steel produced from scrap[20]. The choice of alloy composition for the materials studied in this work is based on previous studies made, and on reasonable levels of copper and nickel that can be found in scrap used for raw material. These levels have been estimated from the literature study and from discussions with steel companies. Other residual elements are at levels found in low alloy steel grades. The phosphorus, sulphur, silicon and carbon in the material come from the base material used to manufacture the tested alloys.

The tests performed on the alloys were chosen to give good background for assessing the influence of copper and nickel on material properties. This included standard tensile testing, microstructure studies in both optical and electron microscopes, texture analysis and analysis of the final chemical composition.

3.1 Alloy production.

3.1.1 Tested alloys

The alloys were selected to test the effects of copper alone and in combination with nickel on the mechanical properties of low alloyed steels intended for cold forming.
The composition was analysed with Glow Discharge- Optical Emission Spectroscopy (GD-OES), carbon levels in the materials were analysed at SSAB tunnplåt in Borlänge by Leco carbon analysis. The Leco carbon analysis was only done on the materials as cold rolled and on the annealed material (not on the castings or on the hot rolled materials). The carbon analysis is done by measuring infrared absorption of the combusted sample, the absorption of the infrared radiation is dependent on absorptivity (a frequency dependent molecular specific characteristic) , optical path length and molar concentration.
Besides the addition of copper and nickel to the material, aluminium and manganese was added to all batches. This was done to keep a consistent level and to get a material that is comparable with earlier studies and normal industrial practice. After heat treatments, cont. and batch. suffix was added to each alloy designation to indicate continuous- and batch annealing respectively. The alloys were analysed as-cast (**Table 2**) and after heat treatments to get the actual composition of the tensile tested materials.

***Table 1** Alloy designations and basic compositions intended for testing. See **Table 2** and **Table 3** for actual chemical composition as analysed from material cast and after heat treatments respectively.*

Alloy designation	Basic intended composition w%.	Comments
R (Reference).	Cu -; Ni -	No addition of elements other than Al and Mn for consistent levels.
Cu1	Cu 0,1; Ni -	
Cu2	Cu 0,2; Ni - ;	
Cu3	Cu 0,35; Ni -	
Cu3 Ni1	Cu 0,35; Ni 0,1	
Cu3 Ni3	Cu 0,35; Ni 0,3	

3.1.2 Production of alloys

Sheet steel from SSAB was alloyed with three levels of copper and with two levels of nickel, an unalloyed sample was also prepared. The sheet steel was cut into suitable sizes and melted in a vacuum furnace (**Figure 14**a) together with the alloying elements. The melt was then cast into ingots with a 35-40mm square cross section and about 150-160mm in length. The ingots (**Figure 14**b) were allowed to air cool to a manageable temperature and were then quenched in water.

Table 2 *Composition analysis by GD-OES (weight %) of the different alloys as cast.*

	Cu	Ni	C	P	S	Si	Al	Mn
Base material as delivered	0,012	0,041	0,0057	0,0072	0,0091	0,0077	0,0312	0,197
Cu1 as cast	0,114	0,042	0,0142	0,0076	0,008	0,0091	0,0704	0,303
Cu2 as cast	0,204	0,042	0,0122	0,0079	0,0077	0,11	0,0577	0,305
Cu3 as cast	0,375	0,042	0,011	0,0079	0,0074	0,0068	0,051	0,293
Cu3 Ni1 as cast	0,377	0,148	0,0119	0,0079	0,0078	0,0076	0,069	0,299
Cu3 Ni3 as cast	0,377	0,367	0,0104	0,0083	0,0074	0,0062	0,0692	0,299
Reference as cast	0,01	0,041	0,041	0,0065	0,0076	0,0063	0,0685	0,303

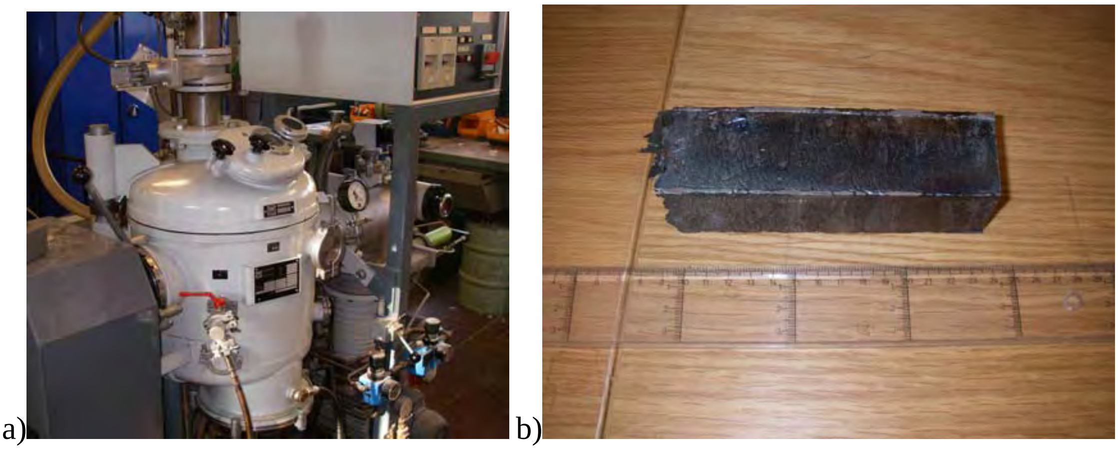


Figure 14: *a) Vacuum furnace used for alloy production. b) Ingot as cast before hot-rolling.*

The ingots were re heated to 1200 °C then hot rolled finishing at 950 °C and thereafter water quenched. The rolled material was then subjected to two different coiling temperatures. Samples were taken from the hot-rolled material for microstructure analysis. The material was then rolled to 70% reduction (**Figure 15**). Two different annealing treatments were carried out. After annealing the different materials are subjected to tensile testing and microstructure analysis.

Figure 15: *The material after cold rolling.*

3.1.3 Annealing and heat-treatments.

The cast ingots were hot-rolled at 1200 ^{o}C with a temperature of 950 ^{o}C at the final pass. After hot rolling the resulting plates were subjected to simulated coiling by heating them to 740 ^{o}C and 530 ^{o}C respectively. They were kept at their respective temperatures for 30 minutes and then cooled in the furnace. The steel coil-simulated at 740 ^{o}C were continuously annealed and the steel coil-simulated at 530 ^{o}C were batch annealed. The chemical composition of the alloys after heat treatments was analysed (**Table 3**).

- **Continuous annealing.** The steel is heated up to 760 ^{o}C in a tube furnace. After reaching 760 ^{o}C the material is cooled down to 490 ^{o}C using forced air convection and is then held there for about 25 seconds in another tube furnace. Thereafter the material is cooled down to 50 ^{o}C, again using forced air convection (**Figure 16**). The aim of this was to simulate the thermal cycle in a hot-dip galvanization line at SSAB tunnplåt in Borlänge.

- **Batch annealing.** The steel is heated at a speed of 30 ^{o}C/h up to 700 ^{o}C, held for 5 hours at 700 °C and then cooled down to room temperature with 30 ^{o}C/h (**Figure 17**).

Table 3 *Composition analysis by GD-OES after heat treatments, carbon composition (marked ▨) analysed with Leco IR-absorption spectography.*

	Cu	Ni	C	P	S	Si	Al	Mn
R cont	0,0110	0,0430	0,0090	0,0056	0,0077	0,0073	0,0737	0,3090
R batch	0,0110	0,0460	< 50 ppm	0,0057	0,0082	0,0077	0,0770	0,3090
Cu1 cont	0,1170	0,0450	0,0101	0,0058	0,0108	0,0107	0,0763	0,3070
Cu1 batch	0,1180	0,0440	< 50 ppm	0,0060	0,0081	0,0096	0,0761	0,3090
Cu2 cont	0,2340	0,0360	0,0104	0,0082	0,0077	0,0140	0,0650	0,3000
Cu2 batch	0,2220	0,0340	< 50 ppm	0,0079	0,0072	0,0120	0,4680	0,3100
Cu3 cont	0,3700	0,0380	0,0113	0,0101	0,0087	0,0120	0,0620	0,3000
Cu3 batch	0,3870	0,0430	< 50 ppm	0,0066	0,0079	0,0084	0,0097	0,3020
Cu3 Ni1 cont	0,3890	0,1410	0,0092	0,0062	0,0092	0,0091	0,0732	0,3020
Cu3 Ni1 batch	0,3890	0,1480	< 50 ppm	0,0068	0,0083	0,0080	0,0746	0,3040
Cu3 Ni3 cont	0,3330	0,2630	0,0102	0,0071	0,0075	0,0080	0,0720	0,3100
Cu3 Ni3 batch	0,3310	0,2560	< 50 ppm	0,0068	0,0079	0,0090	0,0720	0,3100

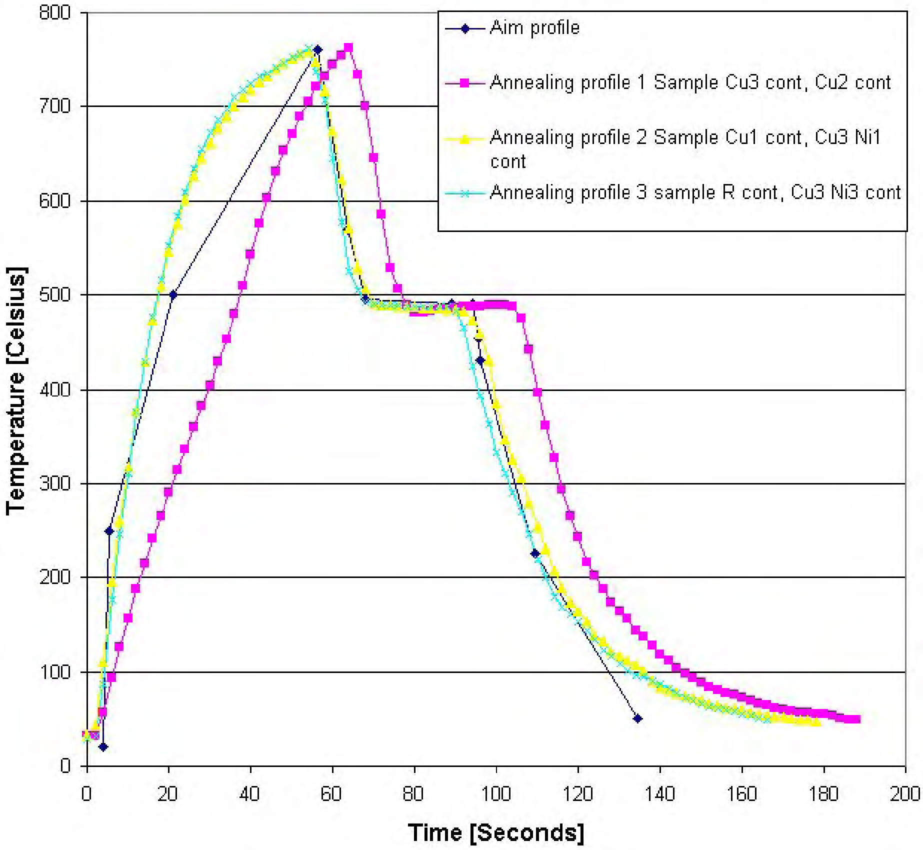


Figure 16 *: Temperature curve for continuous annealing.*

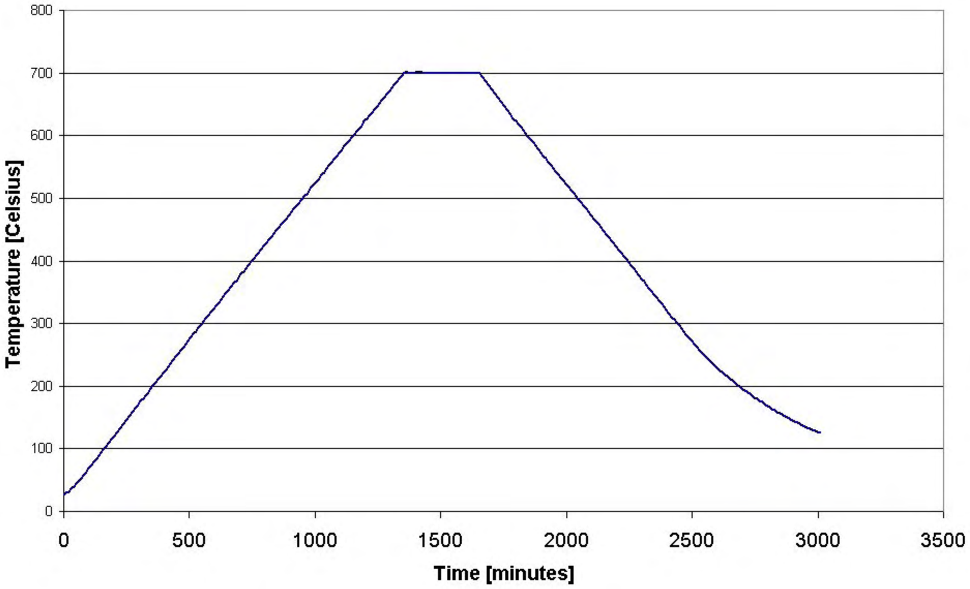


Figure 17 *: Temperature curve for batch annealing.*

3.2 Experimental procedure.

3.2.1 Tensile testing.

For tensile testing, samples are cut from the cold rolled material. The tensile test specimens have a dimension of 15*210 mm with varying thickness, ranging from 1.5 mm down to 0.7 mm. The test pieces have a small waist of 1 mm machined from each side (**Figure 18**). The samples were strained with a speed of 6 mm/minute in an Instron series 9 type 4505 tensile testing rig. The elongation and contraction of the material is measured with two extensiometers.

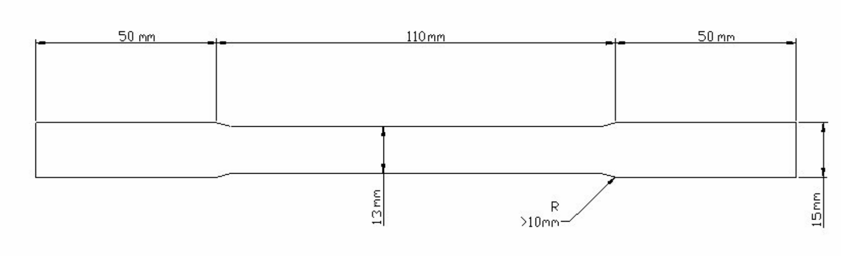


Figure 18 Tensile test specimen dimensions.

3.2.2 Hardness testing.

Vickers hardness testing is performed on a micro hardness tester model FM-700 manufactured by Future Tech. The specimens prepared for microstructure analysis is also used for hardness testing. The materials are polished and a series of five imprints, at a distance of 1.5 mm from each other, are made on each alloy type (**Figure 19**), impression force load used is 1kg.

