

**EFFECT OF COLD DRAWING RATIO AND
COMPOSITION ON RECRYSTALLIZATION
BEHAVIOURS OF STEELS**

**M.Sc. Thesis by
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Programme: MATERIALS SCIENCE AND ENGINEERING

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JUNE 2005

**SOĞUK HADDELEME ORANI VE BİLEŞİMİN
ÇELİKLERİN YENİDEN KRİSTALLEŞME
DAVRANIŞLARINA ETKİSİ**

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FOREWORD

I would like to present my deepest regards and thanks to my supervisor Prof. Dr. Hüseyin Çimenoglu for helping me in carrying out and writing this thesis by his patience and support. I would also like to thank the research assistants Fatih M. Güçlü for helping me during the experiments and Harun Mindivan for helping me in writing the thesis. I am proud of being the member of Uyar family who encouraged me in my lifetime and existing in the iron and steel industry since five years which made me decide to prepare a thesis about steels.

This thesis is dedicated to the memory of my cousin Kerem Erzenci, a civil engineer who was graduated from İstanbul Technical University.

June, 2005

Çağdaş Uyar

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ABBREVIATIONS

FCC	: Face Centered Cubic
DIN	: Deutsches Institut für Normung
Dia.	: Diameter
Red.	: Reduction
HRB	: Rockwell B Hardness
HV	: Vickers Hardness
h	: Hours
min	: Minutes
s	: Seconds

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LIST OF SYMBOLS

T	: Temperature
T_m	: Melting temperature
t	: Time
t_r	: Time for 50% recrystallization
Cu	: Copper
Zn	: Zinc
Al	: Aluminium
Mg	: Magnesium
Fe	: Iron
Si	: Silicon
Mn	: Manganese
B	: Boron
C	: Carbon
Cr	: Chromium
S	: Sulphur
Mo	: Molybdenum
V	: Vanadium
Ni	: Nickel
σ	: Stress
ε	: True strain
τ	: Shear stress
γ	: Shear strain
b	: Burgers vector
G, ν	: Elastic constants
L	: The average distance that a dislocation moved
D	: Density of dislocations after a given deformation
l	: Distance between two slip planes
n	: Number of dislocations
P	: Distance between the slip planes
θ	: Inclination
L_S	: Slip line length
Λ	: Constant ($\approx 4 \times 10^{-4}$ cm)
ρ	: Dislocation density
α	: Constant value between 0,3 and 0,6
X_R	: Change of recovery from the annealed condition
R	: Gas constant (8,314 J/molK)
Q	: Activation energy
Q_{Rex}	: Recrystallization activation energy
X_v	: Volume fraction of the material recrystallized
NR	: Nucleation rate
GR	: Growth rate
F	: A parameter derived from <i>NR</i> and <i>GR</i>
A	: Pre-exponential constant of Arrhenius equation

SOĞUK HADDELEME ORANI VE BİLEŞİMİN ÇELİKLERİN YENİDEN KRİSTALLEŞME DAVRANIŞLARINA ETKİSİ

ÖZET

Plastik şekil verme yöntemiyle malzeme imalatı, üstün plastik deformasyon kabiliyetinden dolayı çelikler üzerinde çok yaygın olarak kullanım alanı bulan bir prosestir. Malzemeye hem sıcak hem de soğuk halde uygulanan bir plastik şekil verme yöntemi olan haddeleme tekniğinde, iki prosesin de kendine özgü avantajları ve dezavantajları mevcuttur. Sıcak haddeleme prosesinin en büyük avantajlarından birisi, malzemedede soğuk halde elde edilemeyecek kadar büyük redüksiyon (deformasyon) oranlarına olanak tanımasıdır. Ancak bu teknikte yüzey kalitesi ve ölçü toleransları kontrolünün soğuk haddeleme tekniği kadar hassas olamaması gibi bir dezavantaj söz konusudur. Soğuk haddeleme yöntemi, endüstriyel uygulamalarda genellikle sıcak haddeleme yönteminin ardından gelen bir proses olarak, hassas ölçü toleransları ve düzgün yüzey kalitesini sağlama amaçlı uygulanır. Soğuk halde deforme olmuş malzeme yapısındaki artan dislokasyon yoğunluğu mukavemet ve sertliği arttırırken, süneklik ve elektriksel iletkenliğin azalması soğuk haddeleme tekniğinin istenmeyen bir etkisidir. Bu olumsuz etkileri gidermek amacıyla soğuk haddelemenin ardından uygulanan yeniden kristalleştirme tavlama malzemedeki artan dislokasyon yoğunluğunu azalttığı gibi, deformasyondan ötürü uzamış olan tane yapısını da ortadan kaldırıp malzemedede eş eksenli bir tane yapısı meydana getirir.

Bu çalışmada, DIN 17115 standartlarına göre endüstride kaynaklı zincir üretiminde kullanılan üç adet çelik kalitesinin (RSt 35-2, 21Mn5, 20NiCrMo3) soğuk çekme işlemlerinden sonraki yeniden kristalleşme davranışları incelenmiştir. Bu tez kapsamında yapılan çalışmalarda, her üç çelikte de yeniden kristalleşme aktivasyon enerjisinin artan soğuk çekme oranı ile azaldığı, yüksek çekme oranlarında ise yeniden kristalleşme aktivasyon enerjisinin deformasyon oranından belirgin olarak etkilenmediği görülmüştür. %35'in altındaki redüksiyon oranlarında RSt 35-2'nin, %35'in üstündeki redüksiyon oranlarında ise 20NiCrMo3'ün en yüksek yeniden kristalleşme aktivasyon enerjisine sahip olduğu saptanmıştır.

EFFECT OF COLD DRAWING RATIO AND COMPOSITION ON RECRYSTALLIZATION BEHAVIOURS OF STEELS

SUMMARY

The production of materials by plastic deformation techniques is widely applied on steels due to their improved formability. Among the plastic deformation processes, rolling can be applied both in hot and cold working conditions as hot rolling or cold rolling (cold drawing). Both conditions of deformation have some advantages and disadvantages. The most important advantage of hot rolling process is application of heavy deformations to the material which is almost impossible to apply under cold working conditions. However, the control of surface quality and size tolerances is difficult during hot working processes. In the industrial applications, the cold drawing process is generally applied after hot rolling in order to reach the desired size tolerances and surface qualities. Increase of dislocation density is the main cause of hardening and strengthening of the material during cold working. However, ductility and electrical conductivity decreases upon cold working. The recrystallization annealing, which is applied after cold working not only decreases the dislocation density but also removes the elongated grain structure of the material by replacing the old grains by new equiaxed grains.

In this study, the recrystallization behaviours of three steel qualities (RSt 35-2, 21Mn5, 20NiCrMo3) which are widely used for welded chain production according to DIN 17115 are examined. Experimental studies have shown that, the recrystallization activation energy of each material decreases by increasing cold drawing ratio as a general trend. However at extended cold drawing ratios, activation energy stays almost constant with respect to applied deformations. Below the drawing ratios of 35%, RSt 35-2 quality steel is found to have the highest recrystallization energy among the examined steels while 20NiCrMo3 steel quality exhibited the highest recrystallization activation energy above this critical deformation ratio.

1. INTRODUCTION

Steel, which is the most widely used metal in the world, can be defined as an alloy of iron containing various amounts of carbon and some other elements, such as manganese, sulfur, nickel, silicon, phosphorus, chromium, molybdenum and vanadium. These elements, when combined with iron, form different types of steels with varying properties over a wide range [1]. Iron is an element that has an history more than 2000 years. Although the iron oxide compounds are known since that time, producing steel from iron oxides by smelting was a problem due to its high costs until the invention of Henry Bessemer, who first produced large amounts of cheap steel by the year of 1856. Since that time, the production of steel has developed itself as being the most important metal in the industry today [2].

Plastic deformation, which is defined as deformation that is permanent or nonrecoverable after release of the applied load is one of the methods which is widely used on steels for producing several specific shapes [3]. In the industry, the plastic deformation is classified as cold working and hot working according to the working temperature. Cold deformation method is applied to steels and also to other several metals to have perfect size tolerance and smooth surface. However after the cold deformation process, some mechanical and physical properties of the metal change in a non desirable way such as increasing dislocation density, decreasing ductility and decreasing electrical conductivity. In order to remove these non desirable properties, an annealing is usually applied to turn the cold deformed metal to its prior mechanical and physical properties. Such annealing processes are examined under the terms of recrystallization [4].

The annealing of deformed metals has an history back to the year of 1829, when French physicist Felix Savart founded that plastic deformation changes the anisotropy of metal and there becomes a structural change on the metal by subsequent annealing. By the year of 1887, the term recrystallization is first named by Sorby who studied the elongated grains in deformed iron in the same year and noted that by heating, a new equiaxed grain structure was produced. It has been

first understood in the early years of the 20th century that the stored deformation energy provides the driving force for recrystallization [5].

In this study, the recrystallization behaviours of three steel qualities (RSt 35-2, 21Mn5 and 20NiCrMo3) are examined. The aim of this study is to determine the effect of cold deformation ratio and chemical composition on the recrystallization activation energy of steels.

2. FUNDAMENTALS OF PLASTIC DEFORMATION

2.1. Introduction

Deformation, which is occurred on a metal by subjecting it to an uniaxial tensile force is mainly examined in two categories. If the metal returns to its original dimensions after the removal of this applied force, the type of deformation is said to be elastic deformation. If the metal is not able to fully return to its original dimensions as it is deformed to such an extent, the type of deformation is said to be plastic deformation. During plastic deformation, the metal atoms are displaced from their original positions translating to new positions. Having the ability to be extensively plastically deformed without fracture makes a metal extremely useful on engineering applications. For instance, the automobile parts such as fenders, hoods and doors which are made from steel can be stamped out mechanically without any metal fracturing thanks to the plastic deformability of steel [6].

When the type of forces applied to the workpiece is considered, the plastic deformation processes can be classified in 5 categories [4].

1. Direct compression-type processes
2. Indirect compression-type processes
3. Tension type processes
4. Bending Processes
5. Shearing Processes

In the direct compression processes, the flow of metal takes place at right angles to the direction of the compression as the force is applied to the surface of the workpiece. Forging and rolling are two main examples that can be given for this type of processes. In the indirect compression processes, the primary forces applied on the workpiece are generally as tensile forces but the reaction of the workpiece with the die develops indirect compressive forces which reach high values. Therefore the flow of the metal takes place under the action of a combined stress state which includes high compressive forces in at least one of the principal directions. Wire drawing and tube drawing, extrusion and the deep drawing of a cup are the

examples for indirect compression processes. For the tension-type forming processes, stretch forming is a good example, where a metal sheet which is under the application of tensile forces is wrapped to the contour of a die. Bending is a type of forming process where bending moments are applied to a sheet, while shearing is the application of shearing forces which have enough magnitude to rupture the metal in the plane of shear. These processes are illustrated in Figure 2.1 in a simplified way [4].

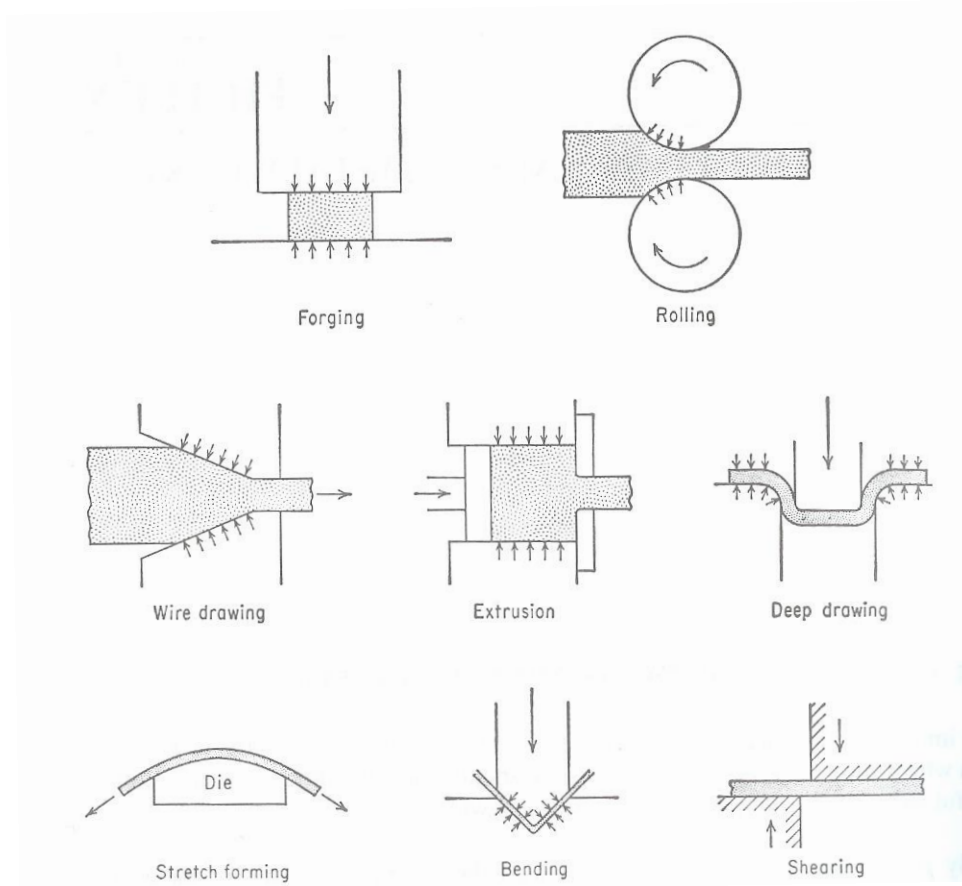


Figure 2.1: The plastic deformation processes [4].

To analyse and understand these processes better, a different distinction can also be made by classifying them as steady state processes and non steady state processes [7].

In steady state processes, all parts of the workpiece are undergone to the same type of deformation. Thus, once the analysis of the situation is made for the deformation zone, it remains valid for the duration of the process.

In non steady state processes, the analysis must be repeated severally for different points in time from the starting condition to the end of the stroke of the deforming machine because the geometry of the part changes continually. There are also some processes which show a transitional character. For instance, in the extrusion process, deformation is non steady state at the beginning and at the end of the process, but it shows steady state characteristics during the extrusion of the greater part of a billet. This classification is illustrated in Figure 2.2 [7].

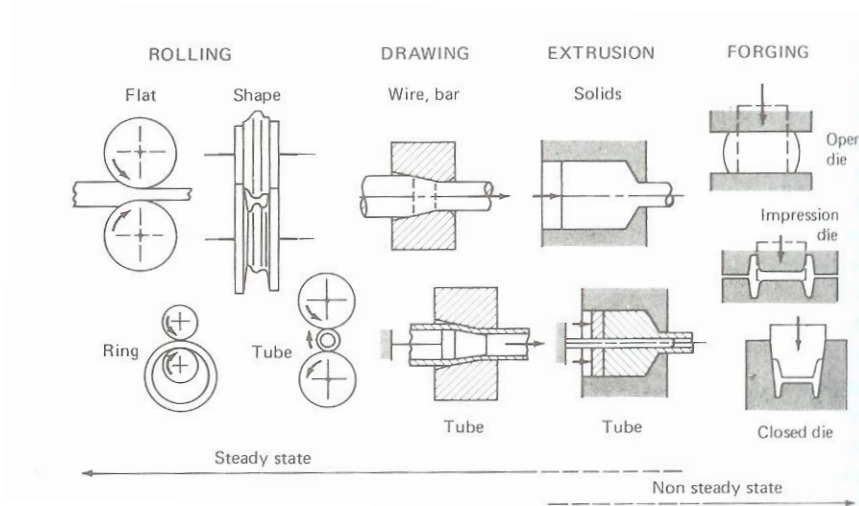


Figure 2.2: The plastic deformation processes classified as steady state and non steady state [7].

2.2. Classification of Plastic Deformation Processes

Depending on the temperature of deformation, forming processes can be classified as “hot working” and “cold working” operations [4].

2.2.1. Hot Working

In hot working, deformation takes place under conditions of temperature and strain rate where recovery processes take place at the same time with the deformation processes so that large strain amounts can be reached without any strain hardening [4]. The temperatures above $0.5T_m$ make the diffusion of atoms considerably easier. Thus, an arrested dislocation has the chance of climbing and moving into another, unobstructed atomic plane. Therefore, many dislocations can rapidly disappear if deformation takes place at the temperatures above $0.5T_m$. As a matter of fact, it can be said that softening processes work simultaneously with dislocation propagation.

Material which is produced by such hot working processes has much lower dislocation density and therefore, is less strain hardened than a cold-worked one [7].

$0,5T_m$ is also approximately the temperature of recrystallization. Thus, we can also define hot working as a process which is carried out above the recrystallization temperature. The definition dynamic recrystallization is referred to the type of recrystallization that takes place during hot working. In many materials dynamic recovery takes place during working, resulting in quite low flow stresses. Recrystallization may still occur as static recrystallization on holding at or cooling from the hot-working temperature. Therefore, the distinctive mark of hot working is not a recrystallized structure but the simultaneous occurrence of dislocation propagation and softening processes, with or without recrystallization during working. The dominant mechanism depends on temperature, strain rate and grain size, and may be conveniently shown on deformation mechanism maps. In general, the recrystallized structure becomes finer with lower deformation temperature and faster cooling rates, and material of superior properties is often obtained by controlling the finishing temperature [7].

Hot working processes such as rolling, extrusion or forging are typically used in the first step of converting a cast ingot into a wrought product. The advantages of hot working can be sorted as below [4]:

- The energy which is required to deform the metal decreases.
- The rapid diffusion at hot working temperatures decreases the chemical inhomogeneities of the cast ingot structure.
- The blowholes and porosity are dissipated by the welding together of these cavities.
- The coarse columnar grains of the casting are broken down and they refine into smaller equiaxed recrystallized grains.
- The changes in structure from hot working result in an increase in ductility and toughness over the cast state.

However, there are also some disadvantages of hot working [4]:

- Due to high temperature, the surface of the metal oxidizes as hot working is done in air and this means a loss of considerable amount of the metal. Oxidation also makes it difficult to produce good surface finishes .
- The dimensional tolerances for hot worked mill products are greater than the ones produced by cold working.

- The structure and properties of hot worked metals are generally not so uniform over the cross section as in metals which have been cold worked and annealed.
- The recrystallized grain size will be finer in the surface layer since the deformation is always greater in the surface layers. Also, grain growth can occur in the interior of large pieces because the interior will be at higher temperatures than the external surfaces for longer times during cooling.

2.2.2. Cold Working

In cold working conditions, the recovery processes are not effective while the deformation is being carried out. Cold working of a metal increases its strength and hardness while the ductility of the metal is decreased. If cold working is excessive, the metal will fracture before reaching the desired size and shape. In order to soften the cold worked metal and restore the ductility, cold working operations are usually done in several steps, with intermediate annealing operations and the possible fractures that might take place are avoided by this way. It is obvious that the need for annealing operations increases the cost of forming by cold working, especially for reactive metals which must be annealed in vacuum or inert atmospheres. However cold working provides a degree of versatility which is not possible in hot working operations. If the cold work-anneal cycle is adjusted conveniently, the part can be produced by achieving the desired degree of strain hardening. If a higher strength is desired on the finished product according to the fully annealed material, then a cold working must be applied as a final step with the proper deformation amount in order to produce the desired strength. This would probably be followed by a stress relief to remove residual stresses. Such a method to obtain a certain combination of strength and ductility in the final product is more successful than trying to achieve the same combinations of properties by partially softening a fully cold worked material, because the recrystallization takes place so quickly and is quite sensitive to small temperature fluctuations in the furnace. If it is desired to obtain the material in fully softened conditions as conclusion, then an annealing step must follow the final cold working step [4]. The other effects of cold working can be sorted as [4];

- Cold working produces elongation of the grains in the principal direction of working. The grains are sheared relative to each other by the generation, movement and rearrangement of dislocations.
- The surface obtained in cold working is smoother than hot working.
- The dimensional tolerances are smaller than hot working.

The dislocation density increases with increasing cold deformation. The exact mechanism of increasing dislocation density by cold working is not completely understood. New dislocations are created by the cold deformation and must interact with those already existing. As the dislocation density increases with deformation, it becomes more and more difficult for the dislocations to move through the existing “force of dislocations,” and thus the metal strain hardens with increasing cold deformation [6].

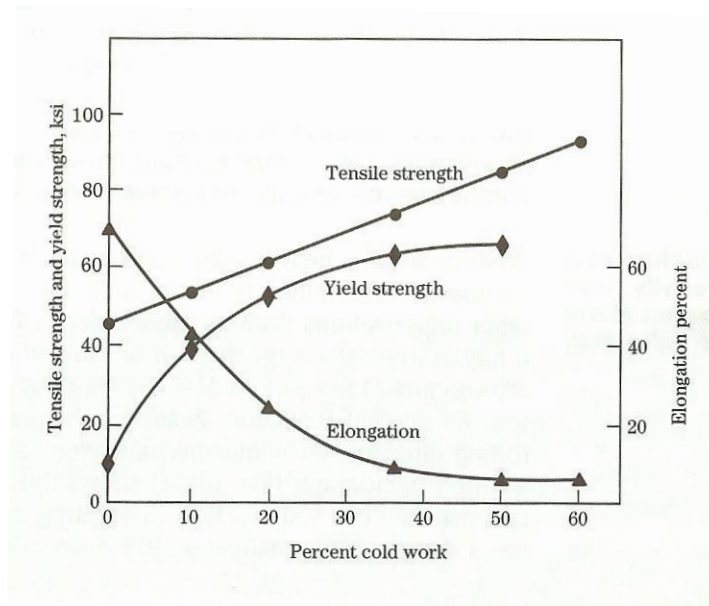


Figure 2.3: The change on the properties of 70% Cu – 30% Zn alloy with increasing cold work amount [6].

In ductile metals such as copper, aluminum and α iron which have been annealed and cold worked at room temperature, strain hardening takes place because of the dislocation interaction. Figure 2.3 shows how cold working at room temperature increases the tensile strength of 70%Cu – 30% Zn alloy from about 30 ksi (200MPa) to 45 ksi (320 MPa) with 30% cold work. Associated with the increase in tensile strength, however, is a decrease in elongation (ductility), as observed in Figure 2.3 with 30% cold work, the elongation of unalloyed copper decreases from about 42% to 10% elongation [6].

The rate of strain hardening can be gaged from the slope of the true stress-strain curve. Generally, the rate of strain hardening is lower for hexagonal closed packed metals than for cubic metals. Increasing temperature also lowers the rate of strain hardening. For alloys strengthened by solid-solution additions, the rate of strain hardening may be either increased or decreased compared with the behaviour for

the pure metal. However, the final strength of a cold worked solid solution alloy is almost greater than that of the pure cold worked to the same extent. Cold working also shows some other changes in physical properties in addition to the changes in tensile properties shown in Figure 2.3. There is usually a small decrease in density in order of a few tenths of a percent, an appreciable decrease in electrical conductivity due to an increased number of scattering centers, and a small increase in the thermal coefficient of expansion. Because of the increased internal energy of the cold worked state, chemical reactivity is increased. This leads to a general decrease in corrosion resistance and in certain alloys introduces the possibility of stress-corrosion cracking [4].

2.3. Theories of Work Hardening

There have been many theories in order to explain the phenomenon of work hardening up today. Determination of how the density and distribution of dislocations vary with plastic strain is the most difficult and also the most important part of these attempts to predict the work hardening behaviour. In other words, it is impossible to understand how a certain amount of strain was accumulated in the crystal with the aid of presence or absence of dislocations because we have no idea about the path that dislocations traversed to obtain that strain. Both the density and the distribution of dislocations are very sensitive functions of the crystal structure, stacking fault energy, temperature and rate of deformation. Due to all these reasons, a unique theory of work hardening that would explain all the aspects does not exist. In fact, Cottrell has observed that work hardening was the first problem that dislocation theory tried to solve and will be the last one to be solved [8].

2.3.1. Early Theories

The earliest theory of work hardening which involved dislocations was put forward by Taylor in 1934. At that time the stress-strain curves of metal crystals such as aluminium were considered to be parabolic, so Taylor created a dislocation model from which such curves could be calculated [8,9]. He appreciated that many dislocations do not reach the surface of a crystal but interact elastically with other dislocations, and form a network within the crystal. This dislocation model can be seen in Figure 2.4 [9].

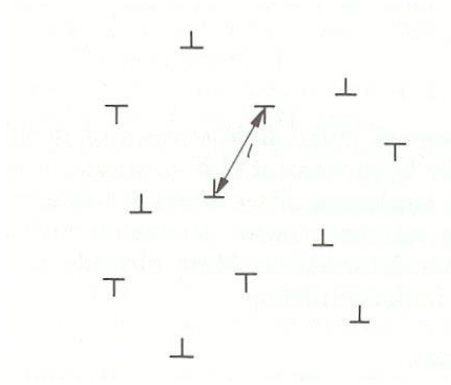


Figure 2.4: Dislocation interaction – Taylor model [9].

This is a progressive state of affairs so the dislocation concentration gradually increases as deformation proceeds, and the stress necessary to force a passage for the later dislocations is raised [9].

Taylor assumed that the average distance a dislocation moved was L before it was stopped. If the density of dislocations after a given deformation is D then the strain ϵ is given by [9];

$$\epsilon = D \cdot L \cdot b \quad (2.1.)$$

where b is the Burgers vector.

The spacing of the dislocations (l) will be $1/D^{1/2}$ and they will interact elastically with their neighbours. The effective internal stress τ as a result of these interactions is the stress just necessary to force the two dislocations past each other and given by [9];

$$\tau = Gb / 8\pi(1 - \nu)l \quad (2.2.)$$

where l is the distance between the two slip planes. G and ν are the usual elastic constants [9].

$$\tau = k \cdot (Gb / l) = k \cdot G \cdot b \cdot D^{1/2} \quad (2.3.)$$

where k is a constant. Substituting for D from equation 2.1. [9];

$$\tau = kG(\epsilon b / L)^{1/2} \quad (2.4.)$$

This is a parabolic relationship between the stress τ and strain ϵ which approximately describes the behaviour of many metals.

However, Taylor's theory is not suitable with experiments as the deformation does not take place by movement of isolated dislocations, but as a result of numerous dislocations on slip bands arising from sources. Another objection is that pairs of dislocations while locked with respect to each other, can be forced to move by other dislocations [9].

After Taylor's theory, Mott came up with his theory by replacing individual dislocation interactions by interactions between piled-up groups of dislocations as the sources of internal stress (Figure 2.5) [8,9].

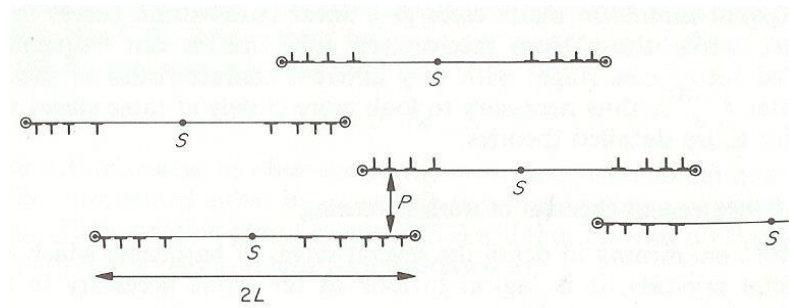


Figure 2.5: Interaction of piled up groups of dislocations on the primary system [9].

In Mott's theory, each dislocation pile-up is considered to be a superdislocation of Burgers vector with a strength of nb where n is the number of dislocations, and a stress at the head of the pile-up n times the applied stress τ . As a result, the distance between pile-ups can be greater than between the dislocations in Taylor's model, and can be of the order of the slip band spacing. If it is assumed that dislocations originating from one source and piled up at each side, the separation distance is said to be $2L$ (each section of dislocation line moves a distance L) and the distance between the slip planes is P , the density of piled up groups is $1/(L \cdot P)$ and the average distance between them is $(L \cdot P)^{1/2}$ [9].

The mean stress τ acting on each pile-up is given by [9];

$$\tau = Gnb / 2\pi(L \cdot P)^{1/2} \quad (2.5.)$$

The plastic strain ϵ is simply the summation of the strains from each pile-up which is the product of $nb \cdot L$ and the pile up density $1/LP$ so [9];

$$\epsilon = nb / P \quad (2.6.)$$

Combining the equations 2.5. and 2.6. [9];

$$\tau = G / 2\pi (\epsilon nb / L)^{1/2} \quad (2.7.)$$

This theory also gives a parabolic relationship between stress and strain for single crystals. However, experimental work since 1950 has shown that most single crystals do not have parabolic stress-strain curves to high strains, especially those from face-centered cubic metals can frequently be divided into three stages with very different characteristics. Thus, the recent detailed theories are developed in this way. [8,9].

2.3.2 Recent Theories

Before the examination of the several stages of hardening which occur in metal crystals, it is logical to look at the stress necessary to make dislocations move in the first instance, i.e. the initial flow stress. Seeger has pointed out that the initial flow stress will be determined not only by the interactions between the first dislocations to be generated on the chosen slip system, but also by the interaction of these dislocations with those which are present in the annealed state. These “grown-in” dislocations can be assumed to be randomly distributed, many of them cutting across the primary slip planes; these are often referred to as the dislocation forest. An idealized model is shown in Figure 2.6, where glide dislocations in two neighbouring slip bands are present as well as several forest dislocations [9].

The common shear stress-strain curve for face-centered cubic metal single crystals is illustrated in Figure 2.7. The three regions on the curve are divided as I,II,III and θ_I , θ_{II} , θ_{III} as the respective inclinations ($d\tau/d\gamma$) of these regions. These three stages which have salient points are explained with separate theories [8].

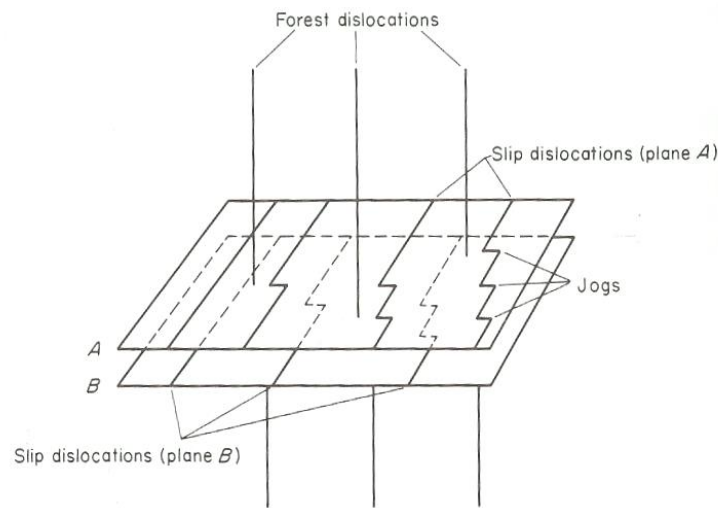


Figure 2.6: Interaction of glide dislocations on plane A with, **a**, other glide dislocations on plane B; **b**, forest dislocations [9].

The stage I starts after elastic deformation at the critical stress τ_0 . This stage, called “easy glide” is a linear region of flow strain hardening rate. θ_I is approximately one-tenth of θ_{II} . This stage is characterized by long slip lines (100 to 1000 μm), straight and uniformly spaced (10 to 100 nm). Stage I does not exist in polycrystals. The extent of stage depends strongly on the crystal orientation. The strain at the end of stage I (γ_2) has a maximum value when the crystal orientation is located in the center of the standard stereographic triangle. The end of stage I is considered to be the start of secondary slip [8].

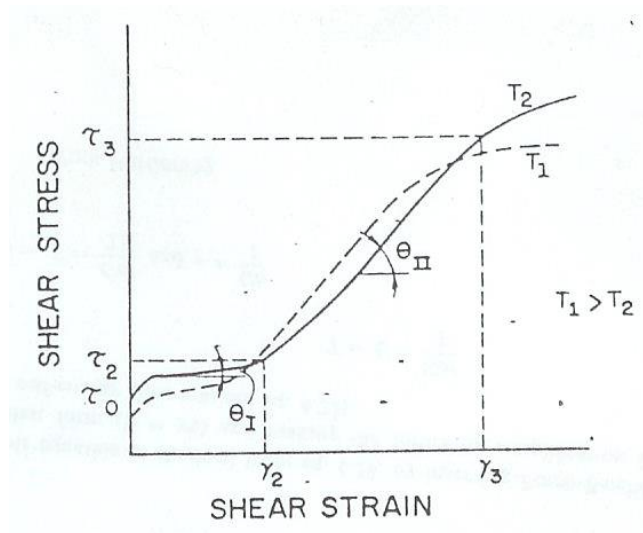


Figure 2.7: Generic shear stress-shear strain curves for FCC single crystals for two different temperatures [8].

The stage II or linear hardening stage has the following important characteristics [8].

1. A linear hardening regime with high θ_{II} .
2. $\theta_{II}/G \approx 1/300$. This parameter is relatively constant for a great majority of metals (maximum variation being by a factor of about 2). θ_{II} approximately equal to $10\theta_I$ and is, relatively, independent of temperature, although it has a significant effect on the extent stage II.
3. Rapid and linear hardening is associated with the secondary slip system activity, although a great part of deformation occurs by slip on the primary system.
4. Experimental observations on dislocation distributions, such as transmission electron microscopy, X-ray topography, or chemical etch pitting indicate that the arrangement of dislocations is very heterogenous, a majority of dislocations being concentrated in small regions and separated by regions of low dislocation density. It has also been observed that the secondary dislocation density in the single crystal is comparable to the primary dislocation density throughout the stage II, in spite of the fact that the macroscopic plastic deformation on secondary systems is very small.
5. Although the primary system is more active, the secondary systems are also operating. Dislocations on active systems interact with those on other systems and form barriers to the continuing movement of dislocations. The barriers increase with increasing deformation and the slip lines on the surface become ever shorter, closely and not so regularly spaced. It has been shown with transmission electron microscopy that the slip-line length, L_S is inversely related to the shear strain in stage II, that is, $L_S = \Lambda / (\gamma - \gamma_2)$ where Λ is constant ($\approx 4 \times 10^{-4}$ cm), γ is the total shear strain and γ_2 is the shear strain at the start of stage II.
6. The flow stress τ is proportional to the square root of the dislocation density ρ . If τ_0 is the stress necessary for moving a dislocation in the absence of other dislocations, then [8]

$$\tau = \tau_0 + \alpha G b \rho^{1/2} \quad (2.8.)$$

where α is a constant with a value between 0,3 and 0,6. This relationship has been observed to be valid for a majority of the cases.

The stage III starts at the point (γ_3, τ_3) and the hardening rate decreases continuously, the change in stress can be described by the equation [8];

$$\tau = \theta_{III}(\gamma - \gamma')^{1/2} \quad (2.9.)$$

where γ' is a constant. This third stage of hardening is parabolic. The stress τ_3 decreases exponentially with increase in temperature and the θ_{III} shows a similar tendency. This stage is characterized by the presence of wavy slip lines as the slip is not restricted to one unique plane [8].

Stacking fault energy, which is a function of the metal that determines the extent to which unit dislocations dissociate into partial dislocations[5], is an important factor that affects the work hardening behaviour of metals. The low stacking fault energy metals show all three work hardening stages at room temperature. The high stacking fault energy metals show only stages II and III. For example, aluminum (a high stacking fault energy metal) shows, at room temperature and above, a poorly developed stage II and stage III. This metal must be tested at 77 K to see all three stages. On the other hand, copper shows all three stages at room temperature. This difference between copper and aluminum is due to their different stacking fault energies which affect the width of extended dislocations in them [8].