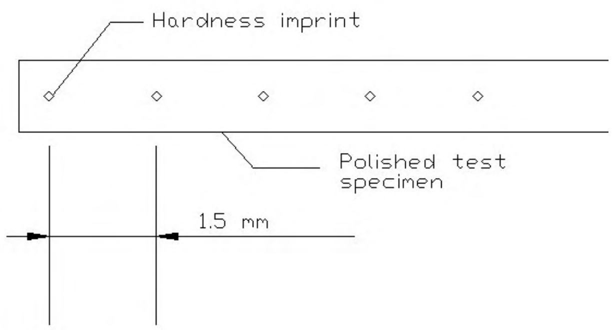


Figure 19 Hardness testing set-up

3.2.3 Microstructure analysis.

Samples are prepared for microscopy from both the hot- and cold-rolled materials. The samples are polished and etched with 2% Nital solution. The different materials are then studied in an optical microscope model LEICA DM IRM to classify grain structure, grain size and microstructure evolution.
The fracture surfaces of four samples, Cu1 Cont.-1, Cu1 Cont.-2, Cu3 Ni3 Cont.-1, Cu3 Ni3 Cont.-3, are studied in a scanning electron microscope (**Figure 36**, **Figure 37**). Grain size is determined by using a view insert slide in a microscope that displays a circle over the view in the microscope ocular. The apparent size of the circle at a given magnification is determined by using a line scale with a 0.01 mm spacing and measuring the circle diameter in the view. When the circumference is known the grain size can be calculated by

dividing length of the circumference with the number of times the circle intersects a grain (**Figure 20** a). By using the same principle, but with an insert that shows perpendicular lines in the view (**Figure 20** b), the quota between the length and cross axis of the grains is determined.

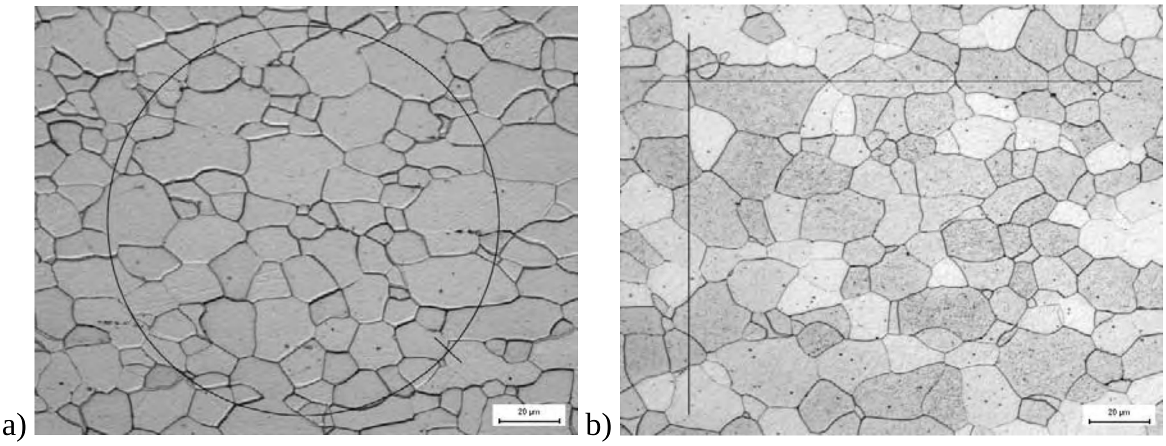


Figure 20 *a) View of the circle intercept as it appears in the microscope b) View of the line intercept in the microscope.*

Crystal grain orientation is measured by using an Electron Back Scatter Diffraction (EBSD) detector. The EBSD determines crystal type and orientation by taking a picture of the diffraction pattern caused by an electron beam on a phosphor screen (**Figure 21**). The electron beam is aimed at the surface of a sample and will be scattered and diffracted by the crystal planes in different grains of the sample. By measuring the position and relative angle of the Kikuchi lines in the resulting diffraction pattern the crystal orientation can be determined[21]. Samples for the texture measurement were prepared by grinding and polishing specimens taken from the tensile test pieces; final polishing was done with a mild etching agent to remove the deformation from grinding and polishing of the surface. The section of the sample surface was made to be as parallel to the rolling direction as possible to avoid having to rotate data after measuring. The measurement was performed in a LEO FEG-SEM 1530 Gemini electron microscope. Approximately 3000 crystal orientation measurements were taken from each sample.

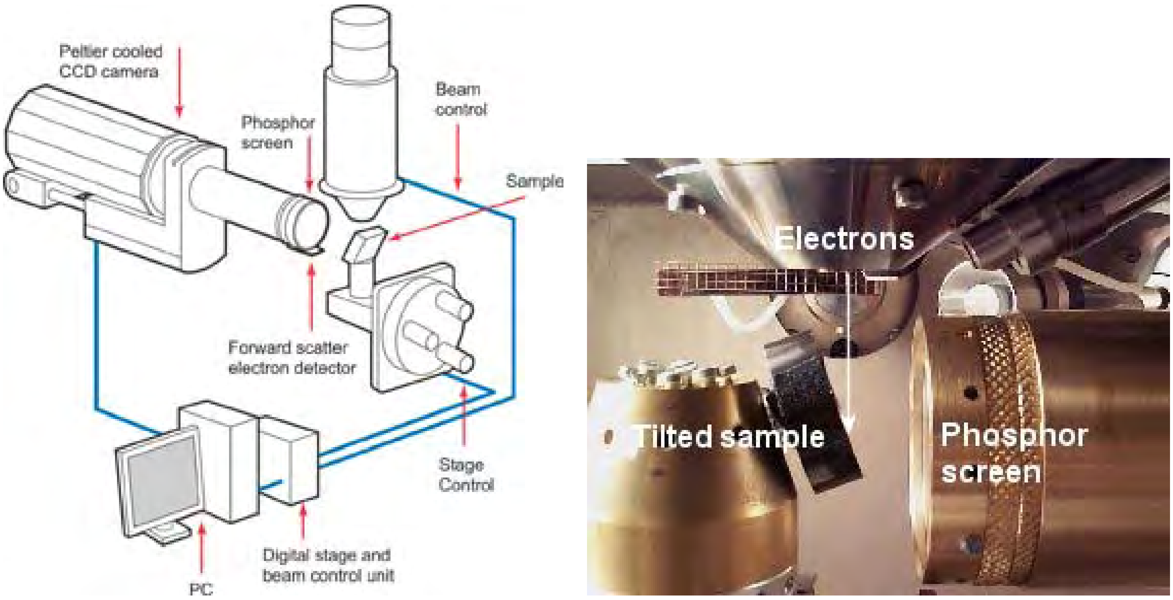


Figure 21 *EBSD set-up. Top: Principal configuration of the test set-up, bottom: view of test chamber in an actual SEM.*

3.3 Results.

3.3.1 Tensile and hardness testing.

The graphic representation of the tensile test summary in **Table 4** is shown in **Figure 22** and **Figure 23**. The results of the hardness testing are presented in **Figure 24**. **Figure 25** and **Figure 26** shows the stress-strain curves of the continuously annealed material while **Figure 27** and **Figure 28** show the stress-strain curves of the batch annealed material. The result of the mechanical testing can be summarised as follows:

Continuously annealed material:

- Lowering of the yield strength with increasing copper content in materials alloyed with copper alone (**Figure 22** a).
- Increase of yield strength in material alloyed with copper and nickel.
- No significant change in ultimate tensile strength in regards to copper (**Figure 22** a).
- Increase in elongation with increasing copper content in materials alloyed with only copper, lowering of elongation in materials alloyed with copper and nickel (**Table 4**).
- Slight decrease of Lüders strain with increasing copper content and in combination with nickel (**Figure 25**,**Figure 26**).
- Alloying with nickel decreases hardness, only the highest copper content (Alloy designated Cu3 containing about 0.37 w% copper) shows significant increase in hardness (**Figure 24**).

Batch annealed material.

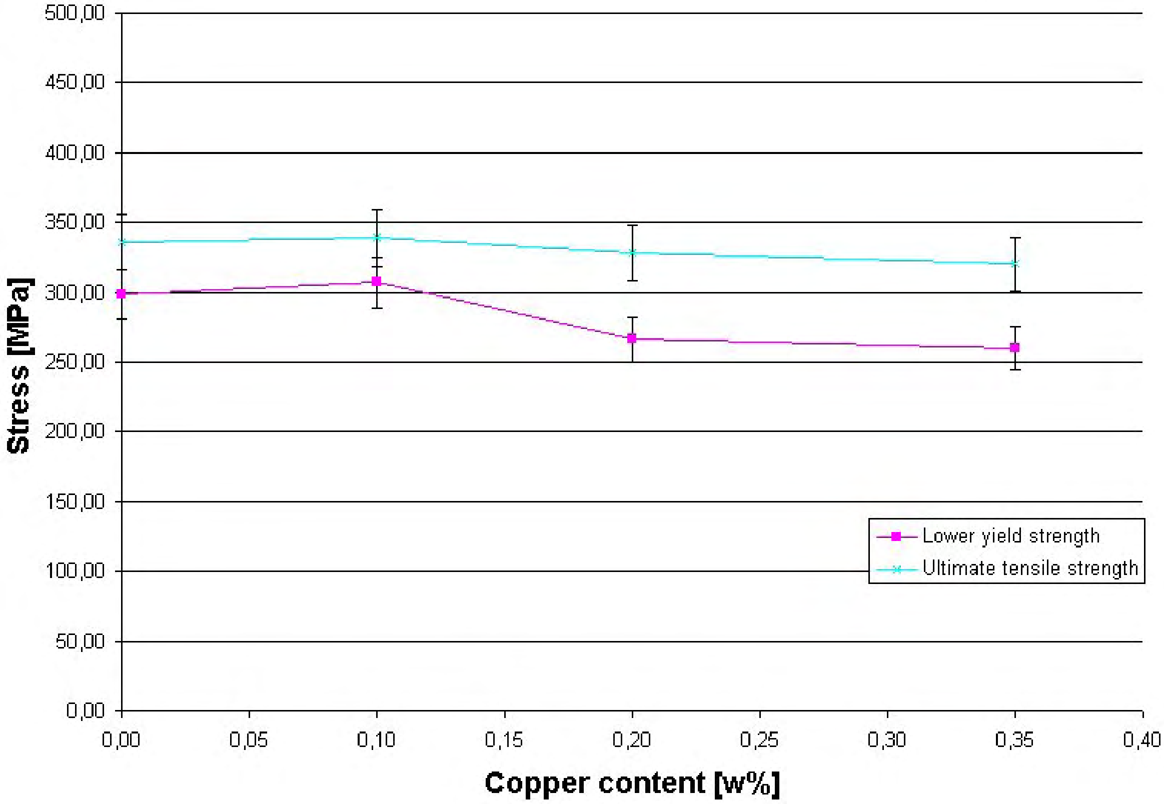
- Increase of yield strength and ultimate tensile strength with increasing copper content. Yield strength also increase in combination with nickel.
- Lowering of elongation with copper content and further lowering in combination with nickel (**Table 4**).
- Lüders strain is <1% of total strain in all of the alloys (**Figure 27** a-c, **Figure 28** a-c).
- No apparent trend in hardness, value varies around 97-98 HV (**Figure 24**).

Table 4 *Mechanical properties in batch- and continuously- annealed materials. UYS= Upper Yield Strength; LYS= Lower Yield Strength; UTS= Ultimate tensile strength.*

Continuously annealed materials:	UYS (MPa)		LYS (MPa)		UTS (MPa)		Hardness [Vickers]		Elongation in %	
	Avg.	Std. dev.	Avg.	Std. dev.	Avg.	Std. dev.	Avg.	Std. dev.	Avg.	Std. dev.
R Cont.	426,5	23,38	298,25	22,68	335,75	12,98	143,5	4,3	32,7	5,11
Cu1 Cont.-1	431		307		339		146,3	8,98	21	
Cu2 Cont.	351	8	266	8	328,5	4,5	140,5	23,6	37,2	5,5
Cu3 Cont.	316,75	17,16	260	6,52	320	5,47	164,0	10,9	36,8	1,92
Cu3 Ni1 Cont.	403	11,22	296,33	10,66	347,33	3,09	93,3	6,7	35,0	4,96
Cu3 Ni3 Cont.	446,33	16,21	316	8,52	365	3,55	114,3	2,5	17,33	2,05

Batch annealed materials:	UYS (MPa)		LYS (MPa)		UTS (MPa)		Hardness [Vickers]		Elongation in %	
	Avg.	Std. dev.	Avg.	Std. dev.	Avg.	Std. dev.	Avg.	Std. dev.	Avg.	Std. dev.
R Batch.	120,25	8,52	114,75	5,35	263,5	2,59	98,8	4,8	45,0	5,78
Cu1 Batch.-1	117,5	5,59	110,5	2,18	270,5	5,17	96,5	3,2	41,56	2,94
Cu2 Batch.	151,5	21,69	142,5	23,93	251	8,21	102,5	4,3	39,26	1,78
Cu3 Batch.	143,66	3,39	136,67	3,68	282	2,82	92,3	3,5	42,33	1,69
Cu3 Ni1 Batch.	175,5	19,83	161,75	16,84	286,25	3,56	98,8	6,3	33,25	2,86
Cu3 Ni3 Batch.	283,33	9,74	260,33	10,20	464	5,71	95,3	1,6	35	4,08