The flow stress of a metal is affected by a change of temperature or strain rate. It is suitable to think of the flow (shear) stress (τ) as consisting of two parts [8]:

$$\tau = \tau^* + \tau_G \quad (2.10.)$$

where τ^* is the temperature-dependent part of flow stress while τ_G is the temperature-independent part of flow stress. τ_G depends on temperature only through variation of shear modulus with temperature. This relative importance of τ^* and τ_G can be studied by measuring the flow stress dependence on temperature or strain rate. (i.e., by measuring the change in flow stress on changing the temperature or strain rate.)

The ratio τ^* / τ_G is observed to be approximately unity in stage I. In stage II, this ratio decreases and reaches a constant value of about 0.1. This constancy of the τ^* / τ_G ratio is often referred to as the Cottrell – Stokes law [8].

3. RECRYSTALLIZATION

3.1. Introduction

The various processes concerning the annealing of deformed metals are examined under the general terms of recovery and recrystallization. The term recovery refers to the processes where new grains do not replace the deformed grains, but nevertheless lead to structural changes on a fine scale within the existing grains. On the other hand, the definition of recrystallization is easier because at the end of this process, the old grains are absorbed by new equiaxed strain free grains. In recrystallization, the mechanical and physical properties of the deformed metal completely return completely to those of the annealed state. The mechanical properties such as hardness, yield strength and ultimate tensile strength change slowly during the recovery period, but during recrystallization these properties alter rapidly over a very small temperature range. Likewise, the elongation of grains sharply increases to become typical of the annealed material. Physical properties such as electrical resistivity and density, while they show minor changes during recovery, also change drastically during recrystallization [9]. Recovery and recrystallization are competing processes as both are driven by the stored energy of the deformed state. Once recrystallization has occurred and the deformation substructure has been consumed, then clearly no further recovery can occur. As the recovery stage plays an important role in nucleating recrystallization, the division between recovery and recrystallization is sometimes difficult to define. There are circumstances in which there is no clear distinction between these two phenomena [5].

3.2. Stages of Recrystallization Annealing

The exact mechanism of recrystallization is a nucleation and growth process. It develops by growing of unstrained nuclei in the deformed metal when the temperature is high enough, and gradually absorbing the whole of the deformed

matrix [9]. In Figure 3.1, the evolution of nucleation and growth of the recrystallized grains of an aluminum alloy containing 0,8% magnesium is illustrated [6].

Primary recrystallization takes place by two principal mechanisms: (1) an isolated nucleus can expand within a deformed grain or (2) an original high angle grain boundary can migrate into a more highly deformed region of the metal. In either case, the structure on the concave side of the moving boundary is strain free and has a relatively low internal energy, whereas the structure on the convex side of the moving interface is highly strained with a high dislocation density and high internal energy. Grain boundary movement is therefore away from boundary's center of curvature. Thus, the growth of an expanding new grain during primary recrystallization leads to an overall decrease in the internal energy of the deformed regions with strain free regions [6].

The rules of recrystallization, which are the statements constituted after the results of a large body of experimental work, explain the effect of the initial microstructure (grain size), and processing parameters (deformation strain and annealing temperature), on the time for recrystallization and on the grain size after recrystallization. If the recrystallization process is considered to be a nucleation and growth phenomenon which is controlled by thermally activated processes and whose driving force is supplied by the stored energy of deformation, then these rules are obeyed in most cases and are easily rationalised [5].

- i. **A minimum deformation is needed to initiate recrystallization.** The deformation must be enough to begin a nucleation for the recrystallization and to supply the necessary driving force to sustain its growth.
- ii. **The temperature at which recrystallization occurs decreases as the time of annealing increases.** This statement is because of the microscopic mechanisms controlling the recrystallization which are thermally activated and the relationship between the recrystallization rate and the temperature which is given by the Arrhenius equation.
- iii. **The temperature at which recrystallization occurs decreases as strain increases.** The driving force for recrystallization is supplied by the stored energy of deformation and it increases with strain. Thus, in this case nucleation and grain growth are more rapid or occur at a lower temperature in a more highly deformed material.

- iv. **The recrystallized grain size depends primarily on the amount of deformation, being smaller for large amounts of deformation.** The strain influences the nucleation rate more than the grow rate. Therefore a higher strain is able to supply more nuclei per unit volume and thus a smaller final grain size.
- v. **For a given amount of deformation the recrystallization temperature will be increased by:**

A larger starting grain size. For the initiation of nucleation, grain boundaries are favoured sites, thus larger initial grain sizes supply fewer nucleation sites, the nucleation rate is lowered, and recrystallization is slower or occurs at higher temperatures.

A higher deformation temperature. If the deformation process is done at higher temperatures, more recovery occurs during the deformation (dynamic recovery). Therefore, the stored energy is lower than for a similar strain at a lower temperature

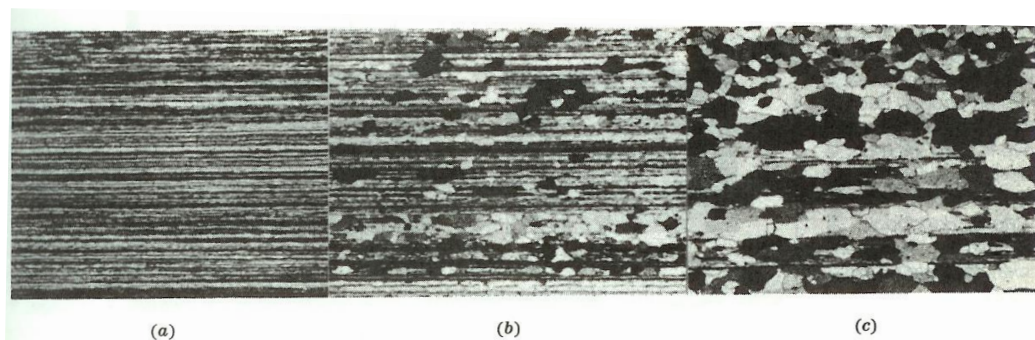


Figure 3.1: Recrystallization evolution of a 85% cold rolled Al alloy containing 0,8% Mg [6].

As the metal structure is recrystallized by the annealing treatment, the tensile strength of a cold worked metal is greatly decreased and likewise its ductility is increased. For example, the tensile strength of a 1 mm sheet of 85% Cu – 15% Zn brass which had been cold rolled to 50 percent reduction was decreased from 520 to 200 MPa by annealing 1 h at 400°C. The ductility of the sheet, on the other hand, was increased from 3 to 38 percent with the annealing treatment [6].

The effect of the annealing on the structural and the mechanical properties of a cold worked metal is illustrated in Figure 3.2 [6]. Recrystallization curves are composed of three distinct regions named as Recovery, Recrystallization and Grain Growth.

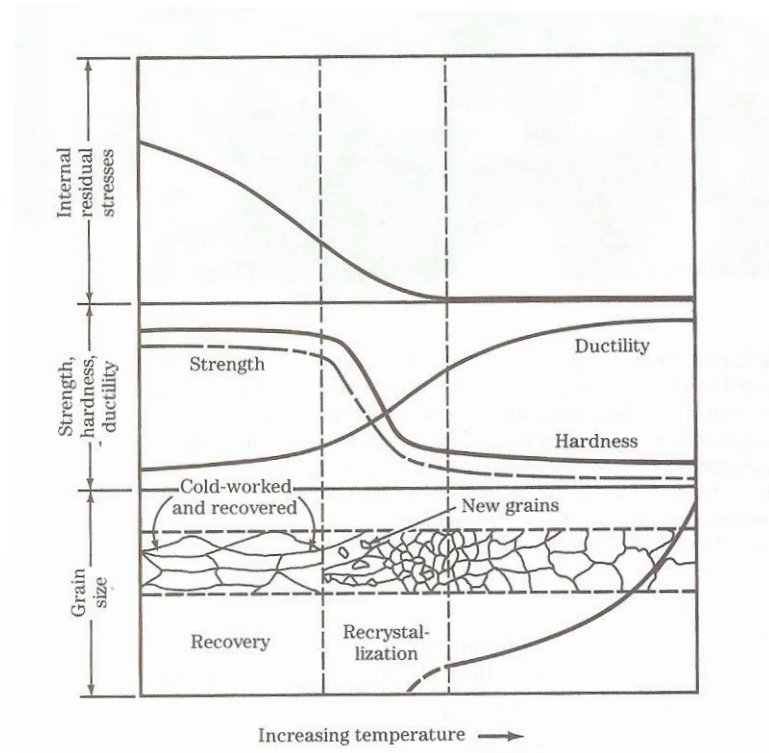


Figure 3.2: The effect of annealing on the structural and the mechanical properties of a cold worked metal [6].

3.2.1. Recovery Region

Dislocation recovery is not a single microstructure process but a series of micromechanisms as (a) cell formation (b) annihilation of dislocations with cells (c) subgrain formation (d) subgrain growth. Whether any or all of these occur during the annealing of a particular specimen will depend on a number of parameters, including the material, purity, strain, deformation temperature and annealing temperature. In many cases some of these stages will have occurred during the deformation as dynamic recovery. Although the recovery stages tend to occur in the order mentioned above, there may be significant overlap between them [5].

As the microstructural changes in a material is subtle and occur on a small scale during recovery, the microstructures as observed by optical microscopy do not usually reveal much change and thus, recovery is often measured indirectly by some bulk technique like following the changes in some physical or mechanical properties. Also, measurement of the changes in stored energy by calorimetry is the most direct method of following recovery as the stored energy is directly related to the number and configuration of the dislocations in the material [5].

For polycrystalline metals, complete recovery can occur if the material has been lightly deformed. However, single crystals of hexagonal close-packed metals such as zinc, which can be deformed to large strains on only one slip system, are able to completely recover their original microstructure and properties on annealing. Single crystals of cubic metals if oriented for single slip and deformed in stage I of work hardening may also recover almost completely on annealing. However, if the crystals are deformed into stages II or III of work hardening, then recrystallization may intervene before any significant amount of recovery has occurred. Another factor affecting recovery is the annealing temperature as it is indicated that complete recovery occurs at higher temperatures [5].

One of the most important parameters affecting recovery is stacking fault energy. This parameter determines the rate of dislocation climb and cross slip by affecting the extent to which dislocations dissociate. In metals of low stacking fault energy such as copper, α -brass and austenitic stainless steel, climb is difficult and little recovery of the dislocation structure normally occurs prior to recrystallization. However, in metals of high stacking fault energy such as aluminium and α -iron, climb is rapid, and significant recovery may occur [5].

The measurements of the recovery kinetics are often done by the changes in a simple parameter, such as hardness, yield stress, resistivity or heat evolution. If the change from the annealed condition is X_R , then the kinetics of recovery, dX_R/dt may be determined from experiment. It is often difficult to obtain a fundamental insight into recovery from such analyses, firstly because the relationship of the parameter X_R to the microstructure is usually very complex, and secondly because of the several micromechanisms that make recovery take place with their own kinetics [5].

Experimental results have been analysed in terms of several different empirical relationships between X_R and t . The two most commonly reported isothermal relationships, which is going to be referred to as type 1 and type 2 are as follows [5]:

Type 1 kinetics [5]:

$$dX_R/dt = -c_1/t \quad (3.1.)$$

which integrates to [5];

$$X_R = c_2 - c_1 \ln t \quad (3.2.)$$

where c_1 and c_2 are constants.

It is clear that this form of relationship cannot be valid during the early stages of recovery ($t \rightarrow 0$) when $X_R \rightarrow X_0$ or at the end of recovery ($t \rightarrow \infty$) when $X_R \rightarrow 0$.

Type 2 kinetics [5]:

$$dX_R / dt = -c_1 X_R^m \quad (3.3.)$$

which integrates to [5];

$$X_R^{-(m-1)} - X_0^{-(m-1)} = (m-1)c_1 t \quad (3.4.)$$

for $m > 1$, and [5];

$$\ln(X_R) - \ln(X_0) = c_1 t \quad (3.5.)$$

for $m=1$.

Martinez-de-Guerenu et al. [10] studied recovery during annealing in a cold rolled low carbon steel and while modelling the kinetics they have assumed that the rate of recovery at a given temperature is inversely proportional to the holding time [10];

$$d\sigma_i / dt = -a / t \quad (3.6.)$$

where σ_i being the internal stress of the material and a being a constant. They have also reported that the same type of equation can be applied if hardness, resistivity or magnetic properties are considered as the measure of recovery. The integration of this equation leads to a type 1 kinetics equation [10];

$$\sigma_i = b - a \ln t \quad (3.7.)$$

where a and b are constants for a given annealing temperature. On a more fundamental base, it has been suggested that recovery may be controlled by thermally activated glide or cross slip of dislocations, leading to an apparent activation energy, $Q(\sigma_i)$, which is a decreasing function of the internal stress, σ_i . By considering a linear decrease with the internal stress $Q(\sigma_i) = Q_0 - c_2 \sigma_i$, the rate of recovery of internal stress can be expressed as [10];

$$d\sigma_i / dt = -c_1 \exp(-(Q_0 - c_2 \sigma_i) / RT) \quad (3.8.)$$

where Q_0 is the intrinsic activation energy, c_1 and c_2 are constants, R is the gas constant and T the absolute temperature. A solution to 3.8 is given by [10];

$$\sigma_i = (Q_0 / c_2) - ((RT / c_2) \ln(c_1 c_2 / RT)) - ((RT / c_2) \ln t) \quad (3.9.)$$

This equation describes the same type of kinetics as the equation 3.7 by taking b as $(Q_0 / c_2) - ((RT / c_2) \ln(c_1 c_2 / RT))$ and a as (RT / c_2) [10].

3.2.2. Recrystallization Region

During recrystallization, an entirely new set of grains is formed. New crystals are nucleated at points of high-lattice-strain energy such as slip-line intersections, deformation twin intersections, and in areas close to grain boundaries. In each case, it appears that nucleation occurs at points of lattice curvature. In this regard it is interesting to note that bent, or twisted single crystals recrystallize more readily than do similar crystals that have been bent, or twisted, and then unbent, or untwisted [11].

As the recrystallization region is divided into two regimes, nucleation which corresponds to the first appearance of new grains in the microstructure and growth during which the new grains replace the deformed material, the progress in recrystallization is commonly represented by a plot of the volume fraction of the material recrystallized (X_v) as a function of $\log(\text{time})$. This plot usually has the characteristic sigmoidal form which is illustrated in Figure 3.3 and it typically shows an apparent incubation time before recrystallization is detected. [5]

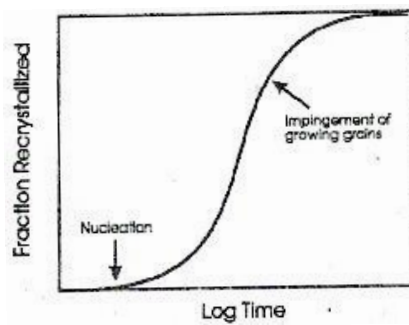


Figure 3.3: The progress of recrystallization during isothermal annealing [5].

Since the kinetics of recrystallization can often be described in terms of these two rates, a number of investigators have measured these quantities under isothermal conditions in the hope of learning more about the mechanism of recrystallization. This requires the introduction of two parameters: NR , the rate of nucleation and GR , the rate of growth[11].

It is customary to define the nucleation frequency, NR , as the number of nuclei that form per second in a cubic centimeter of unrecrystallized matrix because the recrystallized portion is inactive with regard to further nucleation. The linear rate of growth, GR is defined as the time rate of change of the diameter of a recrystallized grain. In practice, GR is measured by annealing for different lengths of time a number of identical specimens at a chosen isothermal temperature. The diameter of the largest grain in each specimen is measured after the specimens are cooled to room temperature and prepared metallographically. The variation of this nucleation can be determined from the same metallographic specimens by counting the number of grains per unit area on the surface of each. These surface-density measurements can then be used to give the number of recrystallized grains per unit volume. Of course, each determination must be corrected for the volume of the matrix that has recrystallized [11].

Several equations have been derived starting with the parameters NR and GR which express the amount of recrystallization as a function of time. The Avrami, Johnson-Mehl (JMAK) equation is [5];

$$X_v = 1 - \exp(-f(NR)(GR)^3 t^4 / 4) \quad (3.10.)$$

where X_v is the volume fraction of material recrystallized, f is a shape factor (it is assumed that the grains are spherical and f is $4\pi/3$ for spheres), NR is the nucleation frequency, GR is the time rate of change of the diameter of a recrystallized grain and t is time. This equation may be written more generally in the form [5];

$$X_v = 1 - \exp(-F t^n) \quad (3.11.)$$

These two concepts, NR and GR , are useful in explaining the effects of several other variables on the recrystallization process [11].

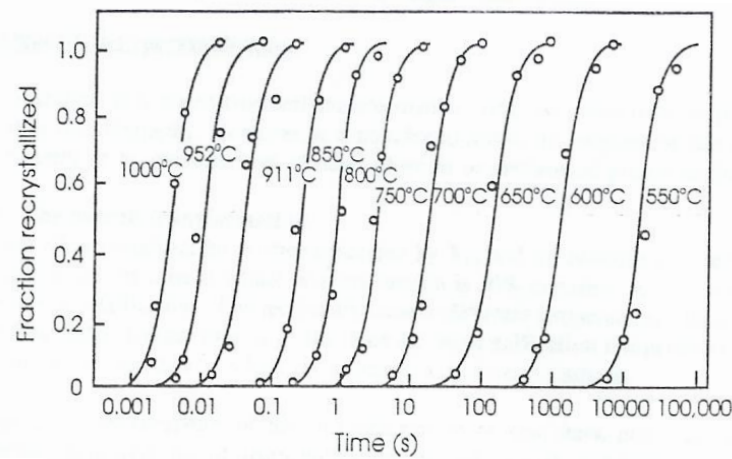


Figure 3.4: The increase on annealing time by decreasing annealing temperature for the material Fe-3,25%Si deformed to 60% [5].