a) Tensile strength by copper content in continuously annealed materials



b) Tensile strength by copper content in batch annealed materials

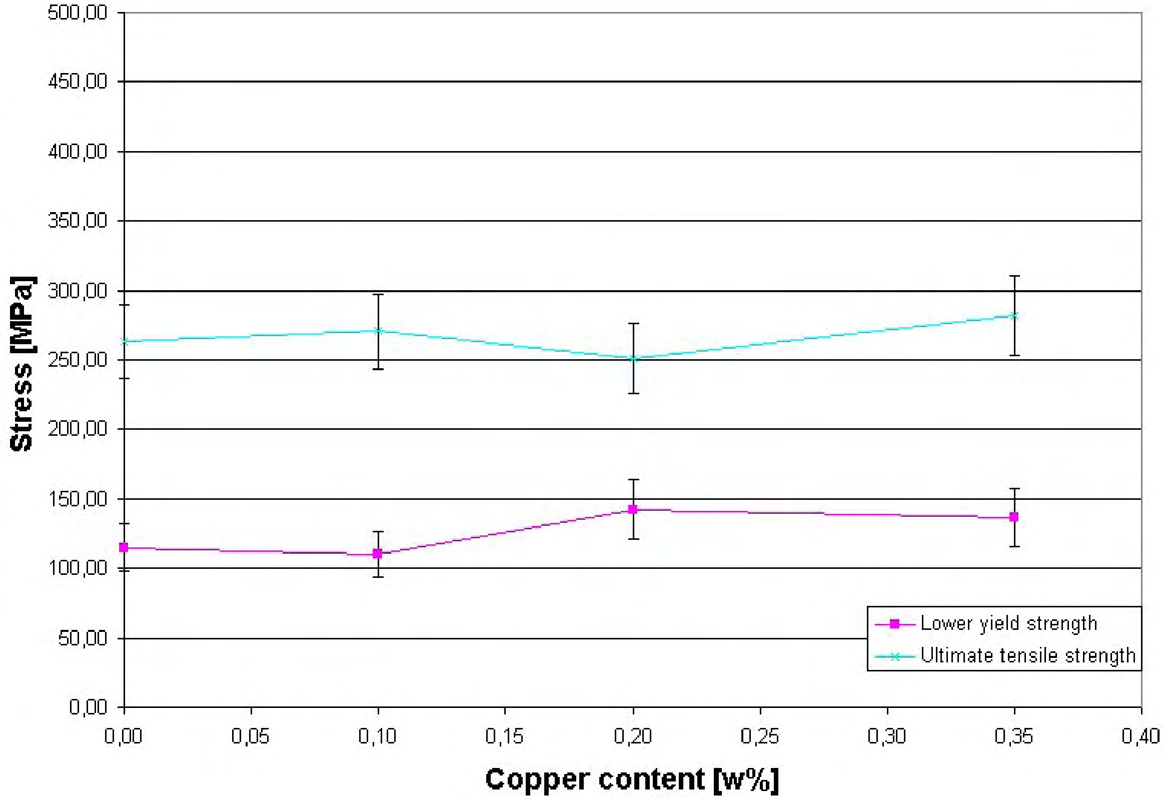
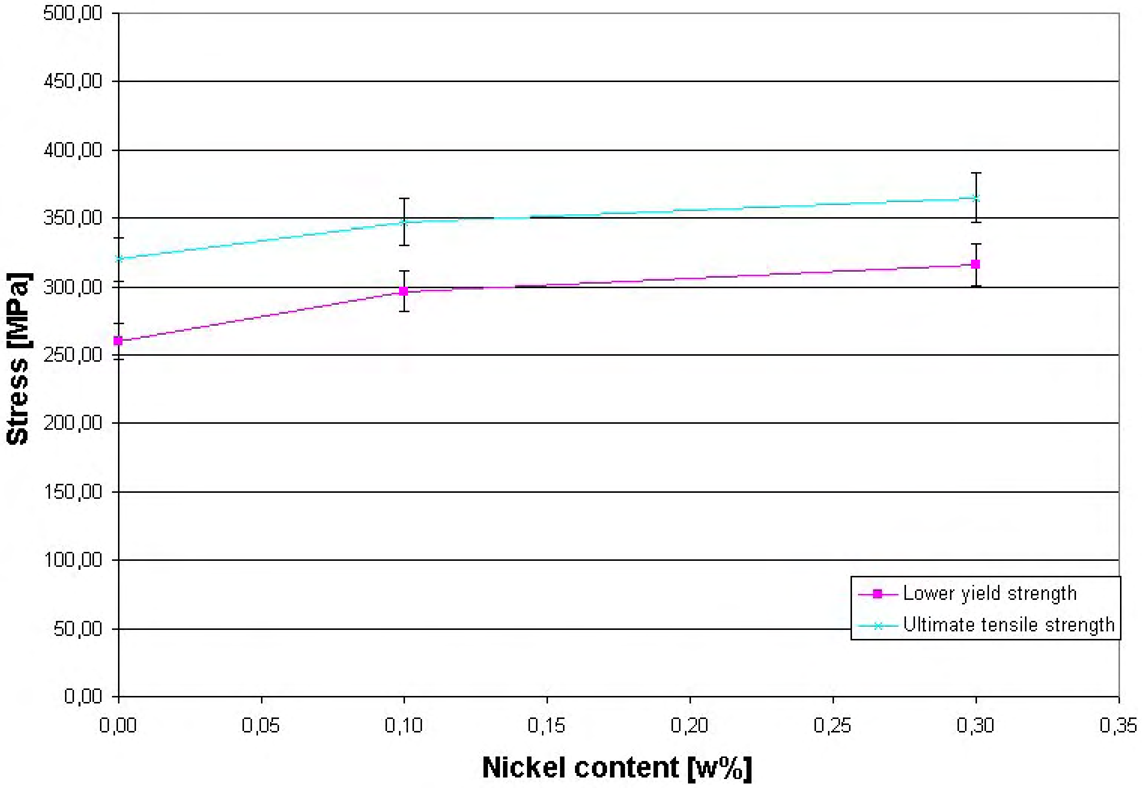


***Figure 22** a) Shows tensile strength by copper content in continuously annealed material and b) shows tensile strength by copper content in batch annealed materials.*

a) Tensile strength by Nickel content in continuously annealed materials.



b) Tensile strength by Nickel content in batch annealed materials.

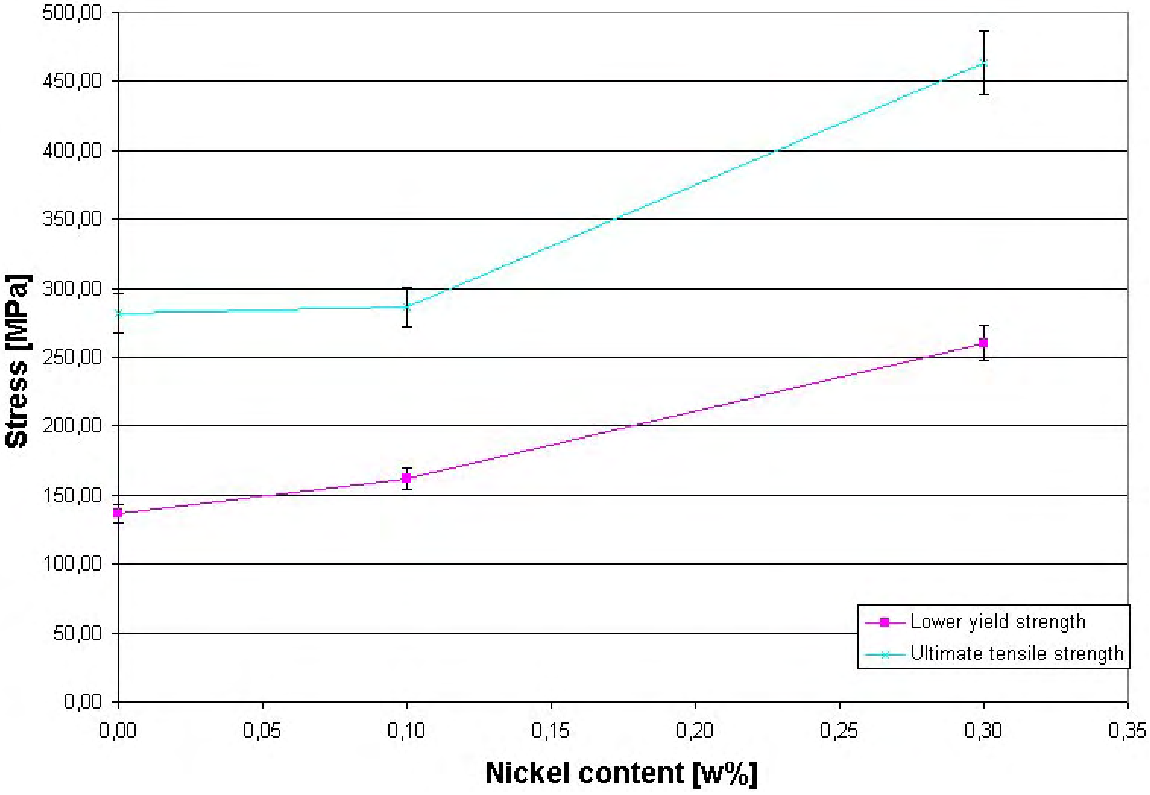


Figure 23 *Tensile strength by varying copper content in a) continuously annealed material and b) batch annealed material. All of the alloys in these diagrams have a copper content of 0.35 w%.*

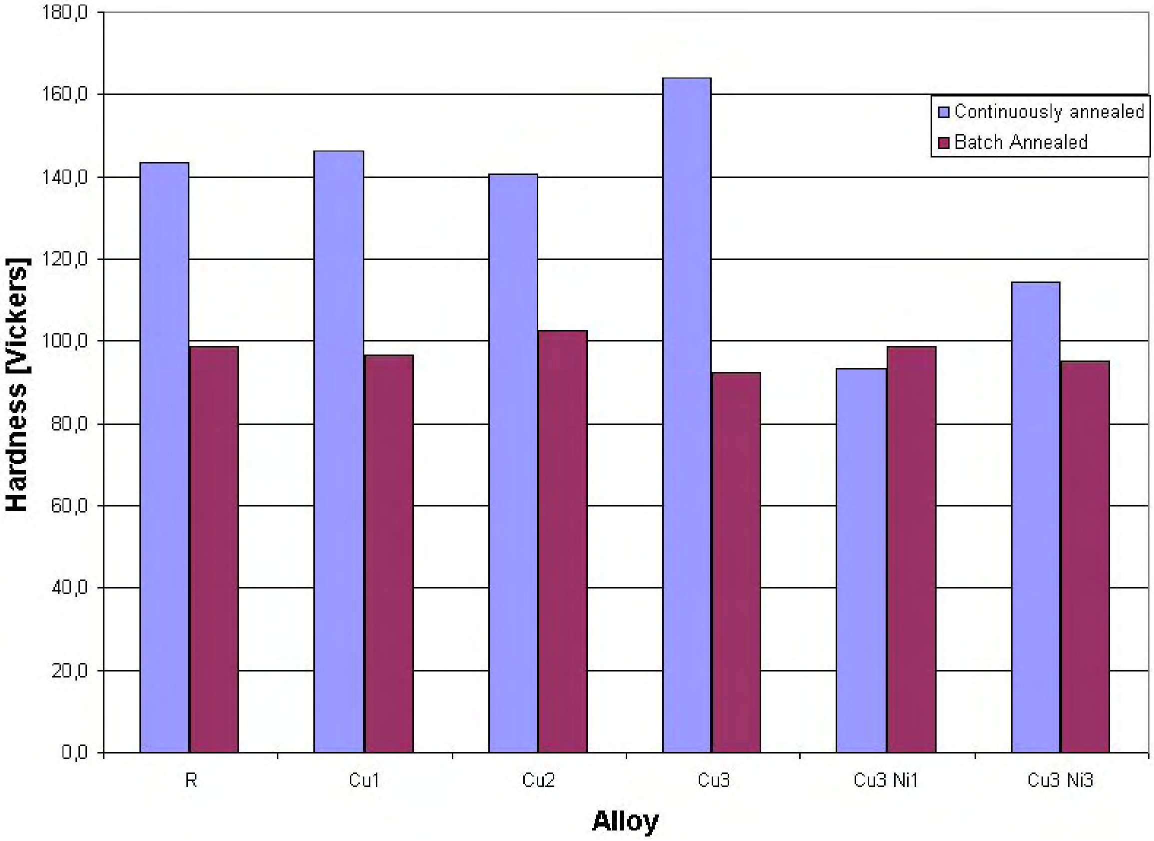
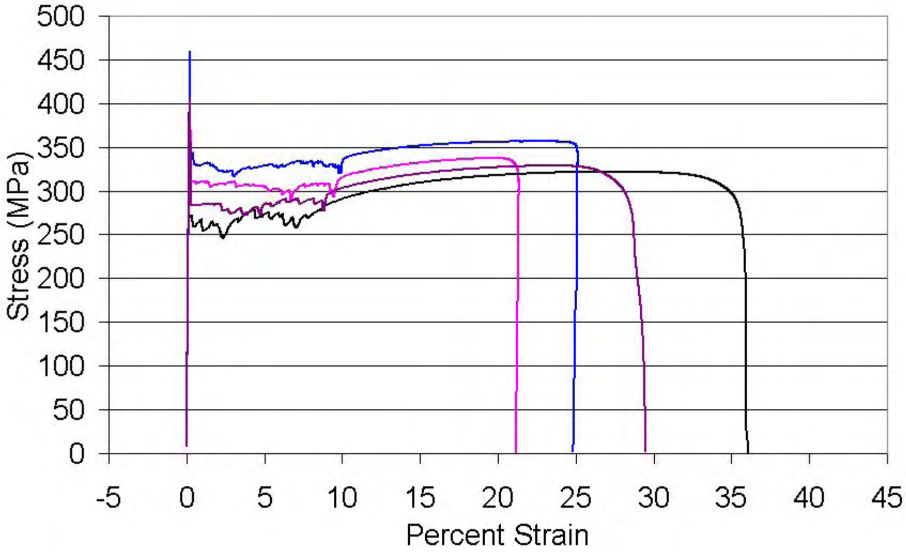
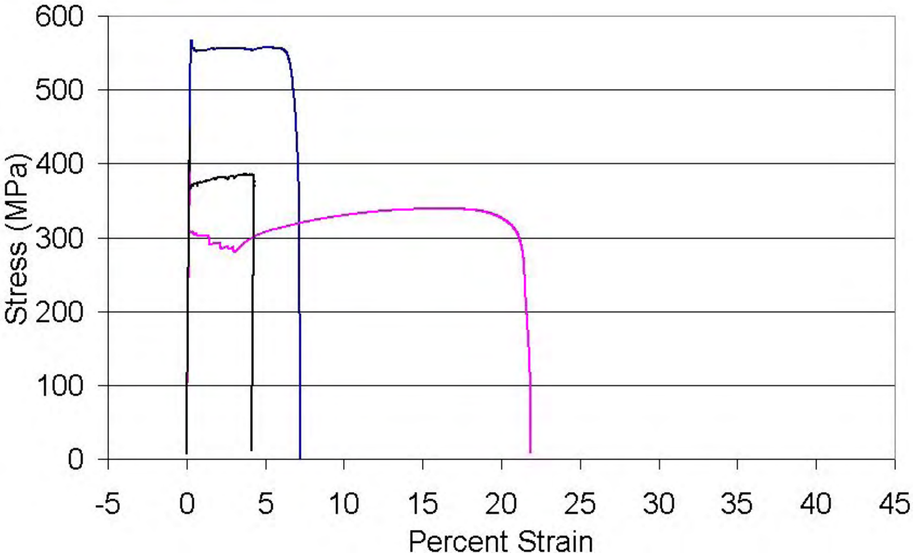