The definite effect of temperature on the time of anneal can be seen in Figure 3.4. Due to the Arrhenius equation, the time of annealing increases exponentially by small decreases on temperature [5,12,13].

If we consider the transformation as a whole and take the time for 50% (t_r) recrystallization as a measure for the rate of recrystallization then a relationship can be made by excepting the Arrhenius equation [5,9,11,13],

$$\text{Rate} = 1/t_r = A \exp(-Q_{\text{Rex}} / R T) \quad (3.12.)$$

where Q_{Rex} is recrystallization activation energy, A is pre-exponential constant, R is gas constant and T is temperature.

In Figure 3.5, the plot of $\ln(t_r) \propto 1/T$, which is drawn according to Figure 3.4, gives a good straight line. The slope of the plot, which corresponds to the activation energy of Equation 3.10, gives a value of 293,3 kJ/mol for Fe-3,25% Si alloy [13]. Although this recrystallization activation energy calculation method may be of practical use, the value obtained from it is not easy to interpret, as if we are considering the transformation as a whole, and it doesn't seem to be constant. For example, such activation energies depend on strain and can vary very significantly with small changes in material purity [5].

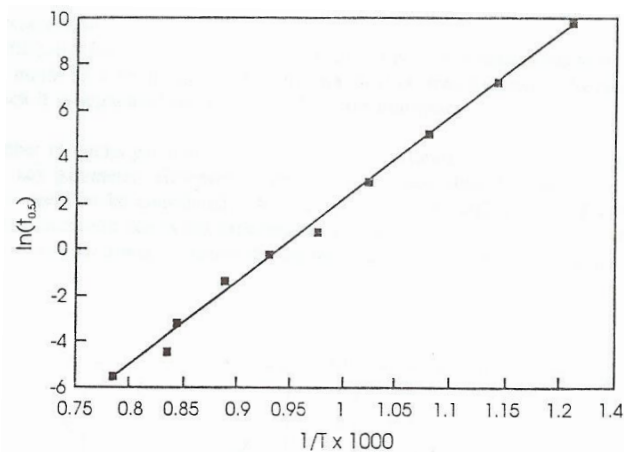


Figure 3.5: Arrhenius plot for the 50% recrystallization of Fe-3,25% Si deformed to 60% [5].

The recrystallization activation energies of some steels are listed in Table 3.1.

Table 3.1: Values of activation energy for some steel compositions.

Steel Type	Deformation Ratio %	Composition %					Recrystallization Activation Energy Q_{Rec} (kJ/mol)	Ref.
		C %	Mn %	Si %	Cr %	B %		
Low Carbon Steel	73	0,01	0,2	0,017	0,035	-	332,3	[12]
Fe-3,25% Si	60	0,019	0,006	3,27	-	-	293,3	[13]
Mn-Cr Steel	60	0,23	0,74	0,22	0,90	-	378,6	[14]
50B44	60	0,47	0,86	0,20	0,47	0,0012	253	[14]
Q235	60	0,18	0,60	0,22	-	-	363,1	[14]

3.2.3. Grain Growth Region

When the primary recrystallization is complete, further growth of the fully recrystallized grains may occur because the structure is still unstable [5]. In a completely recrystallized material, the surface energy of the grain boundaries is the driving force for the grain growth. As the grains grow in size and their numbers decrease, the grain boundary area is reduced and the total surface energy is lowered as a result [11].

Grain growth may be divided into two types as normal grain growth and abnormal grain growth or secondary recrystallization. During normal grain growth, the microstructure changes in a rather uniform way with some narrow range of grain size and shapes while during abnormal grain growth, a few grains in the

microstructure grow and consume the matrix of smaller grains. However as these large grains impinge, normal grain growth may then resume afterwards [5].

The factors affecting grain growth can be summarized as follows:

1. Temperature: As the driving force for grain growth is usually very small, significant grain growth is often found only at very high temperatures.
2. Solutes and Particles: Grain boundary pinning by solutes and second phase particles is relatively an important factor.
3. Specimen Size: The rate of grain growth decreases when the grain size becomes greater than the thickness of the specimen.
4. Texture: A strongly textured material inevitably contains many low angle boundaries and thus, the driving force for the grain growth is reduced [5].

4. COLD WORKABILITY OF STEELS

4.1. Introduction

Cold workability, which is the result of low flow stress and good deformability is the most important property of steels for cold working. Good ductility is also necessary for sufficient die filling and for preventing discontinuities. In addition to these properties, the steel must have all the properties required for good performance in service like strength, toughness, wear resistance, heat treatability and particularly hardenability as applicable. The essential properties offered by the manufacture of parts from bars or wire rods are good machinability, good surface quality, low decarburization depth and high dimensional accuracy of the finished product. Also, maximum material uniformity is an important requirement if mass production is to be done [15].

The requirements and properties listed above are somewhat conflicting. In general, however, the choice of steels is governed by the intended use of the component rather than by the manufacturing processes.

4.2. Properties Required From Cold Workable Steels

i – The flow stress, which is the decisive parameter for cold forming is determined by the initial strength and by the effect of strain hardening. However in practice, the tensile strength of a material is used in determination of cold deformability instead of flow stress because the tensile strength is easier to measure.

ii – Deformability is assessed on the basis of the reduction of area in the tensile test. In connection with cold formability, the reduction area values from 60% to 70% mean very good and 50% to 60% mean sufficient deformability.

iii – The deformability of a steel can also be determined by its microstructure, which is the result of chemical composition and heat treatment. The decisive properties of microstructure in this context are the percentage and shape of ferrite and pearlite,

the degree of spheroidization and size of carbides also being assessed. The degree of spheroidization is the proportion of globular carbides (produced by annealing) to the total amount of carbides .

iv – The requirements concerning the surface quality of the steel products to be cold formed are made by the usual metallographic and nondestructive testing methods. The cold upsetting test which is employed to access this quality is a good example where a cylindrical sample is cold headed to 1/3 or 1/4 to its initial height [15].

The determination of the properties of steels for cold forming are made primarily by the intended use of the components to be made from them, usually by specifying certain values for mechanical properties. Service properties are the main criteria in selecting steels for cold forming. However, several special measures are taken which have a favorable influence on cold formability without affecting other requirements. These measures may consist, for example, in attaining particular microstructures in steels and in fine tuning the chemical composition with regard to the strength of the ferrite or the spheroidization ability of the carbides. Therefore, the chemical composition may be mentioned as a distinct variable in addition to the microstructure. The surface condition concerning stocks and final products is also important, so that, in the final analysis, distinct steel grades are created.

Another distinction must be made between steels with a ferritic matrix and those with austenitic matrix. With ferritic, unalloyed or low alloy steels, it is essential that, whatever the final condition is to be (e.g. quenched and tempered), these steels usually are cold formed in a condition marked by a ferritic-pearlitic structure. The following comments refer to this ferritic state, which may represent an intermediate state [15].

4.2.1. Chemical Composition and Microstructure

The chemical composition of ferritic steels has a bearing on their properties as it determines their microstructure. Carbon and alloy additions to such steels increase the flow stress, which is an decisive factor in connection with cold forming. Carbon has a major impact as it determines the pearlite content which is decisive for the strength and reduction of area. Carbide forming elements such as chromium, molybdenum, vanadium and titanium also contribute to increased strength and thus to flow stress via the carbides, although to a lesser extent. Other alloying elements

such as manganese are also effective through solid solution hardening. However, this effect applies only to steels of lower carbon and ferrite.

The effect of chemical composition on the flow stress of steels with up to 0,6% C in a soft annealed condition is shown in Figure 4.1. The strengthening effect of carbon is outstanding, but the effect of other elements studied, for instance molybdenum, silicon, nickel, manganese and chromium, on the flow stress is much less, and decrease in this order. The carbide forming elements molybdenum and chromium contribute less to the flow stress at higher true plastic strains [15].

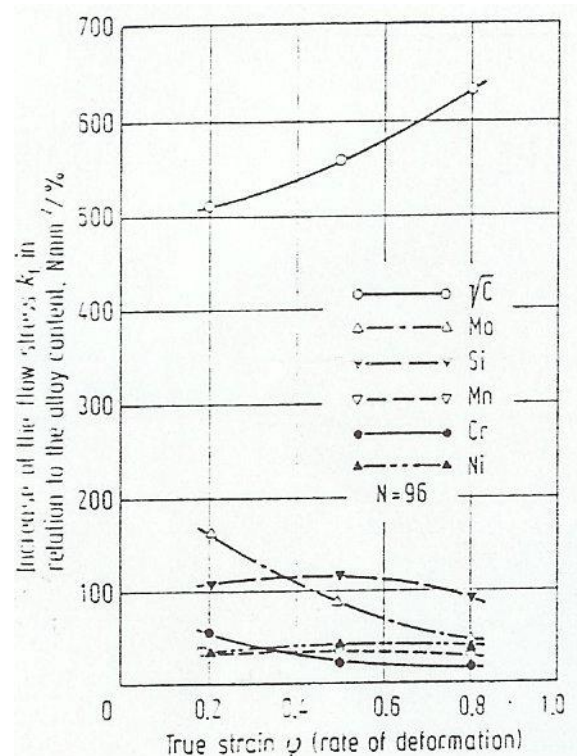


Figure 4.1: The effect of chemical composition on the flow stress of steels with up to 0,6% C in a soft annealed condition [15].

Chromium is usually added to steels which are to be heat treated after being cold formed. Chromium offers the advantage that it only slightly increases the flow stress in an annealed condition, while greatly improving hardenability. In addition, chromium steels can be more easily spheroidized, so that higher reduction of area values are obtained.

Boron has also proven as a convenient element for steels to be cold formed. This element is added to certain steels for increasing hardenability. Solute boron delays the formation of ferrite and thus increases hardenability, but has no effect on

strength in the annealed condition. Thus, despite increased hardenability, the flow stress of boron steel is the same as or even lower than that of boron free steel. By using boron the content of other hardenability increasing elements can be reduced, and cold formability can be improved of larger-sized material can be quenched and tempered because of the increased hardenability. The hardenability increasing effect of boron decreases with increasing carbon content, so that boron steels with a carbon content exceeding 0,50% are seldom used.

The microstructure is another factor which is as important as the chemical composition for cold forming of steels. Steels with low contents of carbon and alloying elements are often cold formed in a ferritic-pearlitic condition, which exists after hot rolling or normalizing. This condition is desirable when cold forming is to be followed by extensive machining. If more highly alloyed steels are also to be cold formed in the microstructural condition then a pronounced ferritic-pearlitic structure is desirable.

Considerable improvement in cold formability can be achieved when the pearlite carbides are spherically shaped by annealing. By this treatment, the tensile strength and the flow stress are lowered, and the reduction area is increased. Therefore, most steels are cold formed in this state.

The importance of the carbide form for the cold formability increases as carbon content increases. Carbide form is the most important variable for steels with a carbon content above 0,35%. Therefore, steels for quenching and tempering are always cold formed in a spheroidized condition. Carbide spheroidization is determined by the chemical composition, annealing temperature and time, the thickness of the carbide layers and the number of grain boundaries and discontinuities[15].

Higher alloyed steels are easier to spheroidize because they have a finer lamellar pearlite or possibly even a martensite-bainite initial microstructure. Spheroidization of unalloyed steels is facilitated, with increasing carbon contents, when cold forming has been done before. Generally speaking, spheroidization is greatly increased in its dependence on the true plastic strain when cold forming is done beforehand, so wire rod and bars are often annealed after a preliminary drawing [15].

4.2.2. Cleanness

Nonmetallic inclusions in steels lead to stress concentrations and cracking due to their notch effect. Under the usual deformation conditions, the standard maximum sulphur levels (e.g. $\leq 0,035\%$) specified for the steels in question generally do not have any adverse effect. If machining is to be done after deformation, then specified sulphur contents may even range from 0,020% to 0,035%. In special instances the number and size of sulphide inclusions are limited by specifying low sulphur contents; maximum levels of 0,0010% S can be achieved. Globular sulphides can be more favourable for cold formability than elongated manganese sulphides as they are produced by adding calcium, selenium or tellurium [15].

Like sulphides, globular oxides impede deformation less than linear ones. In the metallurgical processes employed today, however, the degree of oxide cleanliness achieved is so high that cold formability is not affected [15].

4.2.3. Surface Quality and Dimensional Accuracy

The surface quality of the hot rolled bar employed has a great importance for the costs and the reject rates involved in cold extrusion. However, surface flaws resulting from the production process can not be completely prevented. Therefore permissible crack or flaw depths according to classes and a decarburization depth limit are specified. If extremely exacting requirements must be met, then the bar is used after a machining operation. Surface roughness and strain hardening of the rim zone are also subject to specification. A rough surface is favorable for good phosphate coat blinding, therefore, the roughness of bright bars is sometimes increased by blasting [15].

Extremely close dimensional tolerances almost always must be maintained in connection with bars. Permissible dimensional variations of as-rolled bars often do not suffice but must be further limited. Closer dimensional tolerances are achieved by new rolling processes. Tighter tolerances can be held by cold drawing (bright steel) [15].

4.3. Convenient Steel Grades For Cold Forming

The steels for cold forming are characterized by their application properties. The characteristic steel grades proven in service can be divided into three groups according to the various requirements to be met by the components [15].

1 – Unalloyed steels with a ferritic pearlitic microstructure which derive the required component properties from cold forming without heat treatment.

The most widely used steels for this group are UQSt 36 and UQSt 38 which can be produced killed or unkilld. The microstructure of unkilld steels is not uniform, due to segregation and therefore, the components may show considerable scattering in strength. Thus, greater uniformity of cold-formed component properties can be reached by using special killed steels. Steels for this group are widely used in the cold extrusion of light-duty components in the automotive and other manufacturing industries such as shark-plug bodies and shock absorber tubes. Chemical compositions and mechanical properties of this type of steels are listed in Table 4.1.[15].

2 – Unalloyed and low-alloy steels which are generally first transformed into a condition of good formability, then cold formed followed by heat treatment to achieve the desired properties. Examples are case hardening steels, steels for quenching and tempering, nitriding steels and ball and roller bearing steels.

The steels of this group are the ones suitable for heat treatment processes such as case hardening, quenching and tempering, induction hardening or nitriding depending on their aim of usage. Cq15, 17Cr3, 13MnCrB4, 16MnCr5, 20MoCr4 and 21NiCrMo2 are the ones which are suitable for case hardening. The steels Cq 15 and 17Cr3 are preferred for production of cold extruded piston pins, valve spring retainers, bearing sleeves and self tapping screws. The alloyed case hardening steels 13MnCrB4, 16MnCr5 (also with boron addition) and 20MoCr4 are used in automobile transmissions for pinion gears, medium-duty sliding gears and bevel gears. The steel 21NiCrMo2, which is also a case hardening steel is used for higher duty components of the same type. The chemical compositions and mechanical properties of case hardening steels for cold forming are listed in Table 4.2. The steels suitable for quenching and tempering are generally used for producing cold formed bolts and similarly shaped components. Cq22 and Cq35 are the unalloyed steels which can be given as examples to these kind of steels while the qualities like

Table 4.1: Chemical compositions and mechanical properties of unalloyed steels for cold forming, not intended for heat treatment [15,16].

Steel Grade	Chemical Composition ¹			Mechanical Properties					
				Untreated Condition		Normalized Condition ⁴			
	C %	Si %	Mn%	Tensile Strength N/mm ² (max.)	Reduction of Area % (min.)	Yield Stress N/mm ² (max.)	Tensile Strength N/mm ²	Elongation After Fracture % (min.)	Notch Impact ³ Energy J (min.)
QSt 32-3 ²	≤ 0,06	≤ 0,10	0,2 – 0,4	400	60	170	290 – 400	30	27
QSt 34-3 ²	0,05 – 0,10	≤ 0,10	0,2 – 0,4	400	60	180	310 – 420	30	27
UQSt 36 ²	≤ 0,14	Trace	0,25 – 0,50	430	60	180	120 – 430	30	27
QSt 36-3 ²	0,06 – 0,13	≤ 0,10	0,25 – 0,45	430	60	200	320 – 430	30	27
UQSt 38	≤ 0,19	Trace	0,25 – 0,50	460	55	220	300 – 460	25	27
QSt 38-3 ²	0,10 – 0,18	≤ 0,10	0,25 – 0,45	460	55	220	360 – 460	25	27

¹ Content of phosphorus and sulphur each max. 0,040 %.

² Killed with al (≥ 0,0020 % Al_{total}) or with other elements, having a similar effect for deoxidizing and fixing nitrogen.

³ ISO-V notch test specimens (longitudinal) at 20°C.

⁴ Normalized condition is unusual.

Table 4.2: Chemical composition and mechanical properties of case hardening steels for cold forming [15,16].

Steel Grade	Chemical Composition ¹					Mechanical Properties					
						GKZ ⁴ Condition		Blank Hardened Condition			
	C %	Si %	Mn %	Cr %	Mo %	Tensile Strength N/mm ² (max.)	Reduction of Area % (min.)	Yield Stress N/mm ² (max.)	Tensile Strength N/mm ²	Elongation After Fracture % (min.)	Reduction of Area %(min.)
Cq15	0,12 – 0,18	0,15 – 0,35	0,25 – 0,50	-	-	470	65	355	590 – 790	14	45
17Cr3	0,14 – 0,20	0,15 – 0,40	0,40 – 0,70	0,60 – 0,90	-	530	60	450	700 – 900	11	40
16MnCr5	0,14 – 0,19	0,15 – 0,40	1,00 – 1,30	0,80 – 1,10	-	550	60	590	780 – 1080	10	40
20MoCr4	0,17 – 0,22	0,15 – 0,40	0,60 – 0,90	0,30 – 0,50	0,40 – 0,50	550	60	590	780 – 1080	10	40
21NiCrMo2²	0,17 – 0,23	0,15 – 0,40	0,60 – 0,90	0,35 – 0,65	0,15 – 0,25	590	59	590	830 – 1130	9	40
13MnCrB4³	0,12 – 0,16	≤ 0,20	1,00 – 1,20	0,80 – 1,10	-	540	60	600	800 – 1100	10	40

¹ Phosphorus and sulphur content each max. 0,035%, except 13MnCrB4 with 0,020 – 0,035% sulphur.

² Besides 0,40% to 0,70% Ni.

³ Besides 0,0010% to 0,0030% B.

⁴ GKZ = spheroidizing annealed to globular cementite (carbide).

38Cr2, 41Cr4, 34CrMo4, 22B2, 35B2, 25MnB4 and 37CrB1 are the low alloy steels of this type. 31CrMoV9, which is a steel suitable for nitriding is used for cold-formed piston pins. 105Cr4, which is a ball and roller bearing steel is used for cold formed rollers and bars. The chemical compositions and mechanical properties of the steels for quenching and tempering which are intended for cold forming are listed in Table 4.3.

3 – High-alloy steels with a ferritic or austenitic microstructure which, depending on their intended use, possess properties such as strength at elevated temperatures, high-temperature strength, heat resistance and corrosion resistance. Ferritic steels are cold formed in an intermediate condition, while the austenitic steels are cold formed in the condition as received.

The high-alloy steels of this group must be used for cold heading and cold extrusion of components for special applications requiring high temperatures and aggressive media. Only a small group of these kind of steels are suitable for cold forming. These steels are classified as ferritic, martensitic and austenitic depending on their microstructure. The microstructure also indicates the degree of deformability. Ferritic and martensitic steels are cold formed in a spheroidize-annealed condition and their behaviour is similar to that of unalloyed steels. In comparison, austenitic steels possess higher ductility which, when subjected to plastic deformation, has a larger usable range between yield stress and tensile strength in a quenched hot tempered condition. Therefore, this type of steels are preferred for cold forming and for difficult forming operations. X5CrNi 19 11, X5CrNiMo 18 10, which are the examples for austenitic chrome-nickel steels are used for nuts, bolts and other parts likely to be exposed to corrosion by water, the atmosphere and acids. X2CrNiMo 18 3, an ULC (Ultra Low Carbon) steel is another example for this type containing nitrogen between 0,12% and 0,22% which improves austenite stability and increases strength. The carbon content of this kind of steels are limited to 0,03% for increased resistance to intercrystalline corrosion. This kind of steels are particularly convenient for complicated forming operations. X10CrNiMoTi 18 10, another type of austenitic steel stabilized with titanium or niobium has a less favorable cold formability according to the other ones due to their titanium and niobium content needed for fixing carbon and for improving their resistance to intercrystalline corrosion. The chemical composition and mechanical properties of austenitic steels suitable for cold forming are listed in Table 4.4.

Table 4.3: Chemical composition and mechanical properties of steels for quenching and tempering, intended for cold forming [15,17].

Steel Grade	Chemical Composition ¹					Mechanical Properties						
						GKZ ² Condition		Quenched and Tempered Condition				
	C %	Si %	Mn %	Cr %	B %	Tensile Strength N/mm ² (max.)	Reduction of Area % (min.)	Yield Stress N/mm ² (max.)	Tensile Strength N/mm ²	Elongation After Fracture % (min.)	Reduction of Area % (min.)	Notch ³ Impact Energy J (min.)
Cq 22	0,18 – 0,24	0,15 – 0,35	0,30 – 0,60	-	-	500	60	295	490 – 640	22	50	39
Cq 35	0,32 – 0,39	0,15 – 0,35	0,50 – 0,80	-	-	570	60	365	580 – 730	19	45	29
38Cr2	0,34 – 0,41	0,15 – 0,40	0,50 – 0,80	-	-	600	58	440	690 – 840	15	45	29
41Cr4	0,38 – 0,45	0,15 – 0,40	0,50 – 0,80	0,40 – 0,60	-	620	57	665	880 – 1080	12	45	29
34CrMo4⁴	0,30 – 0,37	0,15 – 0,40	0,50 – 0,80	0,90 – 1,20	-	610	58	665	880 – 1080	12	50	34
22B2	0,19 – 0,25	0,15 – 0,40	0,50 – 0,80	0,90 – 1,20	0,0008 – 0,0050	500	60	370	590 – 740	18	50	60
35B2	0,32 – 0,40	0,15 – 0,40	0,50 – 0,80	-	0,0008 – 0,0050	570	60	510	690 – 830	16	45	40
25MnB4	0,21 – 0,28	0,15 – 0,40	0,90 – 1,20	-	min. 0,0008	520	60	580	750 – 900	14	55	40
37CrB1	0,35 – 0,40	0,15 – 0,35	0,50 – 0,80	-	min. 0,0008	580	59	590	740 – 880	15	45	35

¹ Phosphorus and sulphur content each max. 0,035%.

³ ISO-U notch test specimens (longitudinal) at 20 °C.

² GKZ = spheroidizing annealed to globular cementite (carbide).

⁴ Besides 0,15% to 0,30% molybdenum.

Table 4.4: Chemical composition and mechanical properties of austenitic stainless steels, intended for cold forming [15,16].

Steel Grade	Chemical Composition ¹					Mechanical Properties for the quenched condition			
	C % (max)	Cr %	Mo %	Ni %	Other %	Yield Stress N/mm ² (max.)	Tensile Strength N/mm ²	Elongation After Fracture % (min.)	Notch ² Impact Energy % (min.)
X5CrNi 19 11	0,07	17,0 – 20,0	-	10,5 – 12,0		185	500 – 700	50	60
X5CrNiMo 18 10	0,07	16,5 – 18,5	2,00 – 2,50	10,5 – 13,5		205	500 – 700	45	60
X2CrNiMoN 18 13	0,03	16,5 – 18,5	2,50 – 3,00	12,0 – 14,5	N: 0,14 – 0,22	300	600 – 800	40	60
X10CrNiMoTi 18 10	0,10	16,5 – 18,5	2,00 – 2,50	10,5 – 13,5	Ti ≥ 5 x C %	225	500 – 750	40	60

¹ Besides in all steel grades max. 1% silicon, 2% manganese, max. 0,045% phosphorus and max. 0,030% sulphur.

² ISO-U notch test specimens (longitudinal) at 20°C.

5. EXPERIMENTAL STUDY

The recrystallization experiments have been conducted on cold drawn steel bars. The chemical compositions and cold drawing ratios of the examined bars are presented in Table 4.1 and Table 4.2 respectively. Qualities of the examined steels are; RSt 35-2, 21Mn5 and 20NiCrMo3 according to DIN 17115 [18].

Table 5.1: The chemical compositions of the examined steels.

Standard	Material Number	Steel Quality	C%	Si%	Mn%	P%	S%	Cr%	Mo%	Ni%
DIN 17115	1,0208	RSt 35-2	0,1126	0,2283	0,561	0,0101	0,0184	0,0101	<0,005	0,0575
DIN 17115	1,0495	21Mn5	0,1985	0,2811	1,461	0,0076	0,0019	0,1492	<0,005	0,0633
DIN 17115	1,6527	20NiCrMo3	0,2135	0,2146	0,708	0,0063	0,0014	0,546	0,2252	0,889

Table 5.2: Cold drawing ratio of examined steels.

Steel Quality	Initial Diameter(mm)	Examined Reduction I		Examined Reduction II		Examined Reduction III	
		Dia.(mm)	Red. %	Dia.(mm)	Red. %	Dia.(mm)	Red. %
RSt 35-2	13,5	12,0	23	9,0	56	7,0	74
21Mn5	13,5	11,0	34	9,0	56	7,0	73
20NiCrMo3	11,0	9,0	33	7,0	60	5,0	79

The samples were prepared for recrystallization studies by cutting the bars in the length of 8 mm. Recrystallization annealing temperatures were selected as 625, 650 and 700°C. At 625 and 650°C, samples were annealed in an atmospheric electrical furnace. At 700°C, samples were annealed in a salt bath. The commercial name of the salt bath was Petrofer HS 550. After the annealing processes, the samples were grinded by 120 mesh SiC paper. The hardnesses of the samples have been measured by a Rockwell hardness tester in the scale of HRB. For each specimen, at least 5 hardness measurements were averaged. Development of recrystallization

process was also followed by an optical microscope on the samples prepared by standard metallographic technique. Grinded and polished samples were etched with 2% nital for this purpose.

6. RESULTS AND DISCUSSION

The results of hardness measurements conducted on the samples after recrystallization annealing are listed in Appendix Tables A1-A3. Recrystallization curves of the examined steels are plotted in the terms of “normalized hardness – annealing time” graphs in Appendix Figures A.1 for RST 35-2, A.2 for 21Mn5 and A.3 for 20NiCrMo3 quality steels. Normalized hardness values were calculated by dividing the hardness of the sample annealed for a certain time to that of cold drawn condition. Three regions were evident (recovery, recrystallization and grain growth) on the recrystallization curves presented in Figures A1-A3. At the initial stages of annealing, no decrease in hardness was observed (recovery region). Hardness was sharply dropped at extended annealing times (recrystallization region) and finally reached to a steady state value (grain growth region). It should be mentioned that, recrystallization of 23% cold drawn RSt 35-2 steel yielded a gradual decrease in hardness when compared to other steel qualities and reductions.

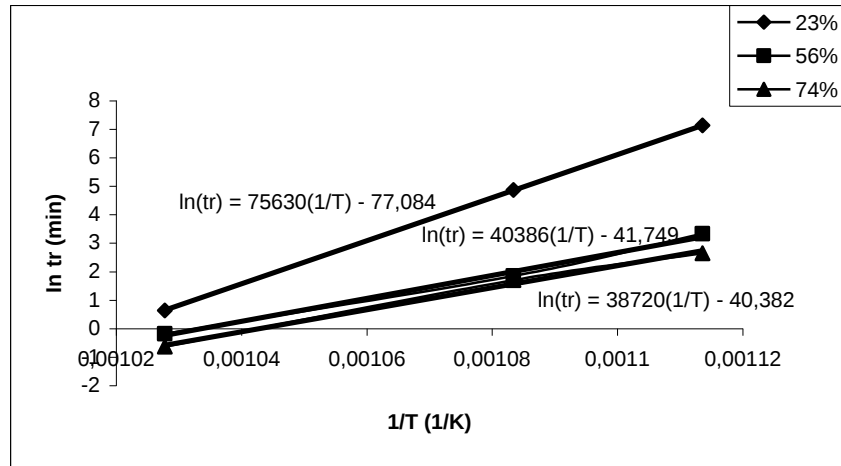
In the present study, 50% decrease in hardness during annealing is considered as 50% recrystallization in the microstructures. Table 6.1. lists the recrystallization times of each annealing temperatures corresponding to 50% decrease in hardness which were determined by the recrystallization curves given in Appedix Figures A.1, A.2, and A.3. Recrystallization curves also indicate that high amount of cold drawing leads short time for recrystallization.

The Arrhenius plots of the examined steels are shown in Figure 6.1 as “ $\ln t_r - 1/T$ ” graphs. According to the Arrhenius equation (Equation 3.12), the slope of the “ $\ln t_r - 1/T$ ” graph is proportional with the recrystallization activation energy (Q_{Rex}). Q_{Rex} values were found by multiplying the slopes of Figure 6.1 by 8,314 J/molK which is the gas constant. The Q_{Rex} values calculated by this method are listed in Table 6.2, along with pre-exponential constant of Arrhenius equation.

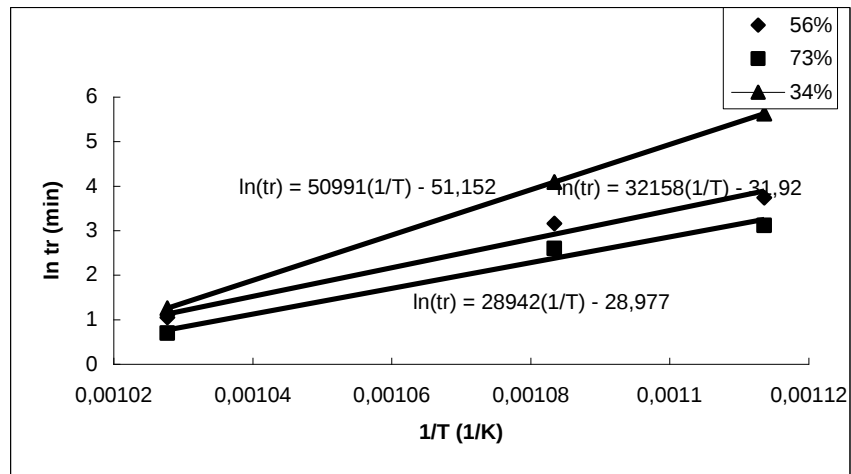
Table 6.1: The recrystallization times of the examined steels.

Cold Drawing %	Annealing Temperature (°C)	50% Recrystallization Time (t_r) (min)
RSt 35-2		
23	625	1250
	650	129,7
	700	1,9
56	625	27,9
	650	6,3
	700	0,7
74	625	14,1
	650	5,5
	700	0,5
21Mn5		
34	625	278,4
	650	60
	700	3,5
56	625	42
	650	23,5
	700	2,9
73	625	22,5
	650	13,4
	700	2
20NiCrMo3		
33	625	172,6
	650	36
	700	2,3
60	625	112,4
	650	25,3
	700	1,6
79	625	50,2
	650	22
	700	1

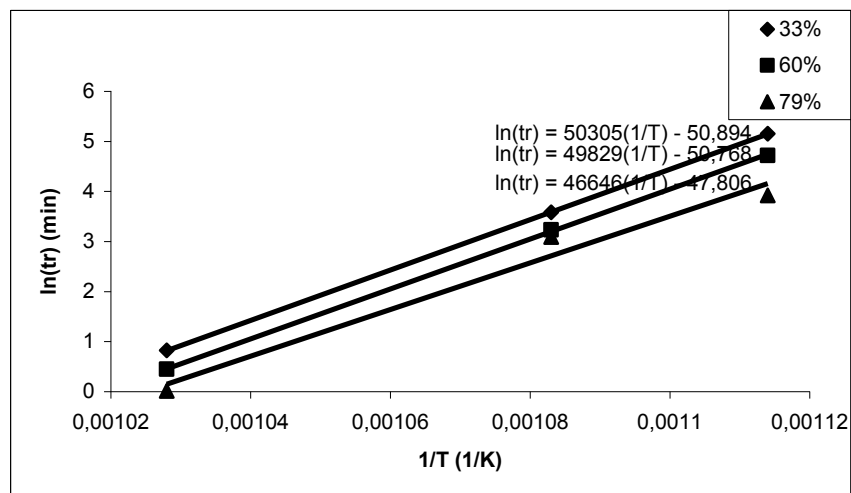
Table 6.2 shows that heavy cold drawing yielded a decrease in Q_{Rex} values. Among the examined steels, 21Mn5 has the lowest and 20NiCrMo3 has the highest Q_{Rex} value. Q_{Rex} values of RSt 35-2 steel listed in Table 6.2 are in accordance with the result of previous studies given in Table 3.1. In the literature [19], it has been reported that, Mn addition extents the Arrhenius plot of steels to longer times. As seen in Figure 6.1, 21Mn5 and 20NiCrMo3 steels yielded longer recrystallization times when compared with RSt 35-2 especially at high cold drawing values.



(a)



(b)



(c)

Figure 6.1: The Arrhenius plots of (a) RSt 35-2, (b) 21Mn5, (c) 20NiCrMo3 quality steels utilized to calculate the recrystallization activation energy.

Table 6.2: Recrystallization activation energies and pre-exponential constant values of the examined steels.

Cold Red. (%)	Q_{Rex} (kJ/mol)	A (1/min)
RSt 35-2		
23	629	$3,33 \times 10^{-34}$
56	336	$7,39 \times 10^{-19}$
74	322	$2,89 \times 10^{-18}$
21Mn5		
34	424	$6,09 \times 10^{-23}$
56	267	$1,37 \times 10^{-14}$
73	241	$2,60 \times 10^{-13}$
20NiCrMo3		
33	418	$7,88 \times 10^{-23}$
60	414	$8,94 \times 10^{-23}$
79	388	$1,73 \times 10^{-21}$

The microstructures developed during recrystallization annealing of 60% cold drawn 20NiCrMo3 is presented in Figure 6.2 as an example. The development of recrystallization is based on the replacement of elongated ferritic grains by new equiaxial grains. After a certain annealing time, new nearly dislocation free grains form (Figure 6.2.d) and grow into the still deformed ferritic surroundings. At extended annealing times, newly formed grains coarsen (Figures 6.2.f). The morphology of pearlitic regions remain unchanged during the annealing processes.

The microstructural changes of the examined steels cold drawn in different reductions are shown in the Appendix Figures B,C,D for different annealing temperatures and times. Microscopic examinations revealed that recrystallization of ferritic grains in slightly cold drawn steels is not evident when compared to the heavily cold drawn condition.