Figure 24 *Hardness[Vickers] in batch- and continuously- annealed alloys.Indents made with 1 kg load.*

a) R cont



b) Cu1 cont



c) Cu2 cont

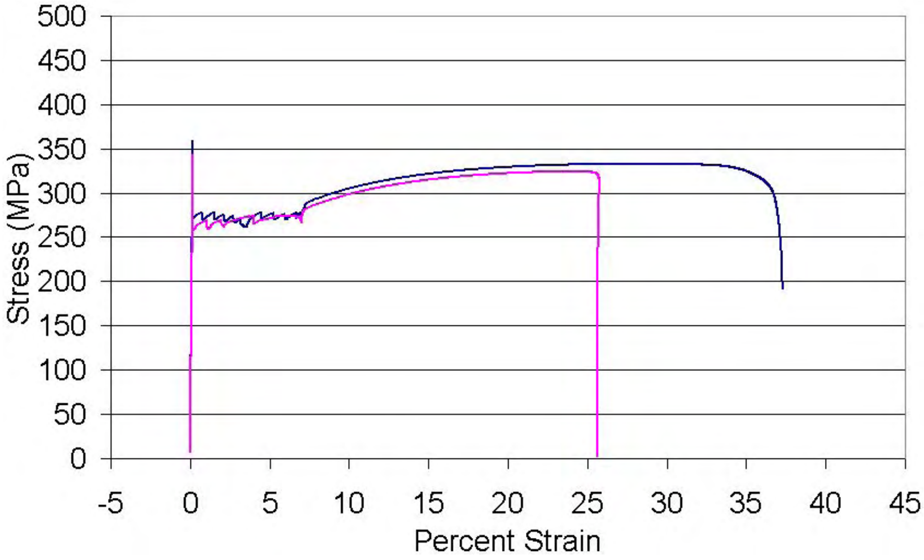
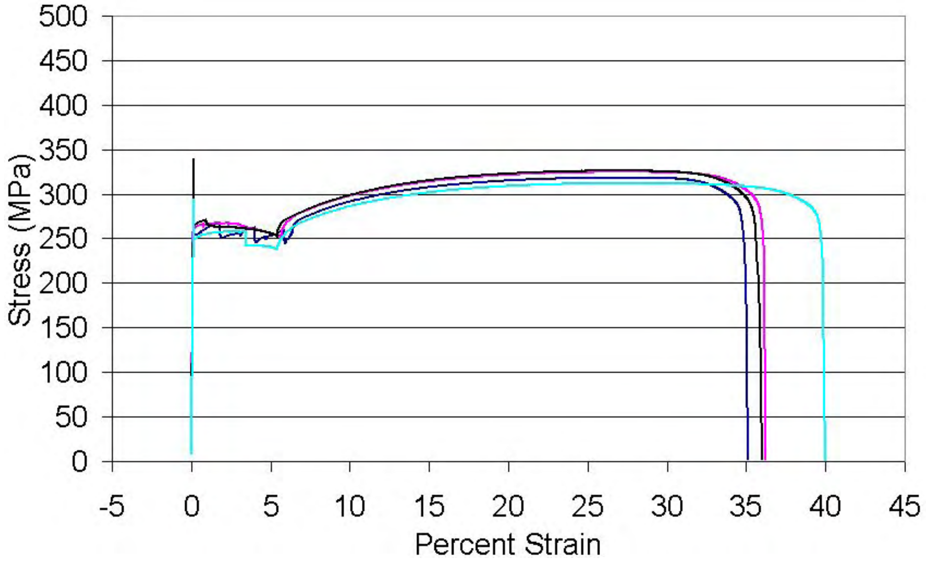
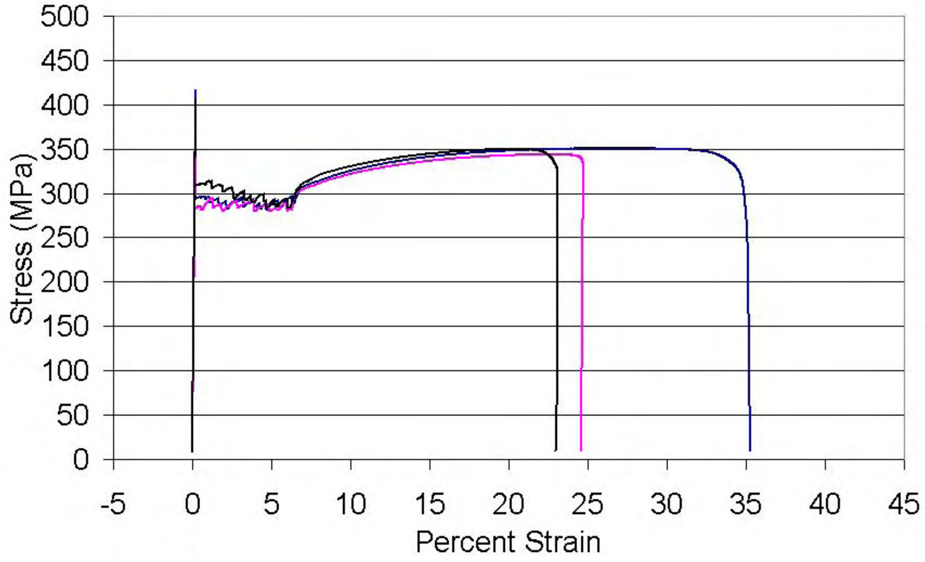


Figure 25 *Stress-Strain curves for continuously annealed materials a) R cont. b) Cu1 cont c) Cu2 cont.*

a) Cu3 cont.



b) Cu3 Ni1 cont.



c) Cu3 Ni3 cont.

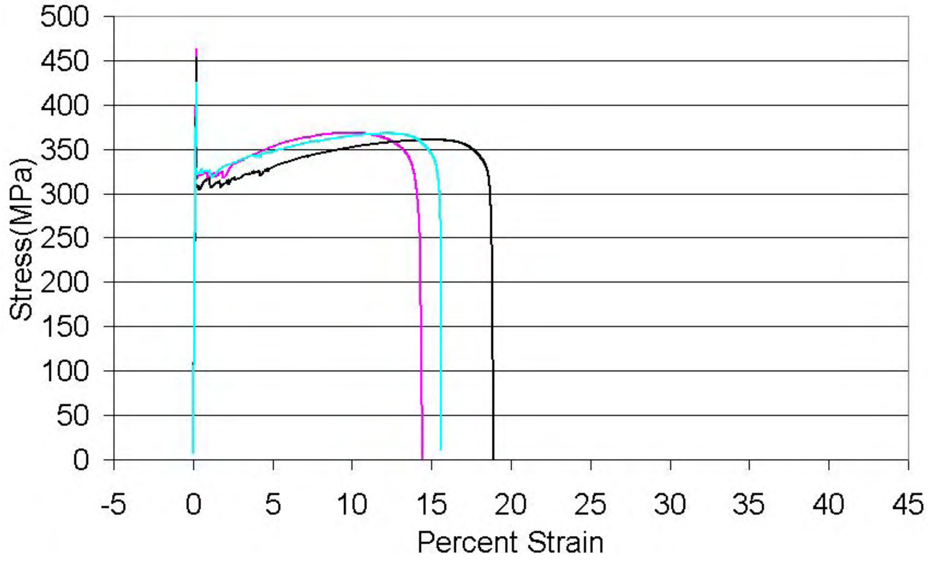
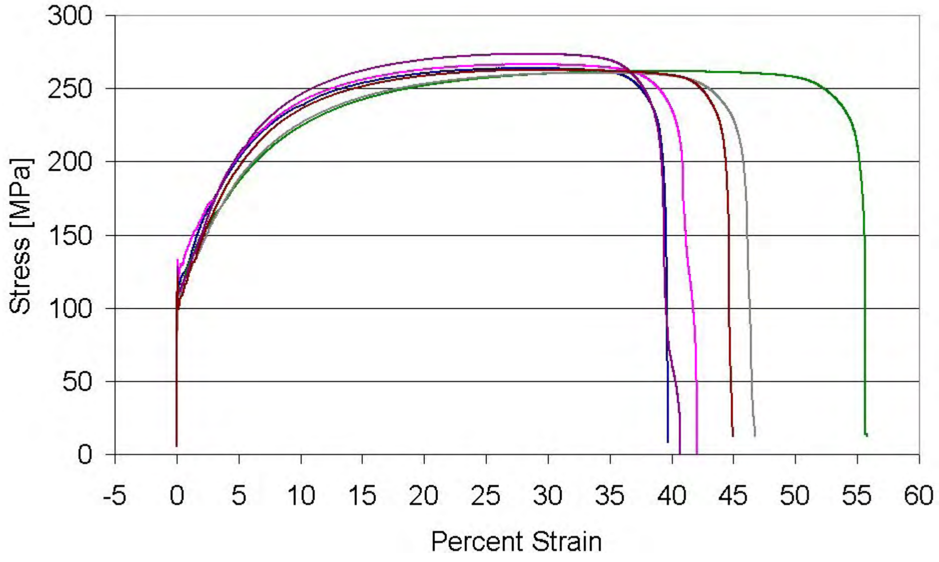
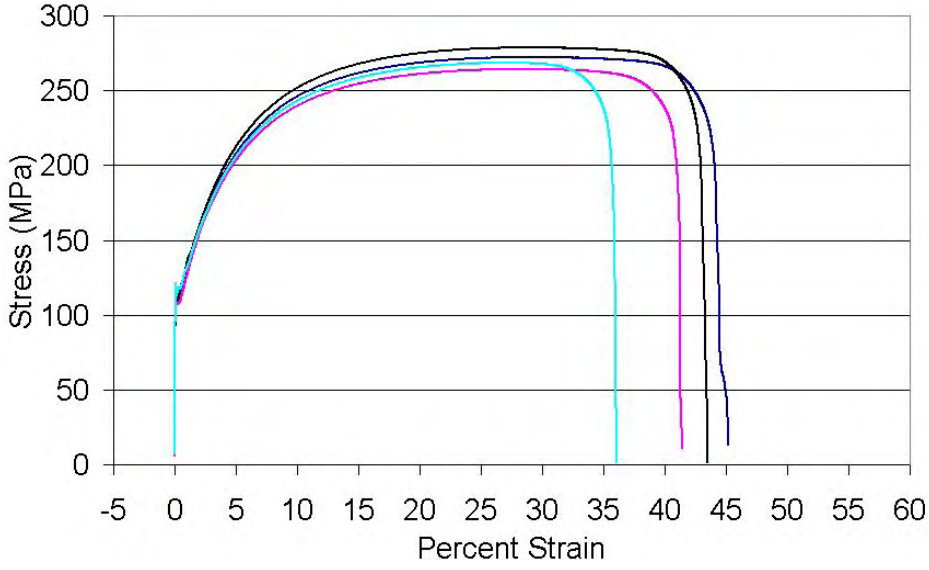


Figure 26 *Stress-Strain curves for continuously annealed materials a) Cu3 cont.. b) Cu3 Ni1 cont. c) Cu3 Ni3 cont.*

a) R batch



b) Cu1 batch



c) Cu2 batch

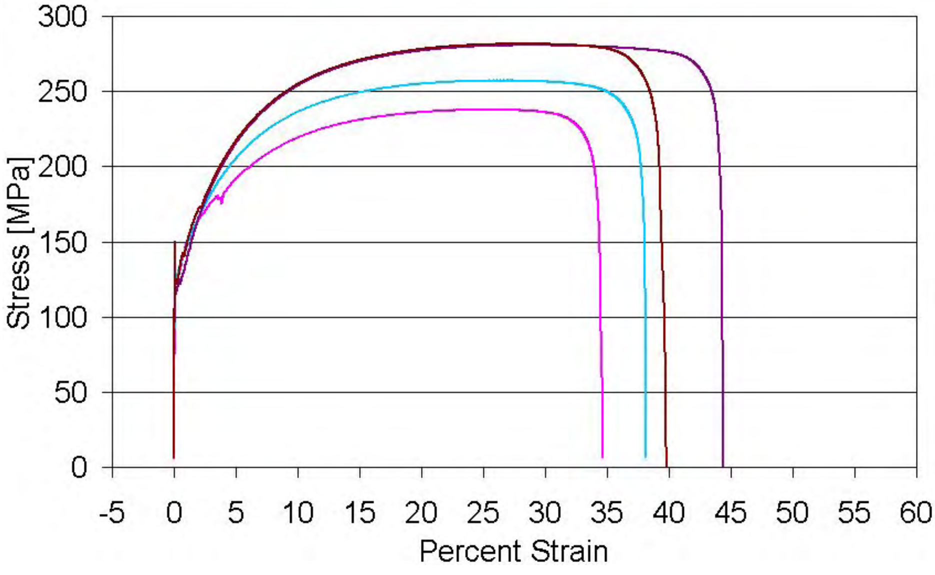
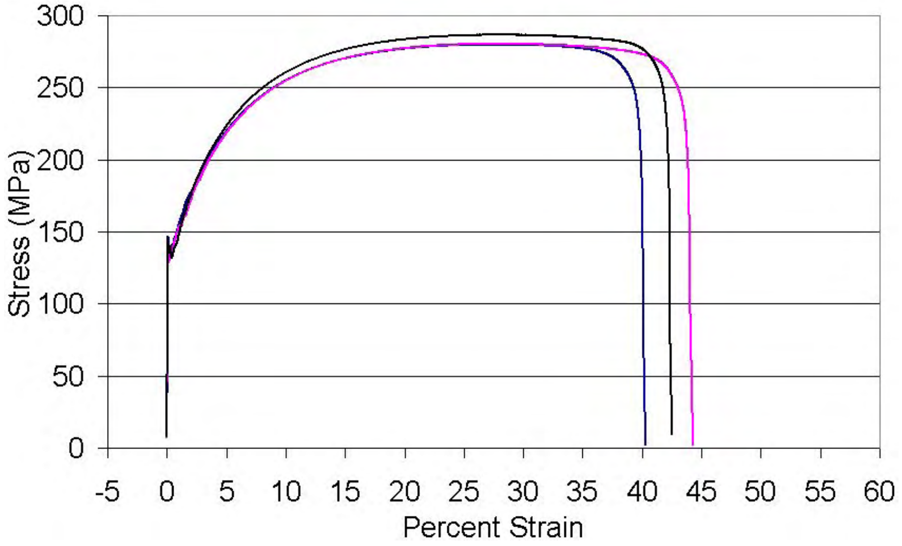
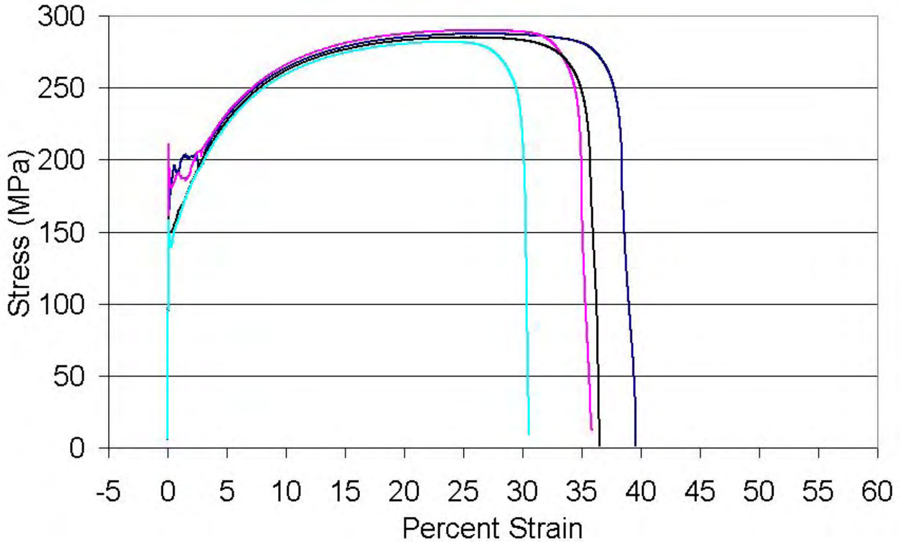


Figure 27 *Stress-Strain curves for batch annealed materials a) R batch b) Cu1 batch c) Cu2 batch.*

a) Cu3 batch



b) Cu3 Ni1 batch



c) Cu3 Ni3 batch, note that the scale for the y-axis is different from the other figures of batch annealed materials

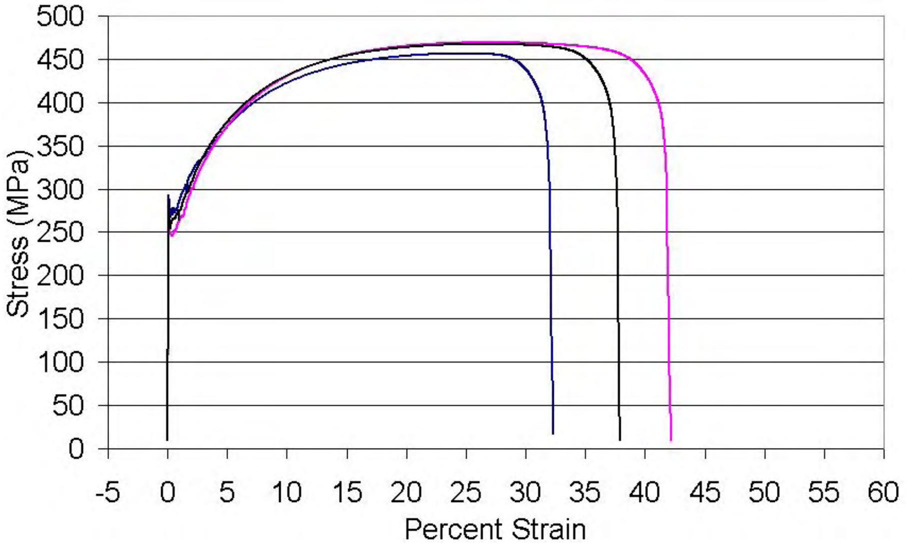


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***Figure 28** Stress-Strain curves for batch annealed materials a) Cu3 batch b) Cu3 Ni1 batch c) Cu3 Ni3 batch.*

3.3.2 Microstructure analysis.

Figure 30 shows the average grain size in both batch continuously annealed material while **Figure 31**shows the ratio between horizontal and vertical grain axes. **Figure 32** and **Figure 33** presents the optical micrographs of the continuously annealed materials while **Figure 34** and **Figure 35**shows the optical micrographs of the batch annealed materials. **Figure 36** and **Figure 37** shows the electron micrographs of the fracture surface of four samples.
The light optical microscope pictures (**Figure 32** - **Figure 35**) are oriented so that the rolling direction is horizontal in the picture plane and the transverse direction is normal to it (**Figure 29**).
Optical metallographic observations can be summarised as follows:

Continuously annealed material:

- Smaller grain size compared to batch annealed material.
- Coarsest grains at 0,35 w% copper, small differences otherwise (**Figure 30**).
- Slight elongation of grains in the rolling direction (**Figure 31**).

Batch annealed material.

- Finer grains at intermediate copper level and with nickel, coarsest structure as unalloyed and with 0,35 w% copper (**Figure 30**).
- Smaller relative differences in grain size between compositions than in the continuously annealed material. More pronounced ‘pancake’ structure (elongated grains in the rolling direction as can clearly be seen in **Figure 35** a-f) compared with the continuously annealed materials (**Figure 31**).

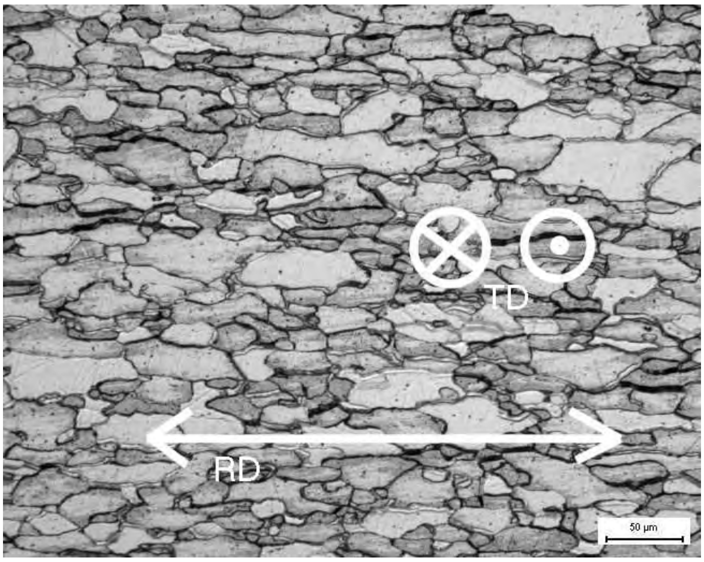


Figure 29 *Rolling direction (RD) an transverse direction (TD) in the picture plane.*

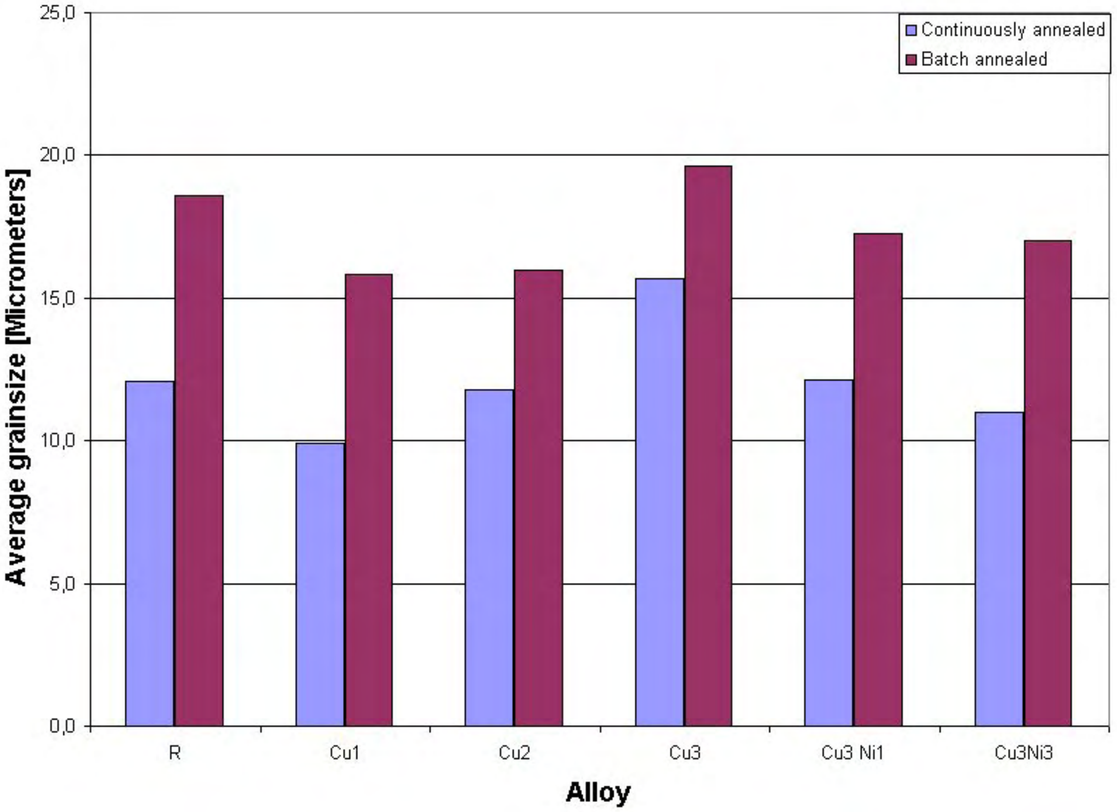


Figure 30 *Average grain size determined with circle intercept (**Figure 20**a) by alloy type and heat treatment*

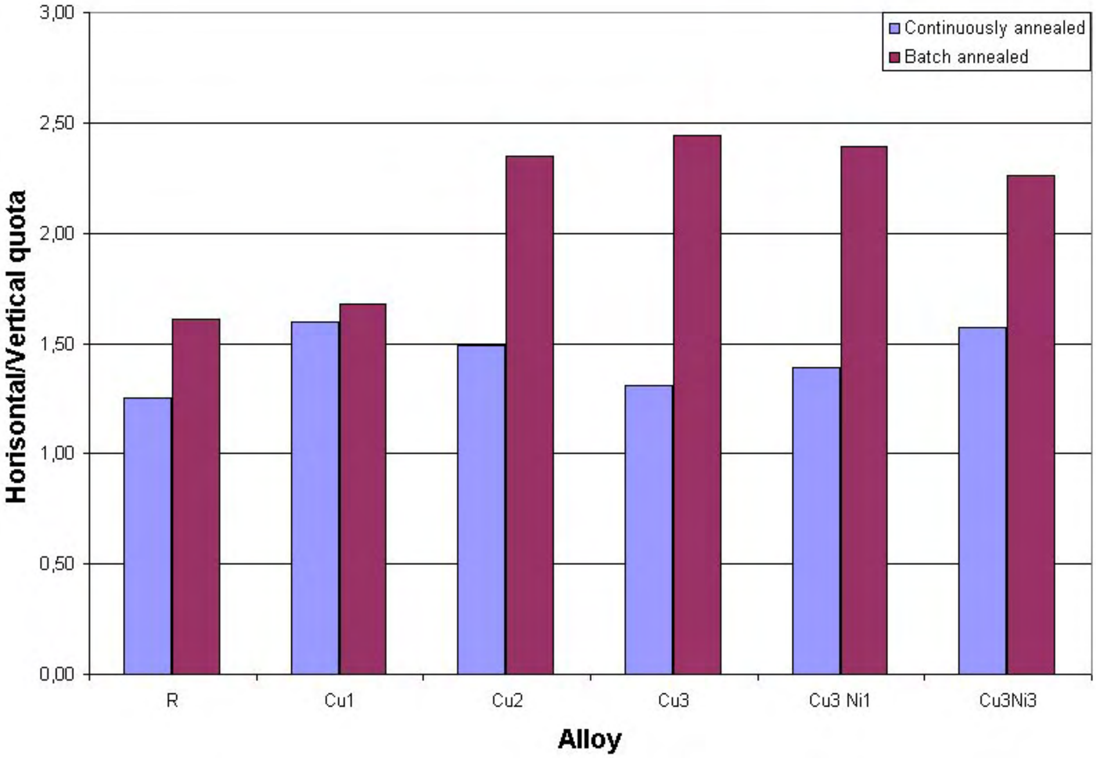
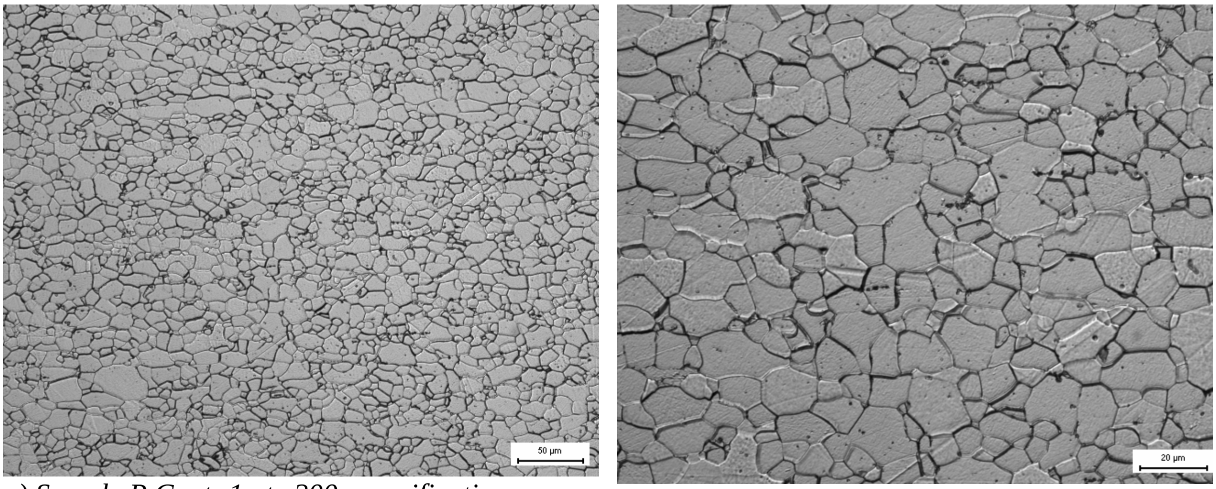
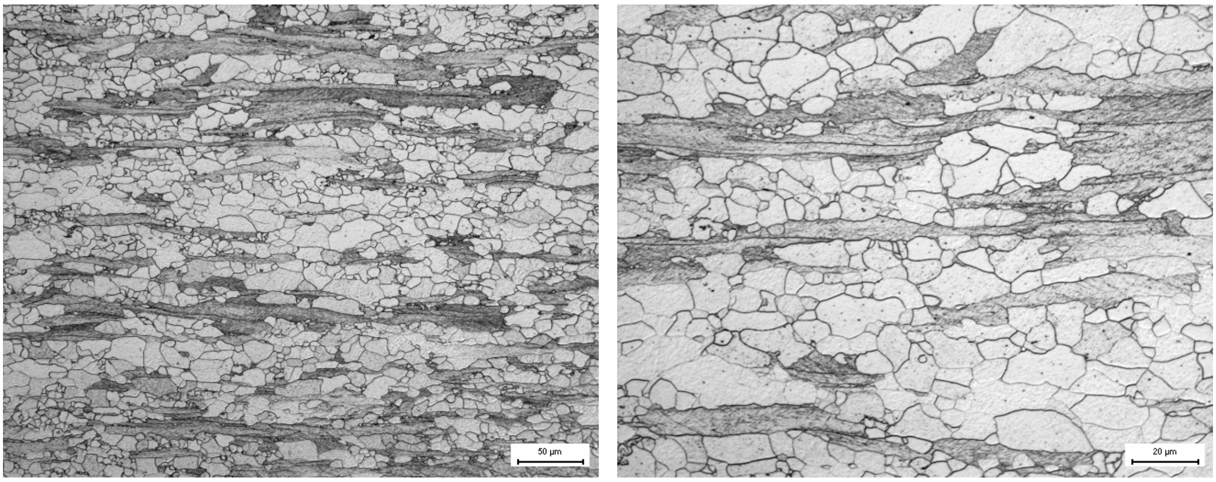