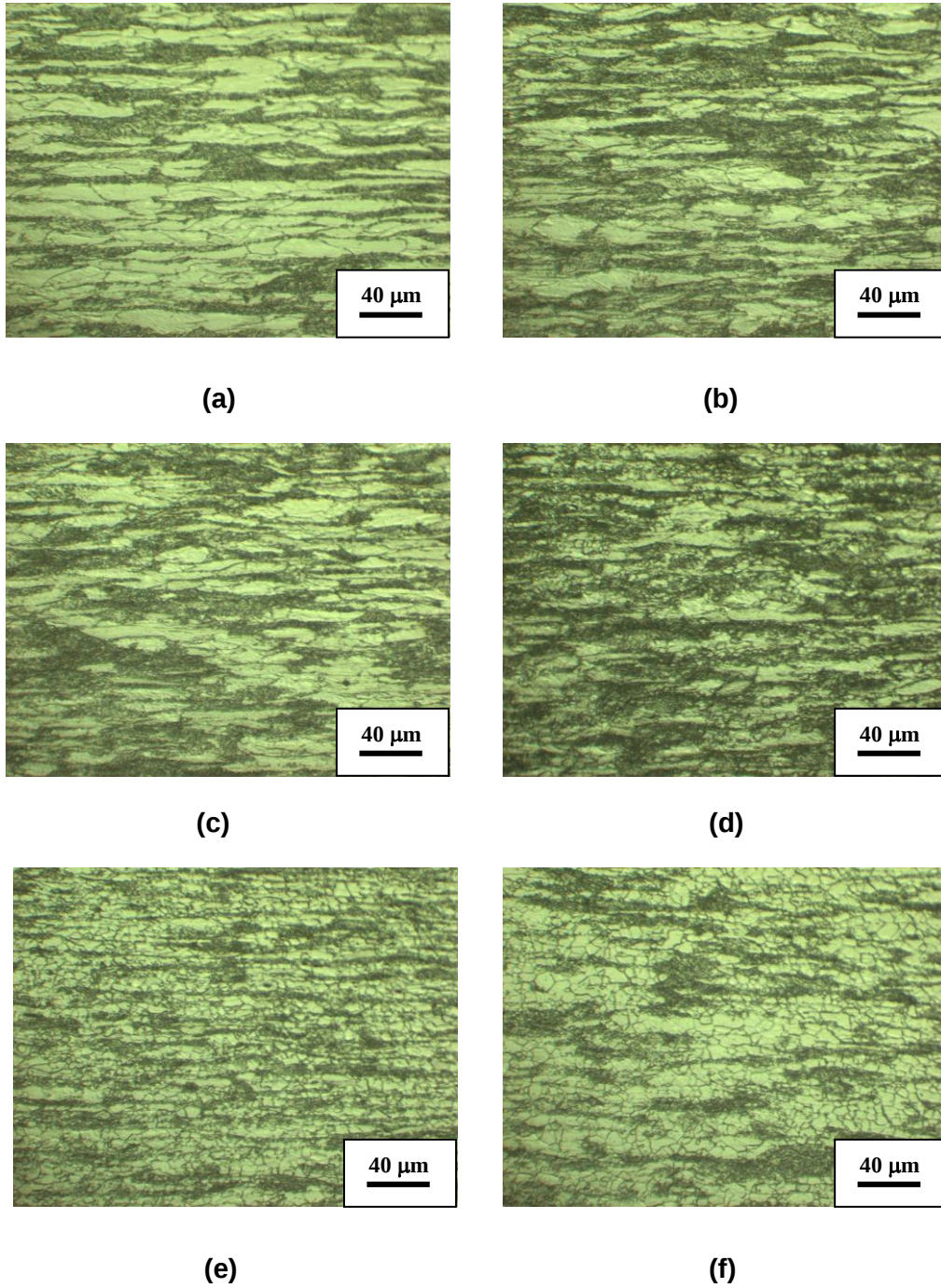


Figure 6.2: Recrystallization evolution of 60% cold drawn 20NiCrMo3 **(a)** As received **(b)** 30 seconds **(c)** 1 minute **(d)** 1,5 minutes **(e)** 2 minutes **(f)** 2,5 minutes (400x).

In this study, a correlation has been made with the results of G. Dönmez [20], who studied deformation behaviour of the examined steels. G. Dönmez reported “True Stress – True Strain” correlation empirically as presented in Table 6.3, from the envelope curves. The variation of strain hardening rate ($d\sigma/d\varepsilon$) calculated from the empirical equation given in Table 6.3 are plotted with respect to the cold drawing ratios in Figure 6.3 along with the corresponding Q_{Rex} values.

Table 6.3: Strain hardening rate ($d\sigma/d\varepsilon$) equations of the examined steels [20].

Steel Quality	Strain Hardening Rate ($d\sigma/d\varepsilon$)
RSt 35-2	$7931,5\varepsilon^4 - 25402,4\varepsilon^3 + 29004\varepsilon^2 - 14114\varepsilon + 2679,4$
21Mn5	$4046\varepsilon^4 - 13125,6\varepsilon^3 + 15257,4\varepsilon^2 - 7804,2\varepsilon + 1736,2$
20NiCrMo3	$1888,05\varepsilon^4 - 7669,6\varepsilon^3 + 11146,2\varepsilon^2 - 6837,2\varepsilon + 1634,1$

Figure 6.3 shows that, Q_{Rex} tends to decrease with increasing cold drawing ratios similar to the strain hardening rate of cold drawing process. It should be mentioned that at regions where $d\sigma/d\varepsilon$ becomes independent from the deformation ratio, Q_{Rex} values also tends to remain constant.

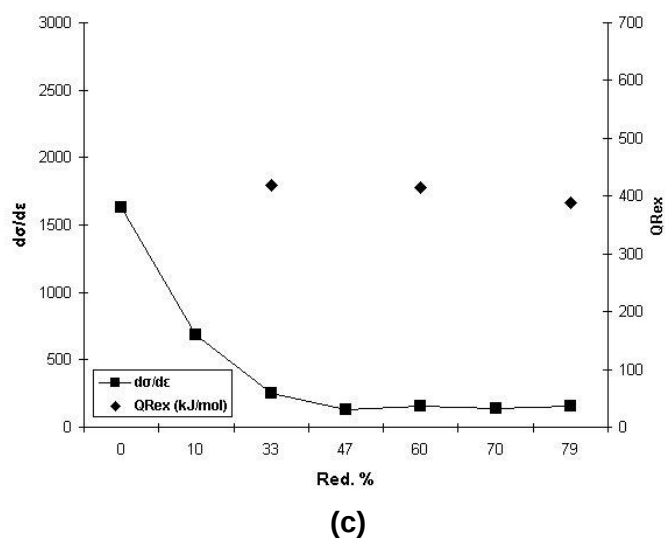
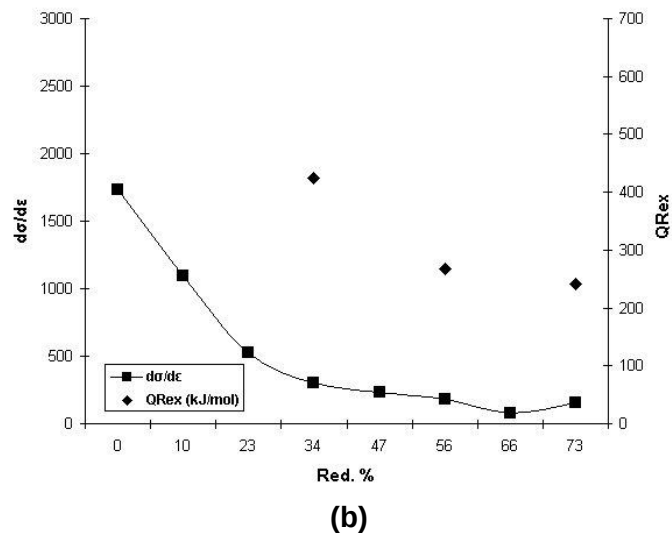
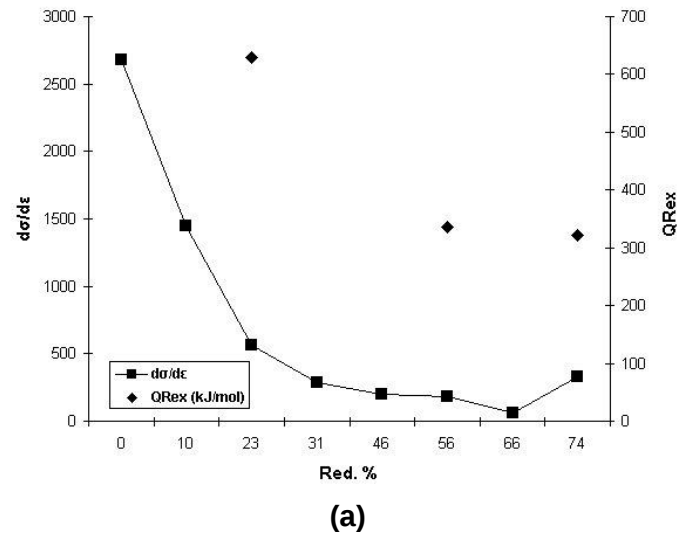


Figure 6.3: The effect of cold drawing ratio on strain hardening rate ($d\sigma/d\epsilon$) and recrystallization activation energy (Q_{Rex}) for (a)RSt 35-2 (b)21Mn5 (c)20NiCrMo3 quality steels.

7. CONCLUSION

In the present study, the effect of cold drawing ratio on the recrystallization behaviours of RSt 35-2, 21Mn5 and 20NiCrMo3 quality steels were examined. The results of the experimental observations can be summarized as below:

1 – As a general trend, the recrystallization activation energies of the examined steels decreased with increasing cold drawing ratio. At cold drawing ratios higher than 35%, Q_{Rex} stayed independent from the cold drawing ratio.

2 – Below cold drawing ratios less than 35%, RSt 35-2 quality steel exhibited the highest Q_{Rex} value while the Q_{Rex} values of 21Mn5 and 20NiCrMo3 steel qualities were almost the same. Above the cold drawing ratios of 35%, 20NiCrMo3 had the highest and 21Mn5 had the lowest Q_{Rex} values.

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APPENDIX

Table A.1: Averaged hardness values of RSt 35-2.

Annealing Time (min)	Hardness (HRB)		
	23% Red.	56% Red.	74% Red.
	As Received		
	94 (224 HV)	98 (241 HV)	100 (260 HV)
	Annealing Temperature: 625°C		
15	94	87	88
30	94	86	80
60	94	82	77
120	94	77	77
240	91	77	
360	90		
480	89		
960	86		
1200	86		
1440	83		
1800	81		
	Annealing Temperature: 650°C		
3	94	98	100
8	94	82	77
15	94	78	77
30	94	77	77
60	90	77	
120	87		
180	80		
240	77		
300	77		
	Annealing Temperature: 700°C		
0	94	98	100
0,5	94	98	100
0,75	94	98	79
1	94	78	77
2	85	77	77
3	79	77	
4	78	77	
5	78	77	
6	77		

Table A.2: Averaged hardness values of 21Mn5.

Annealing Time (min)	Hardness (HRB)		
	34% Red.	56% Red.	73% Red.
	As Received		
	97 (245 HV)	100 (259 HV)	102 (265 HV)
	Annealing Temperature: 625°C		
8	97	100	102
15	97	100	96
30	97	100	91
45	97	90	89
60	97	88	88
120	97	85	86
180	94	84	85
240	87		
300	84		
360	84		
480	83		
600	83		
720	82		
	Annealing Temperature: 650°C		
5	97	100	102
10	97	95	92
15	97	91	91
20	97	90	89
30	97	89	87
60	88	84	83
120	84	82	81
180	80	80	80
	Annealing Temperature: 700°C		
1	97	100	102
2	97	98	93
3	92	90	89
4	90	85	85
5	85	85	85
6	85	85	85

Table A.3: Averaged hardness values of 20NiCrMo3.

Annealing Time (min)	Hardness (HRB)		
	33% Red.	60% Red.	79% Red.
	As Received		
	98 (218 HV)	100 (247 HV)	102 (264 HV)
	Annealing Temperature: 625°C		
30	98	100	102
60	98	100	94
90	98	95	93
120	96	92	90
180	91	86	85
240	88	85	84
300	87	83	83
360	86		
	Annealing Temperature: 650°C		
8	98	100	102
15	98	100	102
20	98	97	97
25	98	93	88
30	94	88	88
45	89	85	85
60	86	83	83
90	85		
	Annealing Temperature: 700°C		
0,5	98	100	102
1	98	100	93
1,5	95	93	89
2	92	87	86
2,5	89	86	86
3	85	85	84
3,5	85	84	84

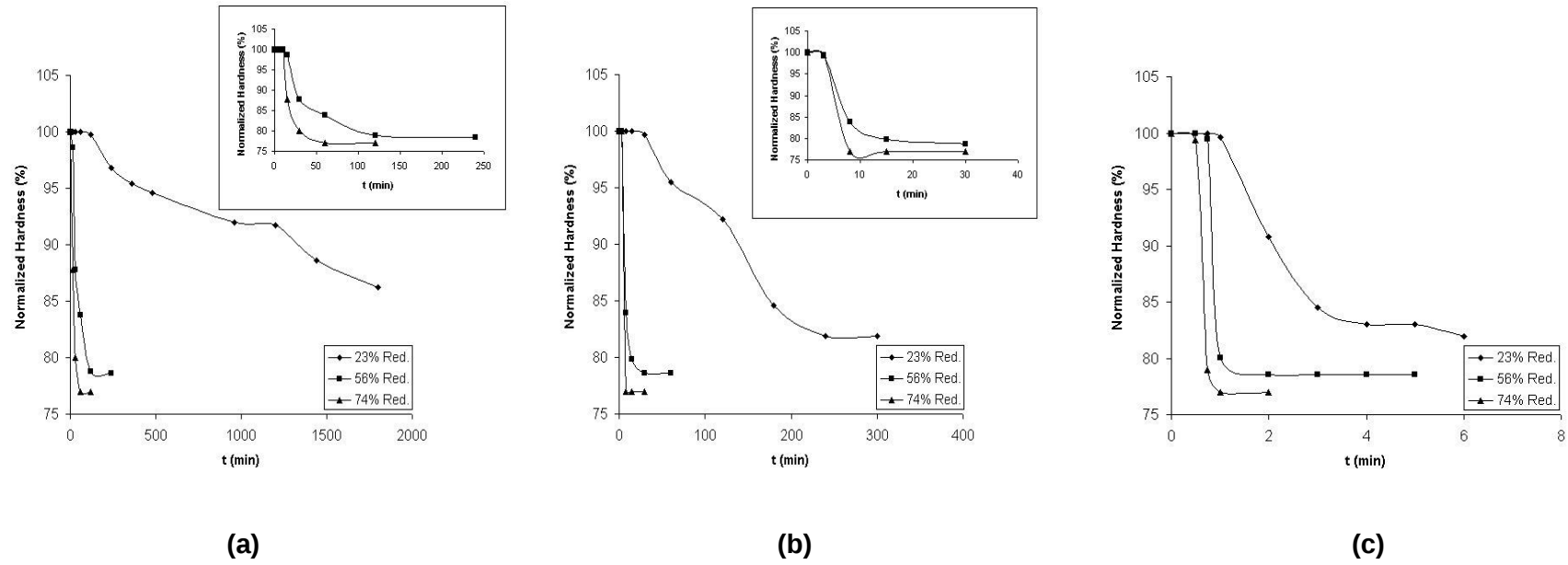
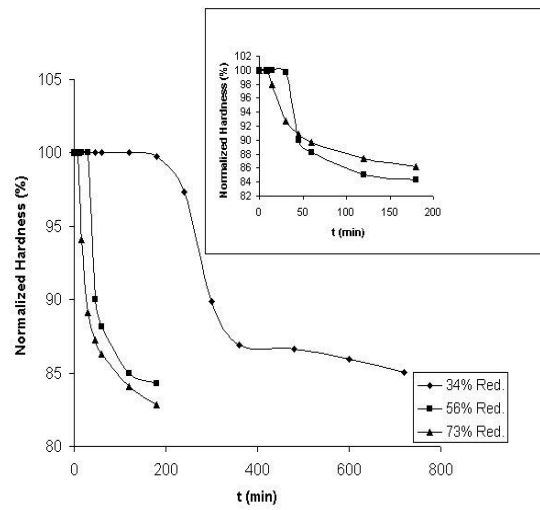
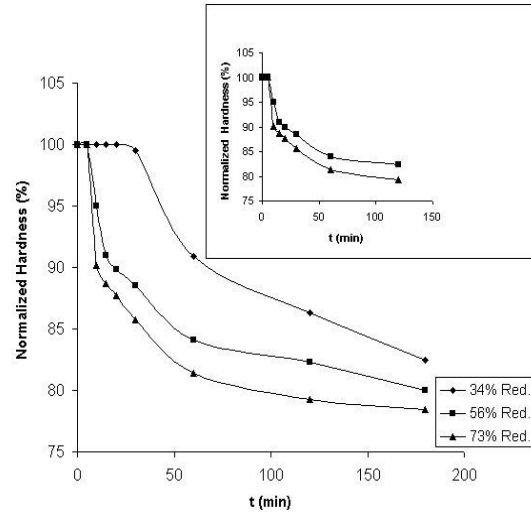


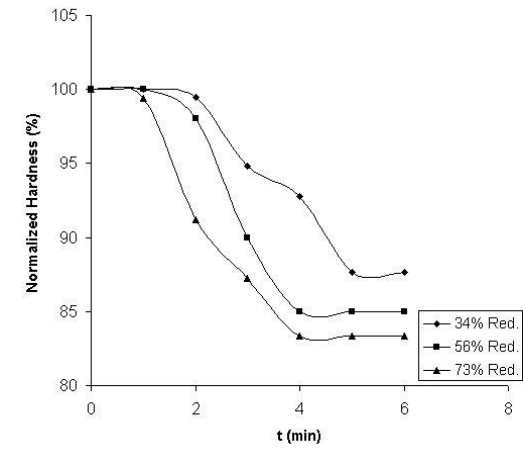
Figure A.1: Recrystallization curves of RST 35-2 quality steel annealed at (a)625°C (b)650°C (c)700°C.