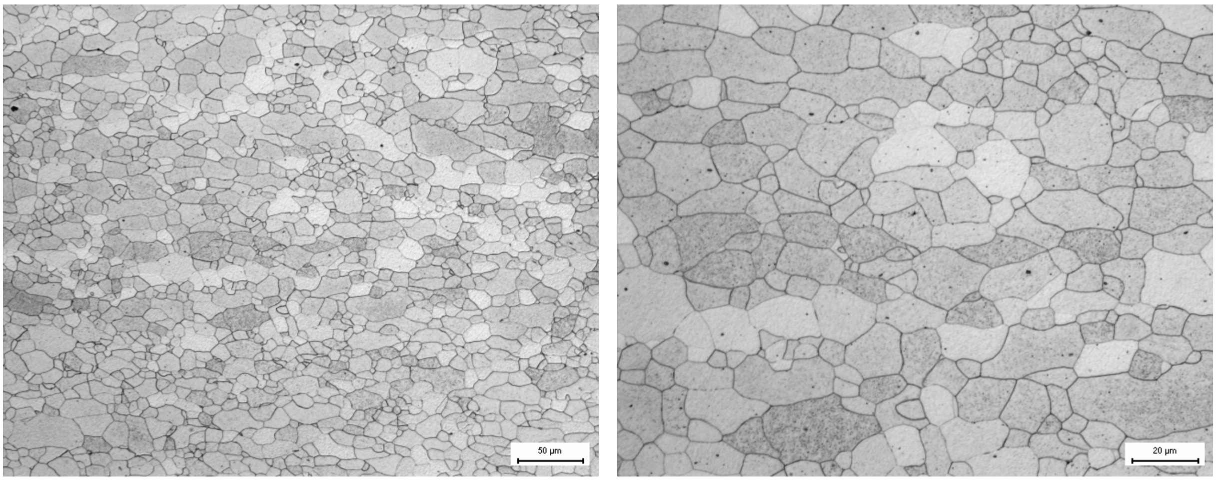
Figure 31 *Ratio between average horizontal axis length, determined with line intercept (**Figure 20**b), and average vertical axis length. A value of exactly 1 would mean that the horizontal and vertical grain axes are of equal length, a value of 2 means that the horizontal axis is twice the length of the vertical etc. The figure shows that all of the materials have at least some elongation of grains in the rolling direction.*



a) Sample R Cont.-1 at x200 magnification. *b) Sample R Cont.-1 at x500 magnification.*

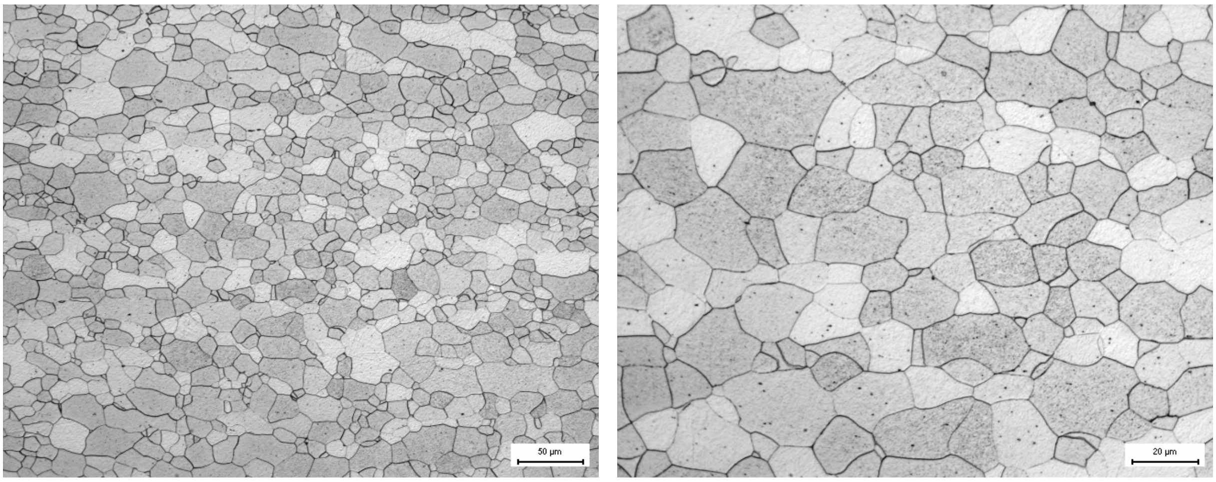


c) Sample Cu 1 Cont.-2 at x200 magnification. *d) Sample Cu 1 Cont.-2 at x500 magnification.*



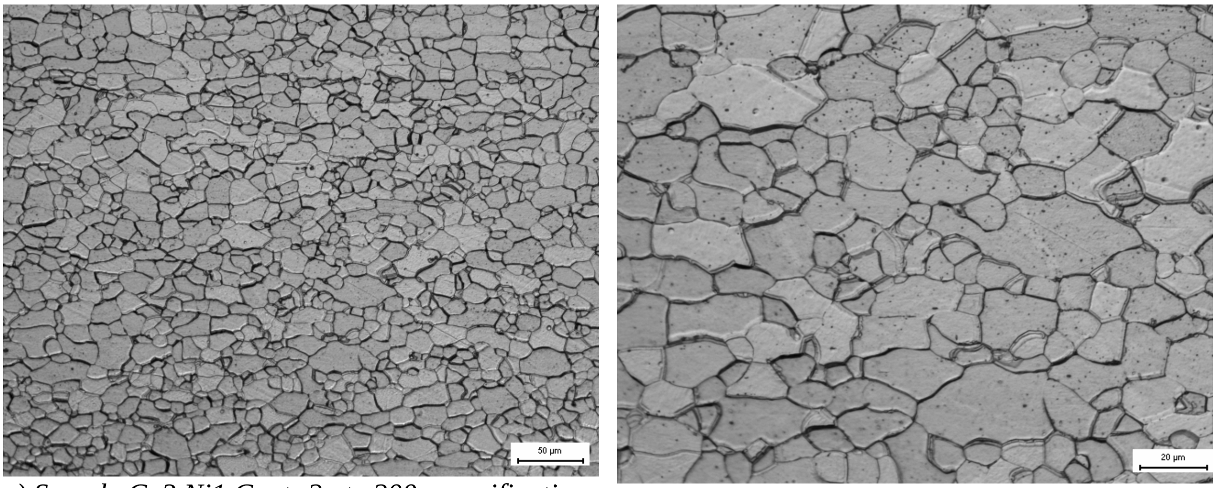
e) Sample Cu2 Cont..-1 at x200 magnification. *f) Sample Cu2 Cont.-1 at x500 magnification.*

Figure 32 *Optical micrographs of the microstructure in the continuously annealed alloys. The material is ferritic and the structure has been made visible by etching with 2% Nital solution (ethanol and nitric acid)*



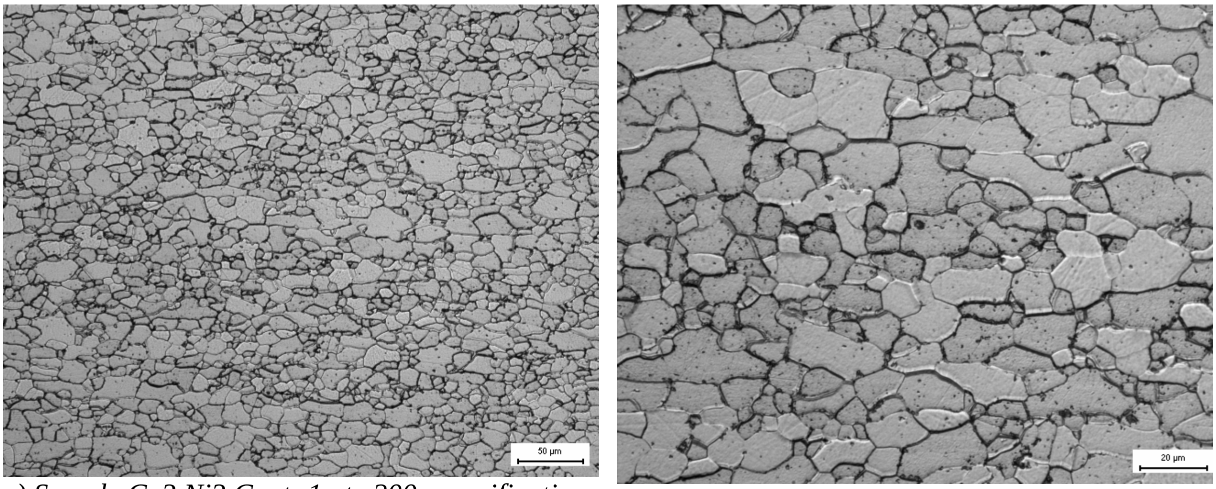
a) Sample Cu3 Cont.-1 at x200 magnification.

b) Sample Cu3 Cont.-1 at x500 magnification.



c) Sample Cu3 Ni1 Cont.-2 at x200 magnification.

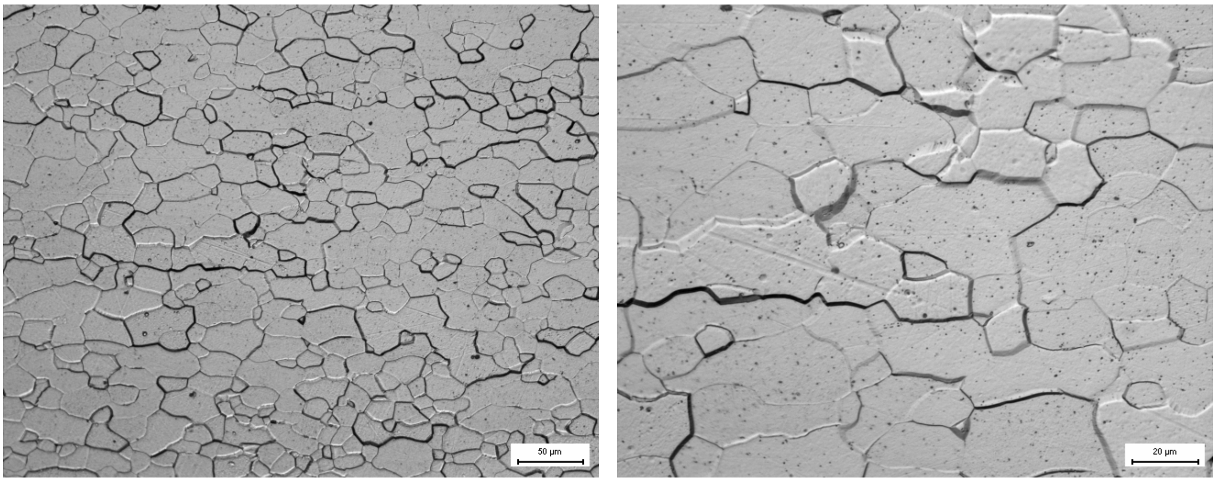
d) Sample Cu3 Ni1 Cont.-2 at x500 magnification.



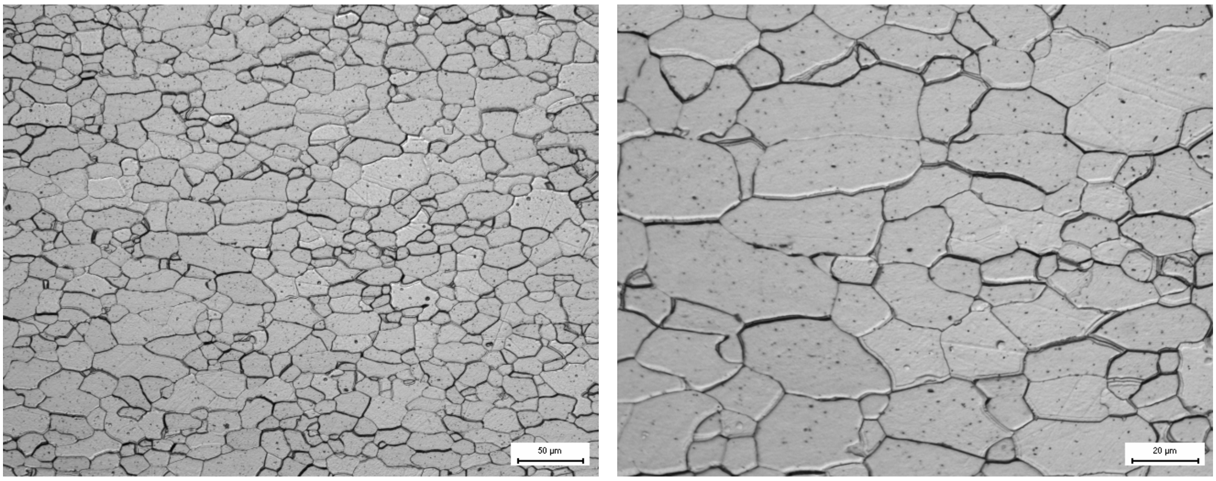
e) Sample Cu3 Ni3 Cont.-1 at x200 magnification.

f) Sample Cu3 Ni3 Cont.-1 at x500 magnification.