(a)

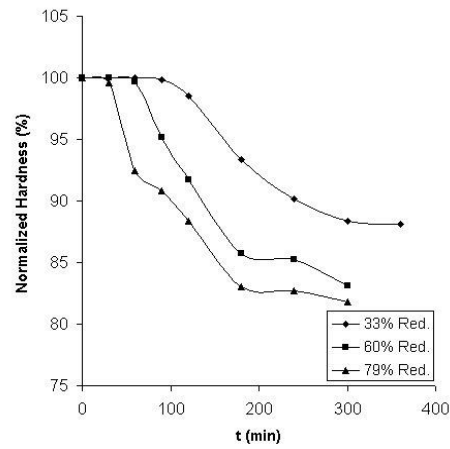


(b)

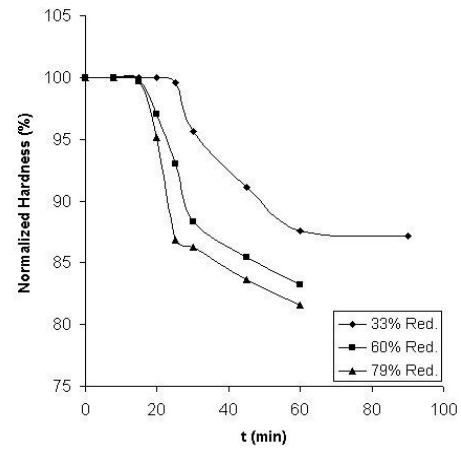


(c)

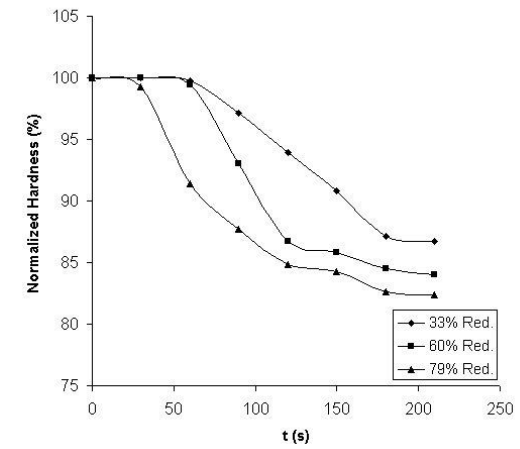
Figure A.2: Recrystallization curves of 21Mn5 quality steel annealed at (a)625°C (b)650°C (c)700°C.



(a)



(b)



(c)

Figure A.3: Recrystallization curves of 20NiCrMo3 quality steel annealed at (a)625°C (b)650°C (c)700°C.

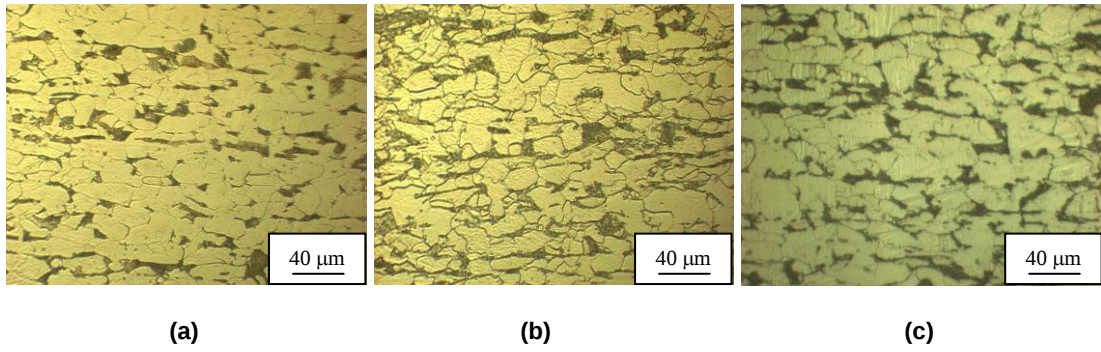


Figure B.1: RSt 35-2 23% Red. annealed at 625°C (a) As received (b) 8 h (c) 22 h (400x).

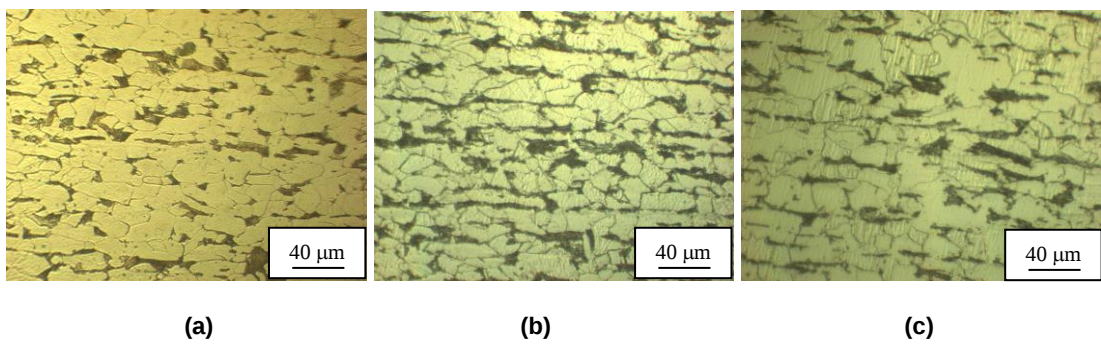


Figure B.2: RSt 35-2 23% Red. annealed at 650°C (a) As received (b) 2 h (c) 4 h (400x).

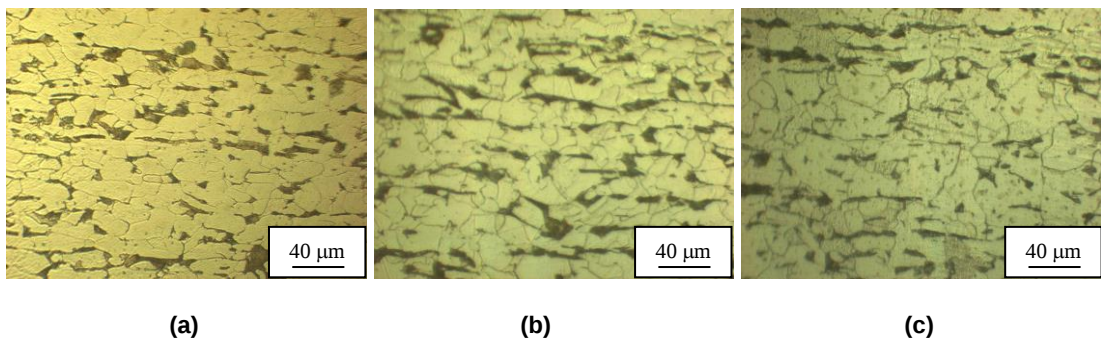


Figure B.3: RSt 35-2 23% Red. annealed at 700°C (a) As received (b) 30 min (c) 2 h (400x).

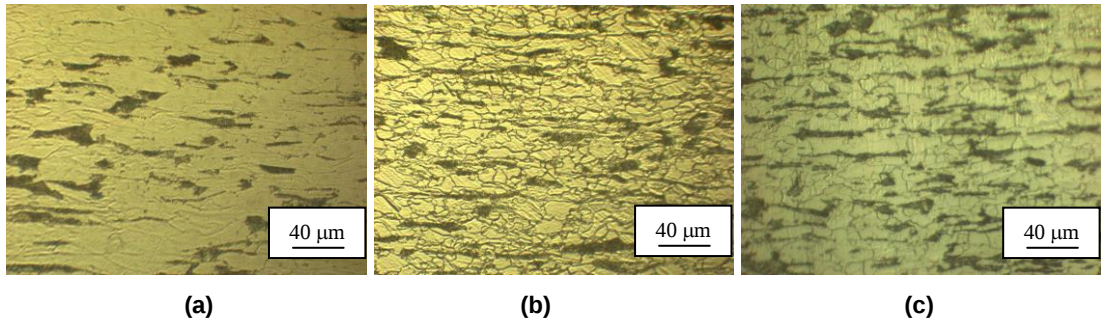


Figure B.4: RSt 35-2 56% Red. annealed at 625°C (a) As received (b) 30 min (c) 2 h (400x).

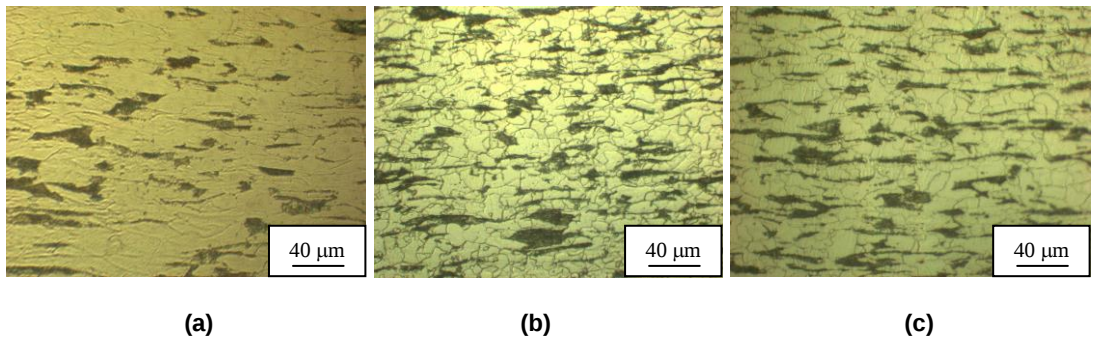


Figure B.5: RSt 35-2 56% Red. annealed at 650°C (a) As received (b) 8 min (c) 30 min (400x).

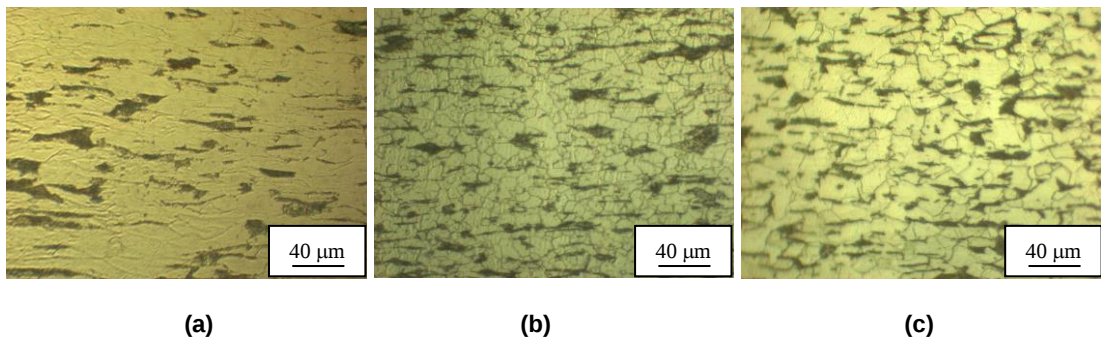


Figure B.6: RSt 35-2 56% Red. annealed at 700°C (a) As received (b) 30 s (c) 1 min (400x).

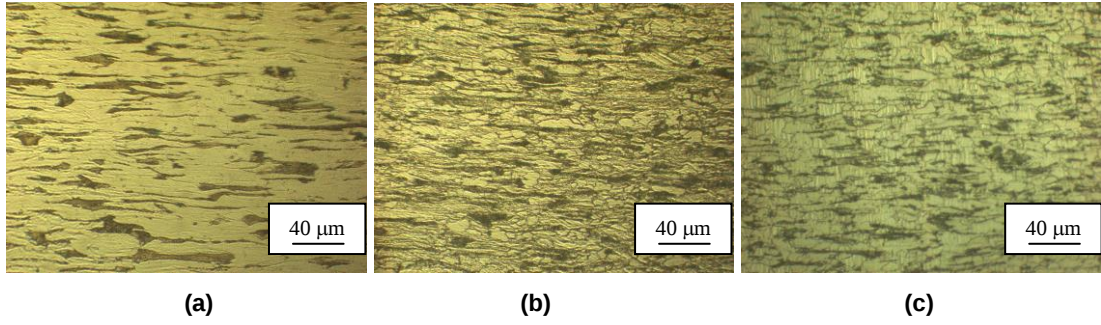


Figure B.7: RSt 35-2 74% Red. annealed at 625°C (a) As received (b) 15 min (c) 1 h (400x).

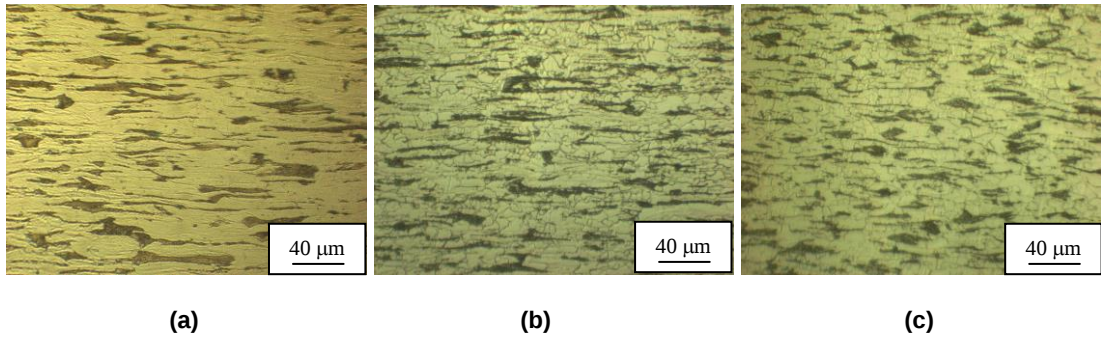


Figure B.8: RSt 35-2 74% Red. annealed at 650°C (a) As received (b) 7 min (c) 30 min (400x).

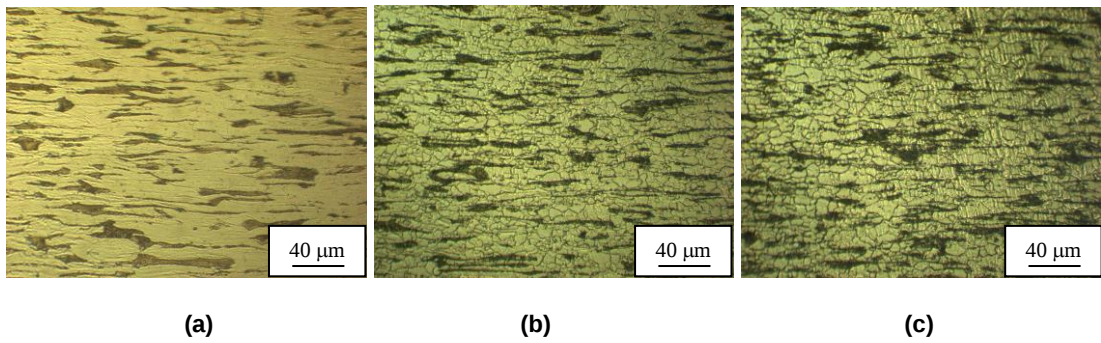


Figure B.9: RSt 35-2 74% Red. annealed at 700°C (a) As received (b) 45 s (c) 1 min (400x).

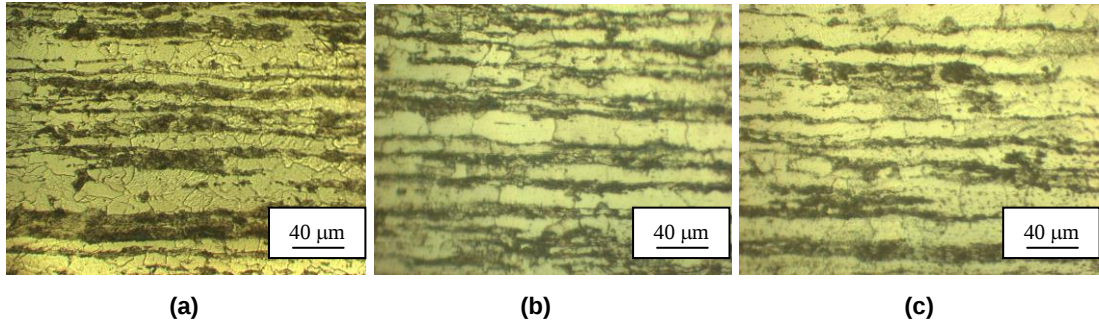


Figure C.1: 21Mn5 34% Red. annealed at 625°C (a) As received (b) 5 h (c) 10 h (400x).

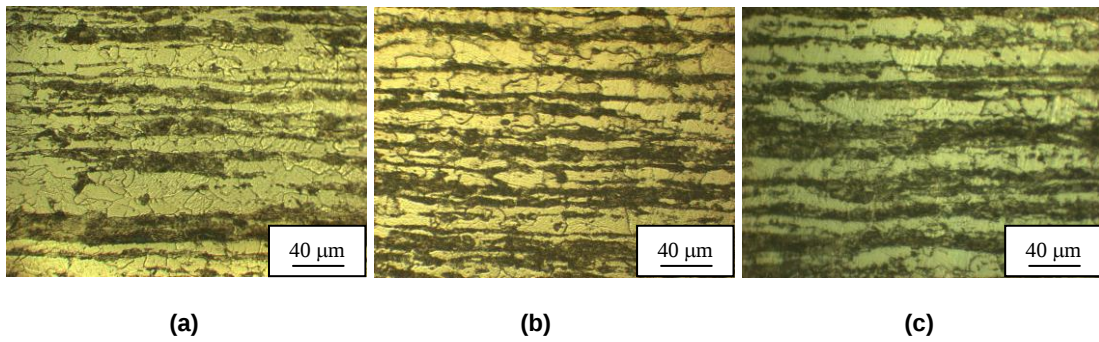


Figure C.2: 21Mn5 34% Red. annealed at 650°C (a) As received (b) 90 min (c) 3 h (400x).