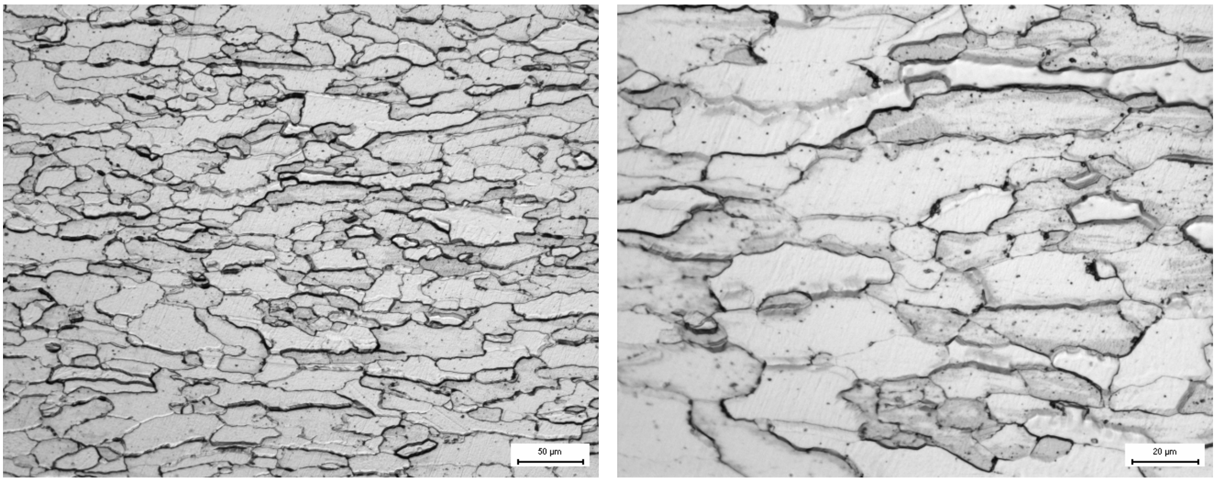
Figure 33 *Optical micrographs of the microstructure in the continuously annealed alloys. The material is ferritic and the structure has been made visible by etching with 2% Nital solution (ethanol and nitric acid)*



a) Sample R Batch.-4 at x200 magnification. *b) Sample R Batch.-4 at x500 magnification.*

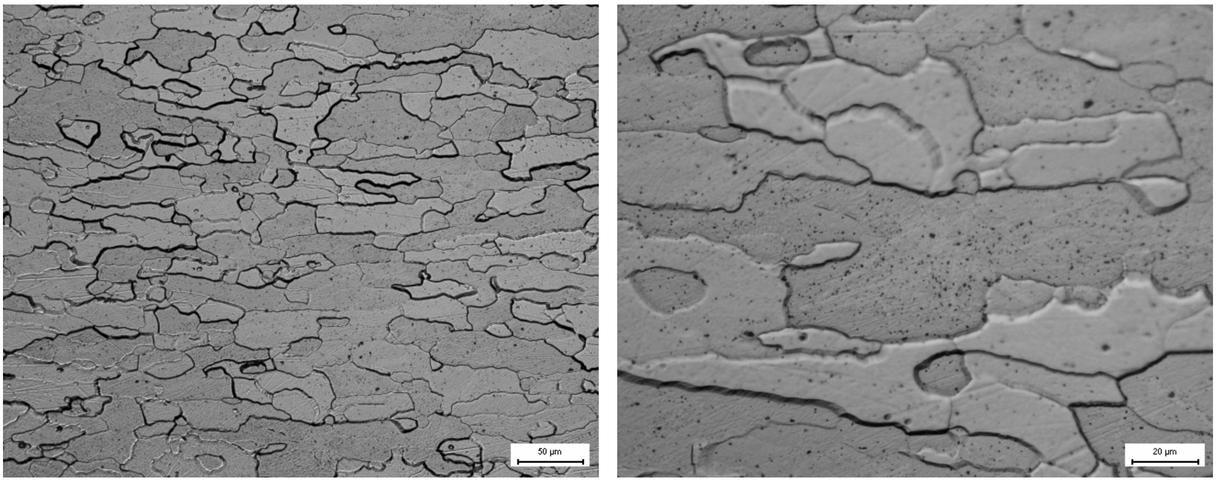


c) Sample Cu 1 Batch.-1 at x200 magnification. *d) Sample Cu 1 Batch.-1 at x500 magnification.*

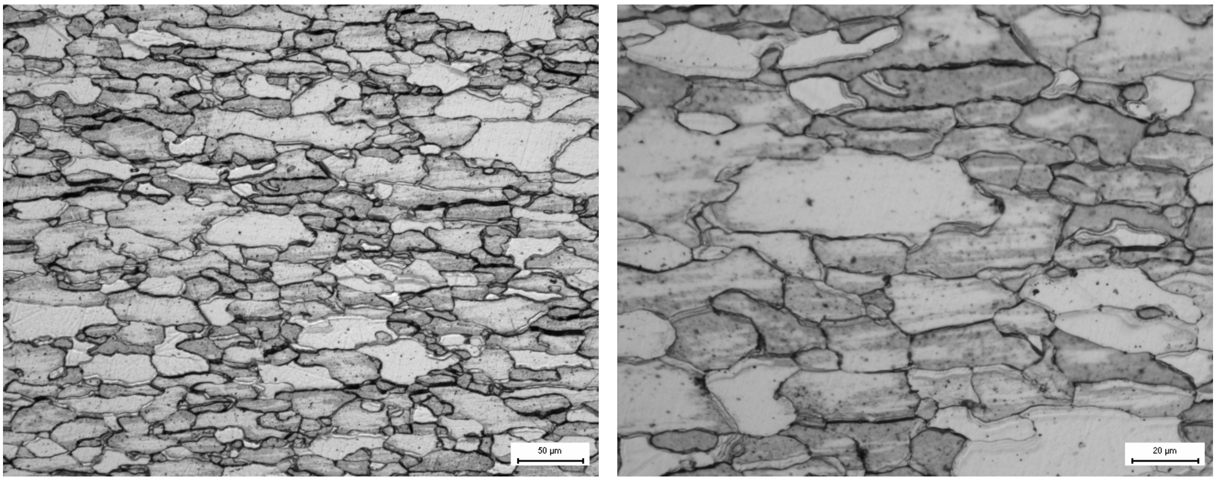


e) Sample Cu2 Batch.-4 at x200 magnification. *f) Sample Cu2 Batch.-4 at x500 magnification.*

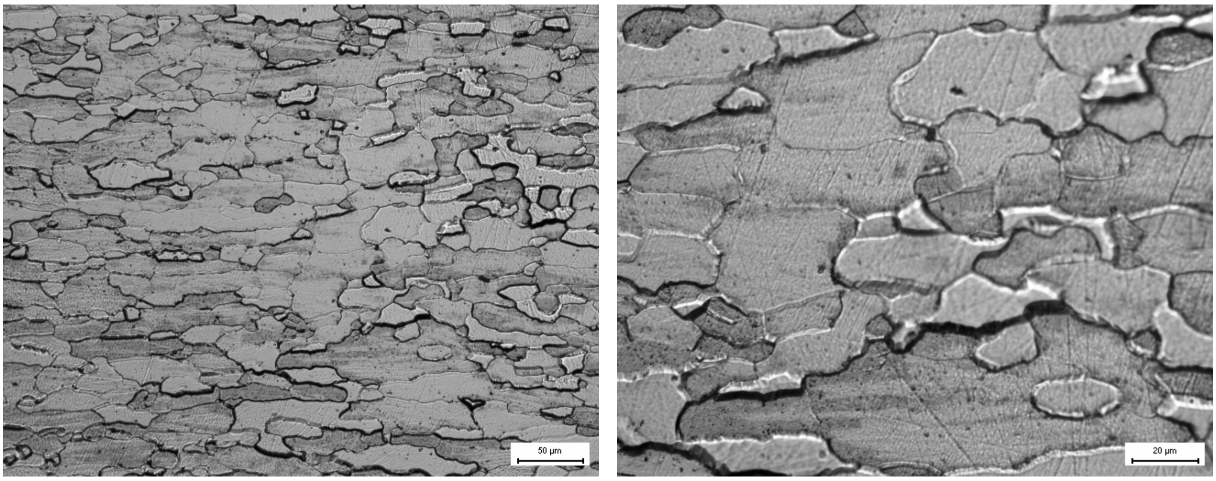
***Figure 34** Optical micrographs of the microstructure in the batch annealed alloys. The material is ferritic and the structure has been made visible by etching with 2% Nital solution (ethanol and nitric acid)*



a) Sample Cu3 Batch.-3 at x200 magnification. *b) Sample Cu3 Batch.-3 at x500 magnification.*

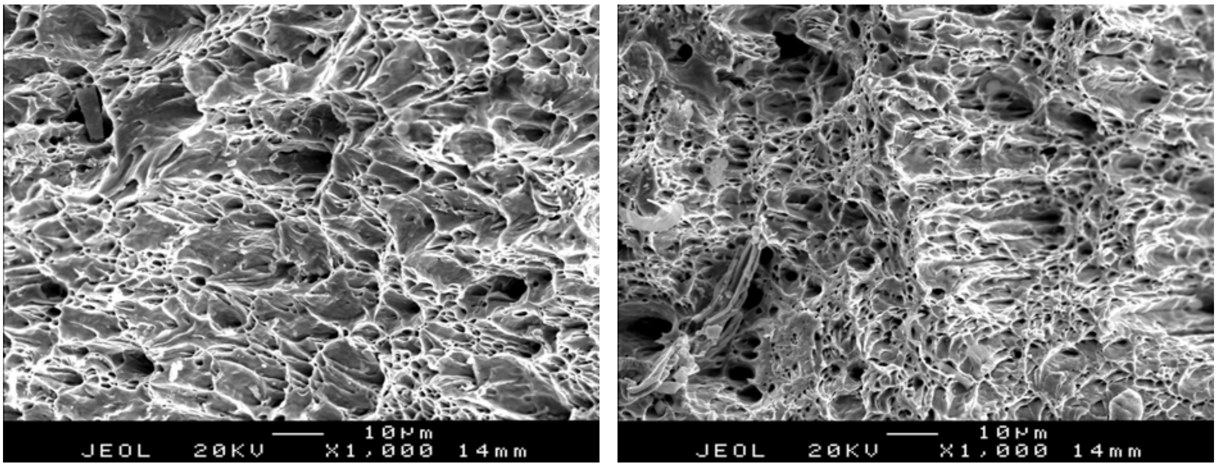


c) Sample Cu3 Ni1 Batch.-2 at x200 magnification. *d) Sample Cu3 Ni1 Batch.-2 at x500 magnification.*



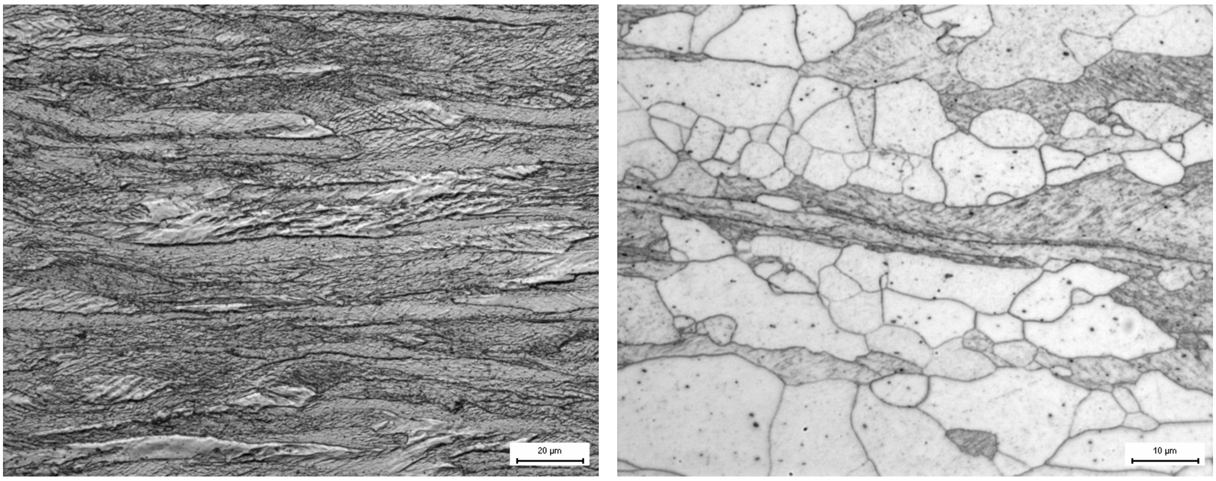
e) Sample Cu3 Ni3 Batch.-3 at x200 magnification. *f) Sample Cu3 Ni3 Batch.-3 at x500 magnification.*

Figure 35 *Optical micrographs of the microstructure in the continuously annealed alloys. The material is ferritic and the structure has been made visible by etching with 2% Nital solution (ethanol and nitric acid)*



a) Asem picture of the fracture surface of sample Cu 1 Cont.-1 x1000 magnification.

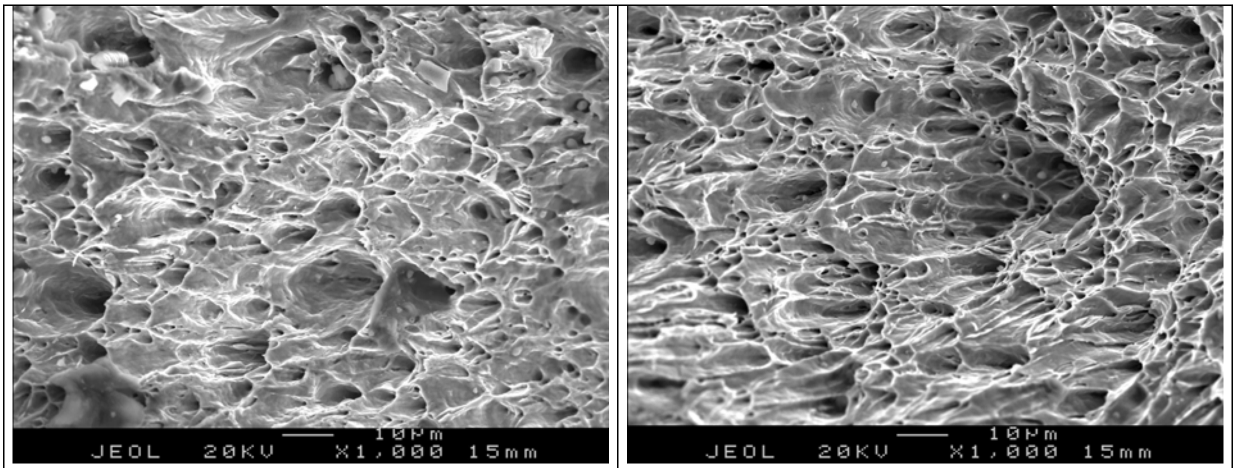
b) Asem picture of the fracture surface of Cu 1 Cont.-2 x1000 magnification.



c) Optical micrograph of sample Cu 1 Cont.-1 at x1000 magnification.

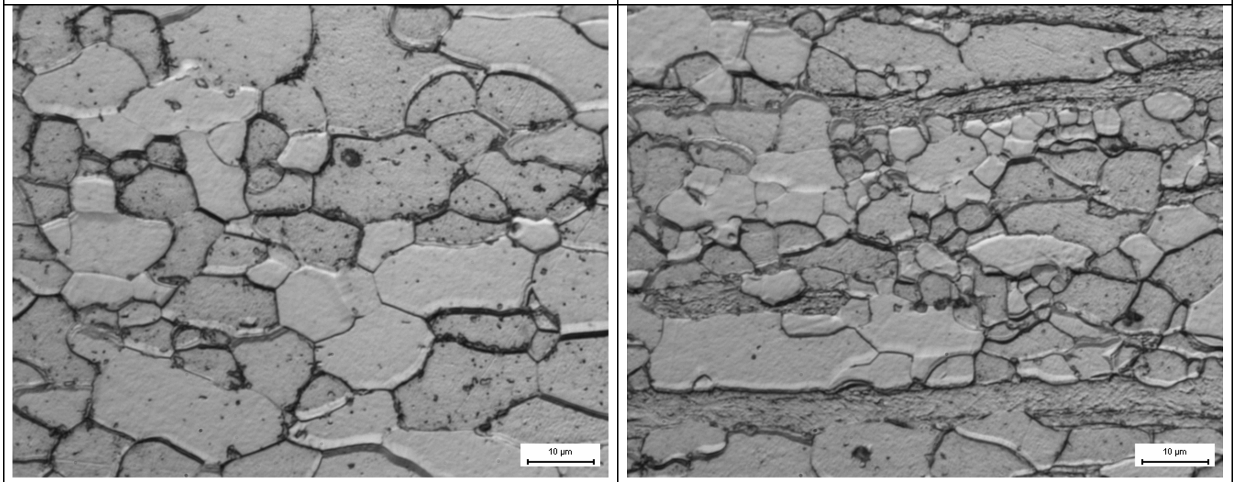
d) Optical micrographs of sample Cu 1 Cont.-2 at x1000 magnification.

***Figure 36** Electron microscope pictures of the fracture surfaces of two tensile test specimens with the light optic microscope pictures of the same material directly below. The sample on the left (Cu1 Cont.-1 a,c) has not recrystallized, which can be seen clearly on the etched microstructure. Cu1 Cont.-2 on the right has partially recrystallized.*



a) Asem picture of the fracture surface of sample Cu3 Ni3 Cont. 1 at x1000 magnification

b) Asem picture of the fracture surface of sample Cu3 Ni3 Cont.-3 at x1000 magnification.



c) Optical micrograph of sample Cu3Ni3 Cont.-1 at x1000 magnification.

d) Optical micrograph of sample Cu3Ni3 Cont.-3 at x1000 magnification.

***Figure 37** Electron micrographs of the fracture surfaces of two tensile test specimens with the light optic microscope pictures of the same material directly below. Note in c) that the material is fully recrystallized but the sample shown in d) has not fully recrystallized.*

3.3.3 Texture measurements.

The textures (**Figure 40**) are shown as orientation distribution functions (ODF) of the bcc crystal. The reason for displaying ϕ_2=°45 sections of the data is that all of the relevant data for this type of steel can be seen in this section. The numbers within the [] parenthesis, denoting crystal direction [uvw], shows the direction parallel to the rolling direction. The numbers within the (), denoting crystal planes (hkl), shows the crystal plane that is parallel to the sheet surface. The (111) crystal planes parallel to the sheet in this figure correspond to the marked line in **Figure 38** and **Figure 39**. These (111) orientations are referred to as the γ-fibre (**Figure 39**). The gamma fibre texture is desirable to have in products intended for deep drawing and cold forming as it optimises the ductility and related properties of the material. In practical terms this means that the (111) crystal planes should lie parallel to the normal plane of flat products (⊥ to both the rolling- and transverse direction) and evenly distributed across all rotations around the crystal plane normal [22].

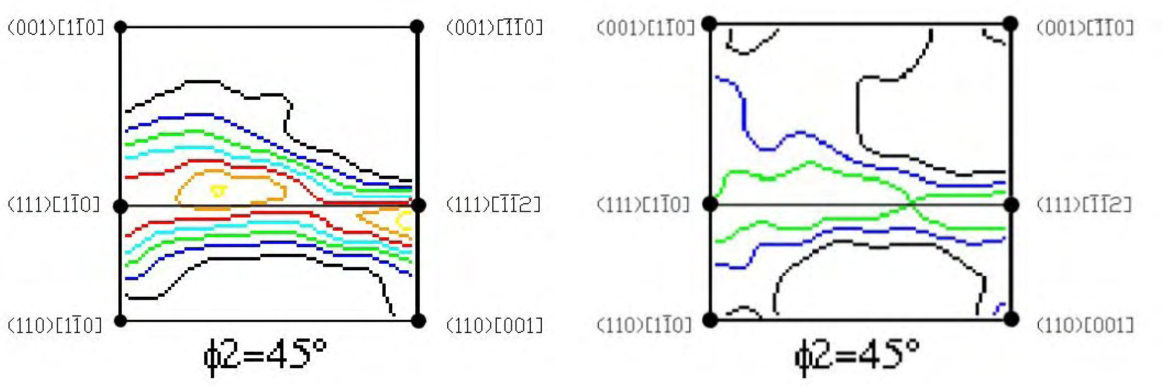


*Figure 38 γ-fibre line (**Figure 39**) superimposed over the orientation distribution functions of Cu3 Ni1 batch on the left, and over Cu3 Ni1 cont. on the right.*

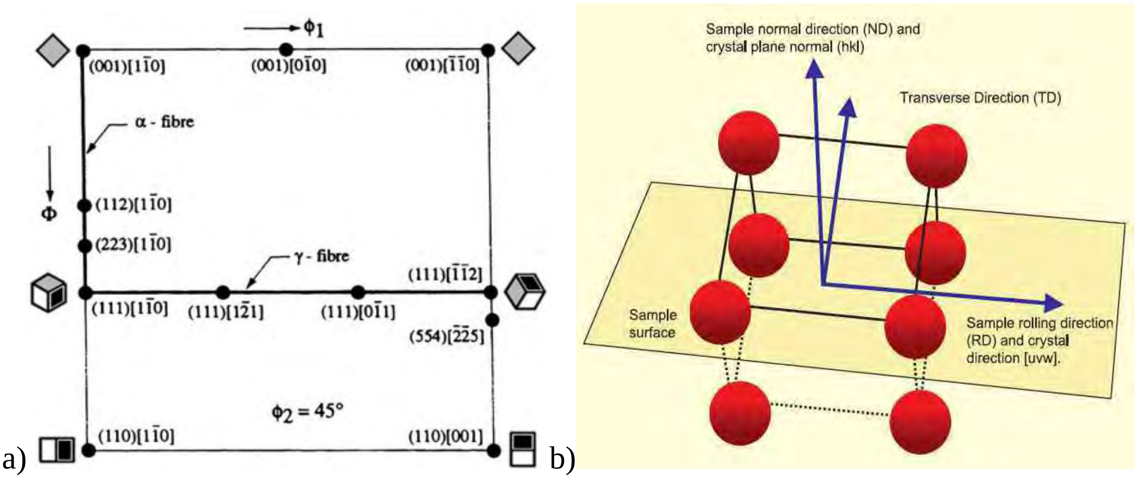
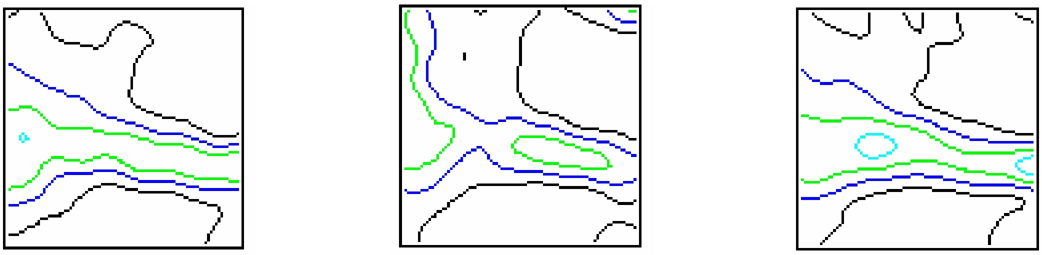


***Figure 39** a) Locations of texture components in the ϕ_2= °45 ODF figure. b) Illustration of crystal orientations in relation to sample orientation, in the case shown in the illustration the ideal crystal orientation would be (010)[001].*

The texture measurements from **Figure 40** show that:

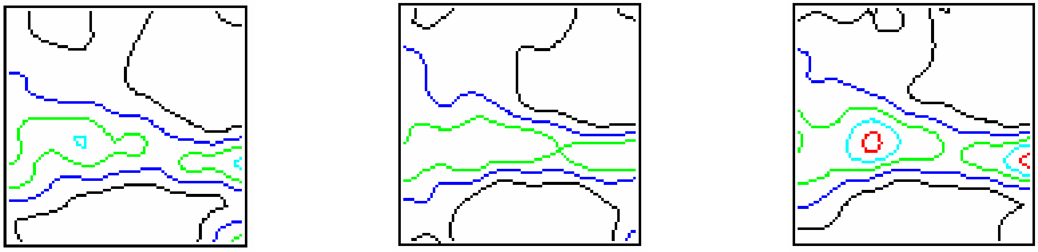
- The γ-fibre is more pronounced in the batch annealed materials.
- The γ-fibre is stronger in the Cu2, Cu3 Ni3 alloys in both batch- and continuously annealed materials (**Figure 40**).
- Cu1 cont. has not properly recrystallized, as can be seen by the area forming a laying L shape. This indicates the presence of α-fibre (**Figure 39**) and is also confirmed by the optical micrographs (**Figure 32** c, d)



ϕ2=45°
R cont.

ϕ2=45°
Cu1 cont.

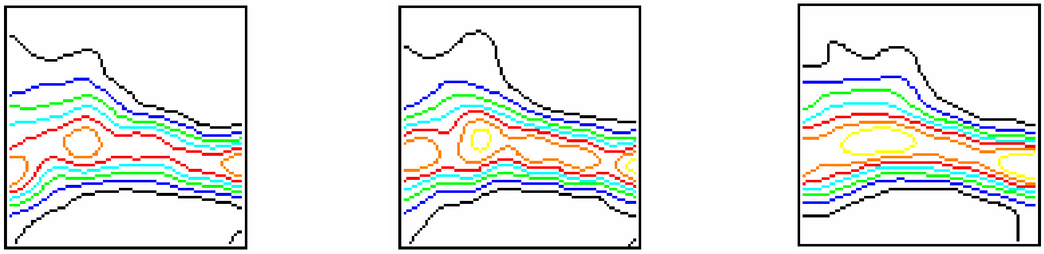
ϕ2=45°
Cu2 cont.



ϕ2=45°
Cu3 cont.

ϕ2=45°
Cu3 Ni1 cont.

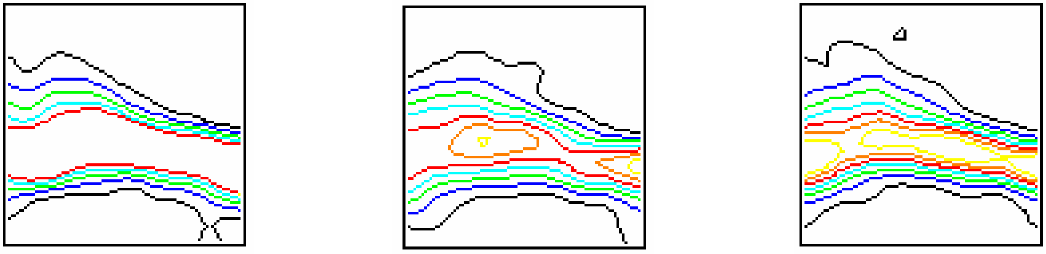
ϕ2=45°
Cu3 Ni3 cont.



ϕ2=45°
R batch.

ϕ2=45°
Cu1 batch.

ϕ2=45°
Cu2 batch.

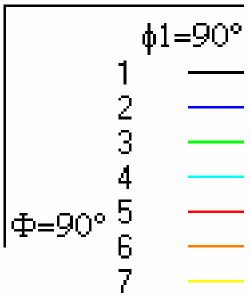


ϕ2=45°
Cu3 batch.

ϕ2=45°
Cu3 Ni1 batch.

ϕ2=45°
Cu3 Ni3 batch.

***Figure 40** Orientation distribution functions for all alloys, showing the gamma-fibre intensity.*



Intensity reference.

3.4 Discussion.

3.4.1 Tensile test results.

The continuously annealed material behaves differently from the batch annealed material in the tensile tests as can be expected. But it still has a significant Lüders strain and shows a typical stress-strain behaviour regardless of the low carbon content. The slight decrease in Lüders strain is encouraging as it at least does not indicate a negative effect from copper and nickel. As before mentioned, a Lüders strain as small as possible is desirable for materials intended for extensive cold forming.

One feature to be studied, that was noted in a previous test of the same nature as this experiment, was the disappearing of the Lüders strain in batch annealed material alloyed with copper. In this experiment however, none of the batch-annealed material showed any Lüders strain regardless of other alloy content. The tensile tests of the batch annealed material show almost interstitial free steel behaviour. This has been confirmed to be due to decarburization during the annealing process (**Table 3**). In previous internal studies done at KIMAB, Lüders strain was observed in unalloyed batch annealed material but could not be observed when alloyed with copper up to 1,5 w%.

In regards to the other mechanical properties such as ultimate tensile strength (i.e. maximum stress in the plastic deformation area) the materials shows ductile and soft behaviour. The decrease in hardness with the addition of nickel could possibly be related to the fact that nickel increases the solubility of copper in austenite. This would then have caused the copper to disperse more easily in the base matrix during melting and casting of the alloys and not precipitate and would lessen the dispersion hardening effect. The dispersion hardening effect of copper in this case though is most likely very small even in the 0,35 w% copper alloyed material, and thus affording no easy explanation of the softening effect of nickel.

The mechanical behaviour of the material shows no negative effects of copper in regards to formability of the material. The lowering of the yield strength is in this case positive since a high yield strength would require more force for deformation of the material.

3.4.2 Microstructure.

The light optical microscope pictures show a typical recrystallized ferrite structure with elongated grains in the rolling direction. The continuously annealed alloys have a smaller average grain size compared to the batch annealed ones which is consistent with heat treatments and subsequent mechanical properties. As can be seen in **Figure 32** c, d the continuously annealed Cu1 material has not properly recrystallized. The reason for this is hard to discern since the other alloy (Cu3 Ni1 cont.) that was heat treated at the same time as Cu1 cont. was properly recrystallised. The behaviour is still ductile though as is confirmed by the fracture surfaces seen in **Figure 36** and **Figure 37**. Even with material not recrystallized at all (as seen in **Figure 36** c) the fracture surface still shows the typical dimples of a ductile break. The material that failed to recrystallize even partially (**Figure 36** c), Cu 1 cont. sample 1, was discarded from the data gathered on mechanical properties etc. But it was examined in the optical microscope and in the SEM to determine the reason for its deviating behaviour.

The texture measurements show a gamma fibre in both the batch annealed material and in the continuously annealed materials. The relative intensity reference shows that the gamma fibre is nearly twice as strong in the batch annealed material as in the continuously annealed. This is also consistent with previous studies done at KIMAB. The combination of copper and nickel seems to increase the relative strength of the gamma fibre with both heat treatments. The lower level of nickel does not seem to have any significant influence on the gamma fibre but the alloy with the higher level of about 0.25 w% nickel in combination with 0.33 w% copper shows an intensity increase in the gamma fibre.

3.5 Conclusions

- The addition of copper shows no major disadvantages to the mechanical properties in regards to material intended for extensive cold forming.
- There is no discernible trend in ultimate tensile strength with increasing copper content within the tested interval.
- Elongation remains essentially constant with copper level and in combination with lower nickel levels in the continuously annealed materials.
- There is no discernible influence from copper or nickel on the average grain size in the continuously annealed materials at the composition levels within this study.
- The Lüders elongation decreases slightly with increasing copper content and with the addition of nickel in the continuously annealed material.
- The continuously annealed material shows a finer microstructure than the batch annealed, that has a more pronounced elongation of grains in the rolling direction.
- The addition of copper and nickel increases the strength of the gamma fibre texture that is favourable for deep drawing abilities of low alloyed steel.

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