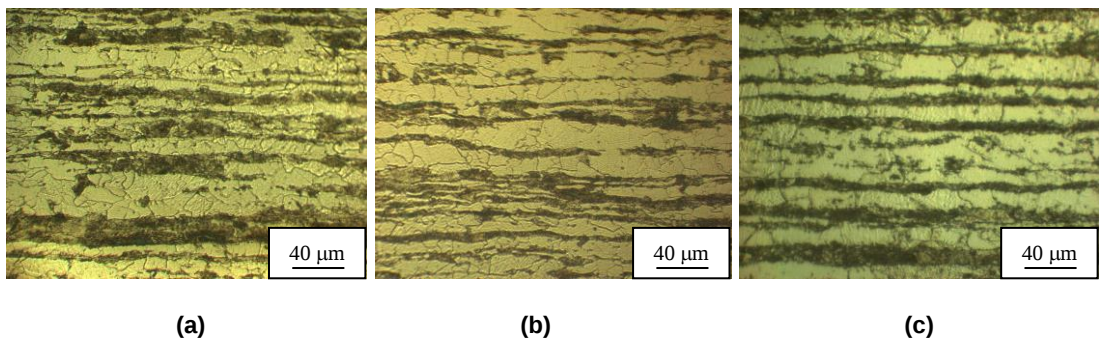


Figure C.3: 21Mn5 34% Red. annealed at 700°C (a) As received (b) 4 min (c) 6 min (400x).

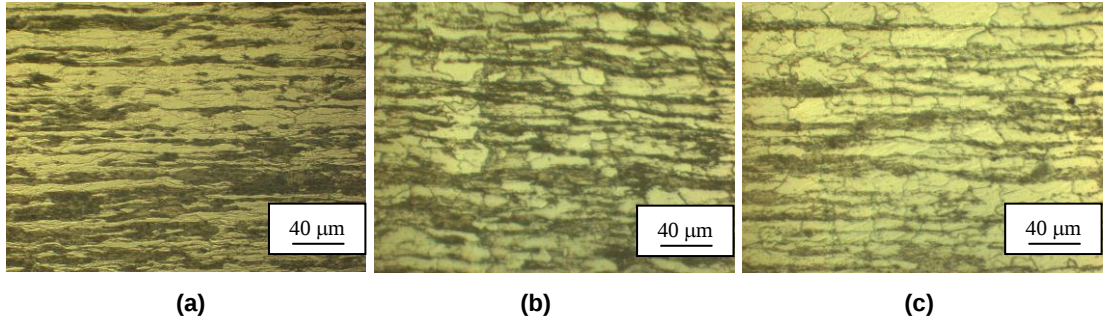


Figure C.4: 21Mn5 56% Red. annealed at 625°C (a) As received (b) 1 h (c) 3 h (400x).

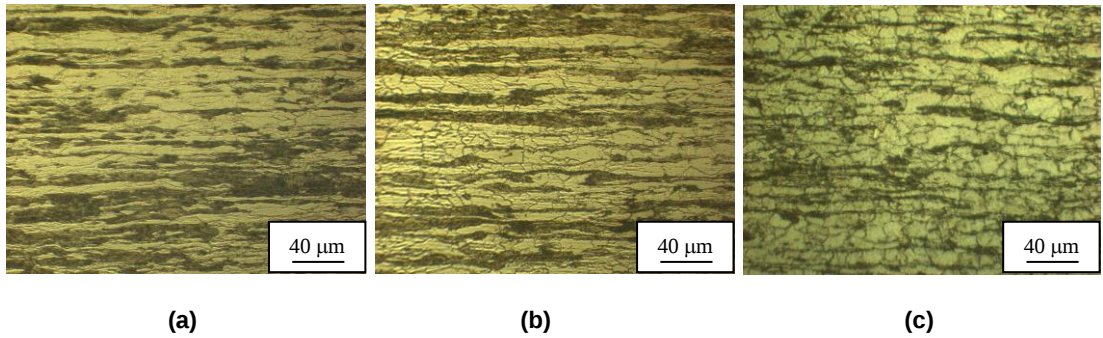


Figure C.5: 21Mn5 56% Red. annealed at 650°C (a) As received (b) 1 h (c) 3 h (400x).

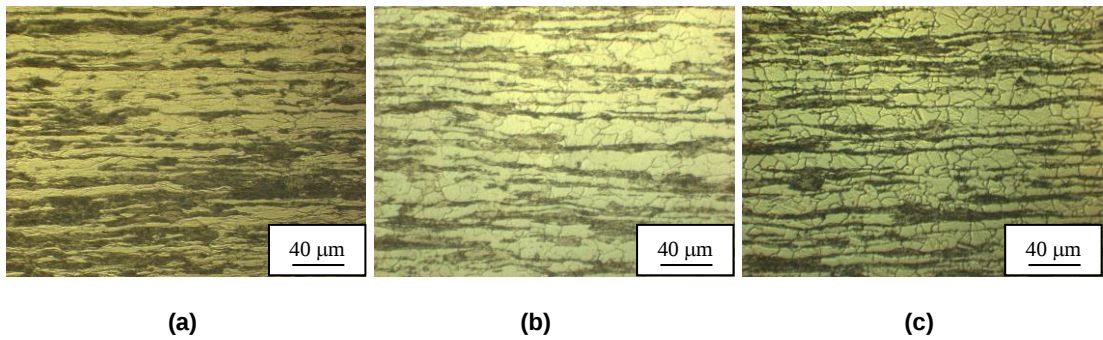


Figure C.6: 21Mn5 56% Red. annealed at 700°C (a) As received (b) 3 min (c) 5 min (400x).

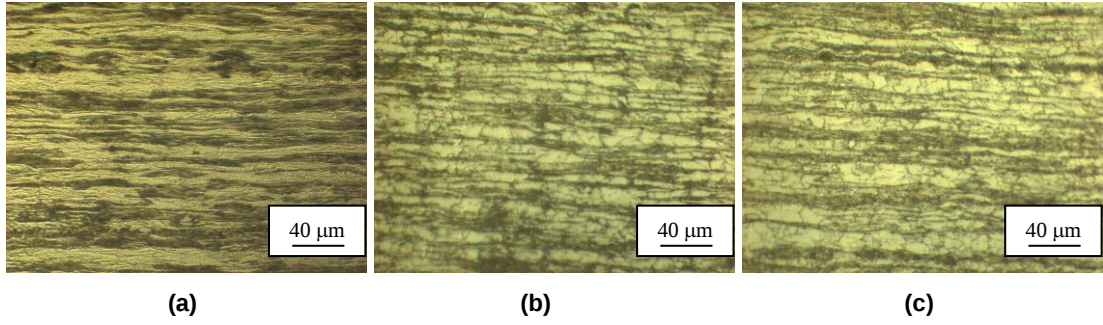


Figure C.7: 21Mn5 73% Red. annealed at 625°C (a) As received (b) 30 min (c) 3 h (400x).

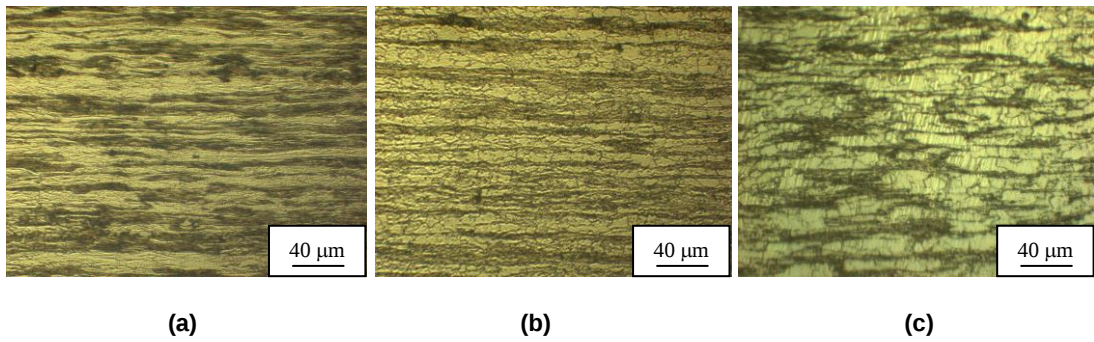


Figure C.8: 21Mn5 73% Red. annealed at 650°C (a) As received (b) 30 min (c) 2 h (400x).

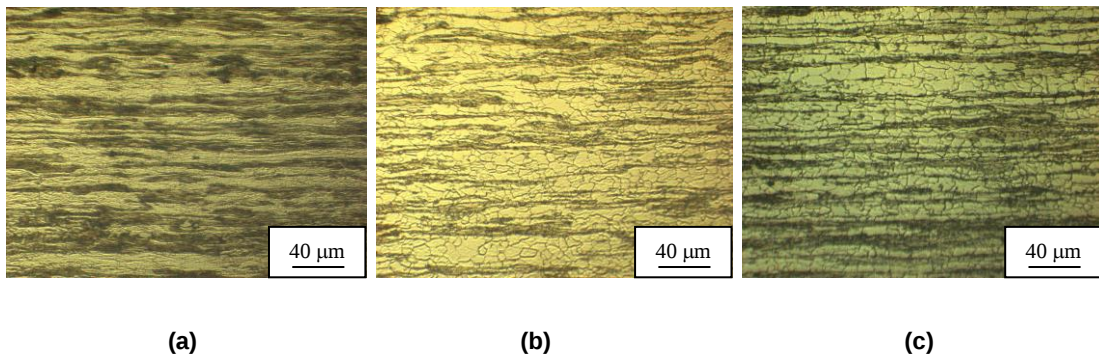


Figure C.9: 21Mn5 73% Red. annealed at 700°C (a) As received (b) 2 min (c) 4 min (400x).

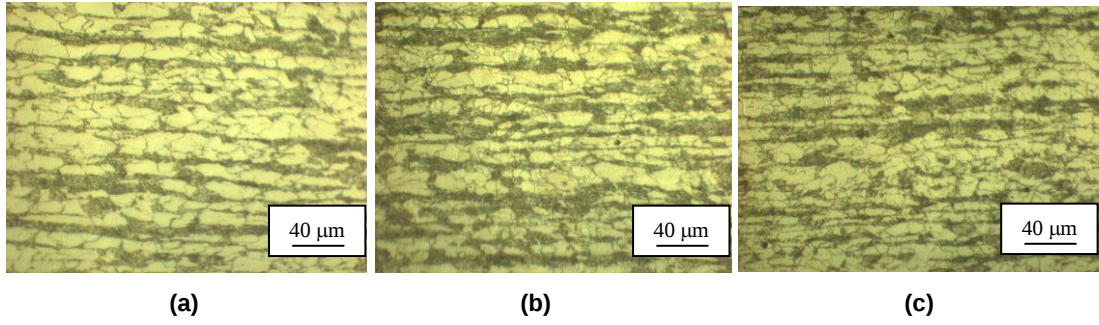


Figure D.1: 20NiCrMo3 33% Red. annealed at 625°C (a) As received (b) 4 h (c) 6 h (400x).

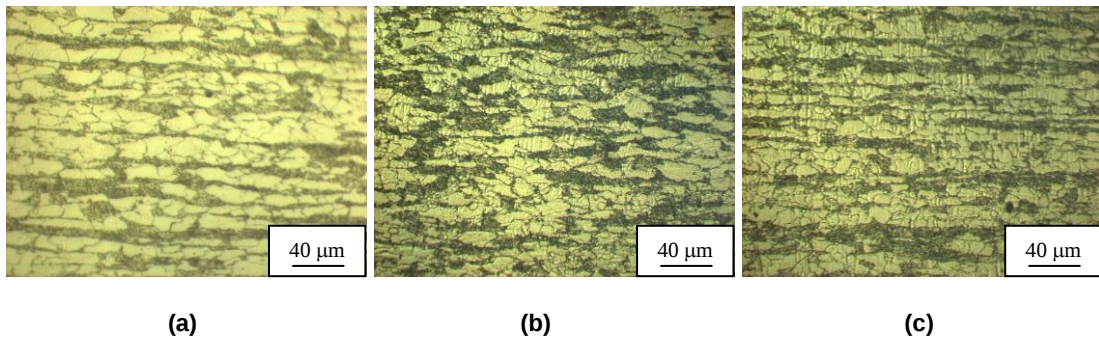


Figure D.2: 20NiCrMo3 33% Red. annealed at 650°C (a) As received (b) 30 min (c) 1 h (400x).

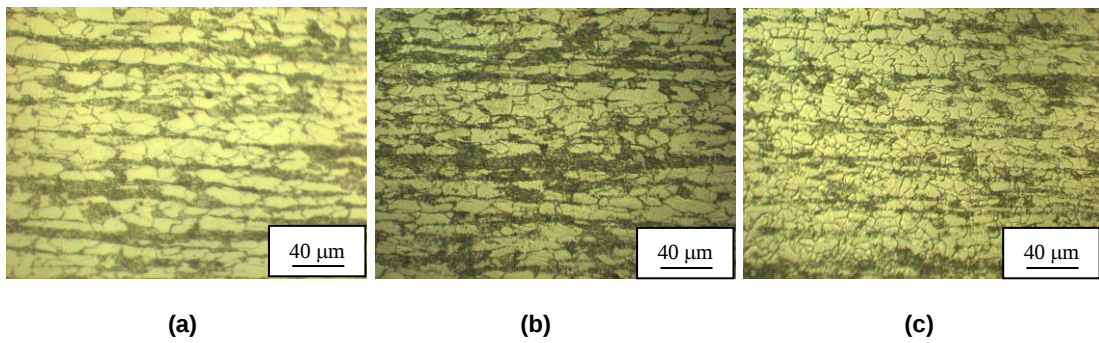


Figure D.3: 20NiCrMo3 33% Red. annealed at 700°C (a) As received (b) 2 min (c) 3 min (400x).

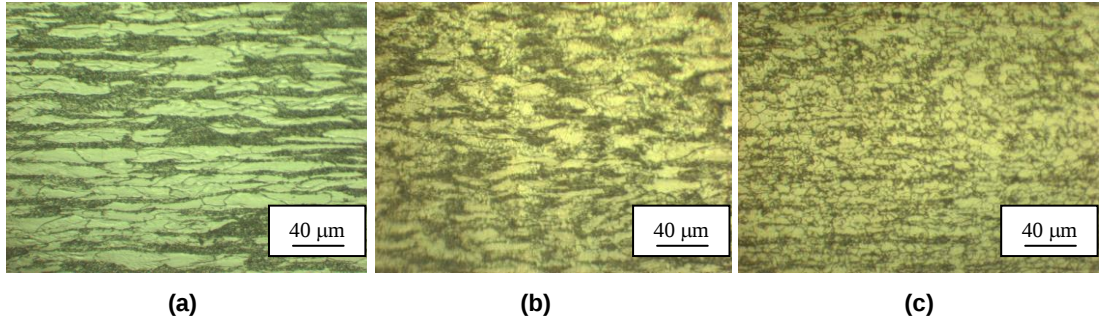


Figure D.4: 20NiCrMo3 60% Red. annealed at 625°C (a) As received (b) 2 h (c) 4 h (400x).

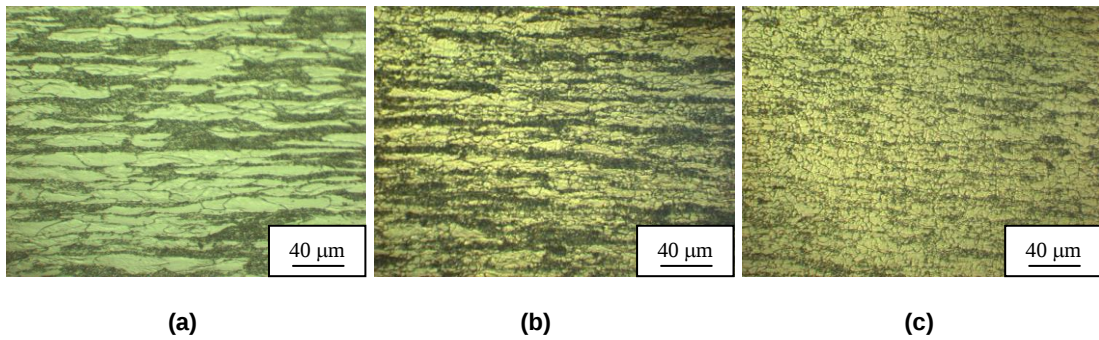


Figure D.5: 20NiCrMo3 60% Red. annealed at 650°C (a) As received (b) 25 min (c) 45 min (400x).

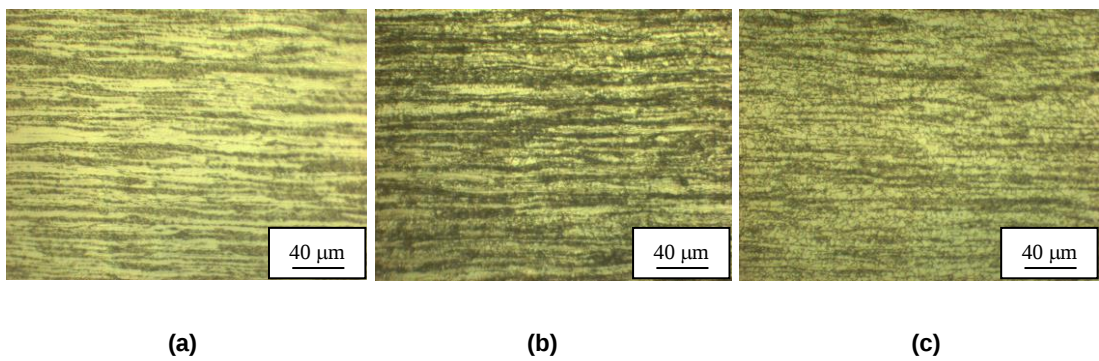


Figure D.6: 20NiCrMo3 79% Red. annealed at 625°C (a) As received (b) 1,5 h (c) 3 h (400x).

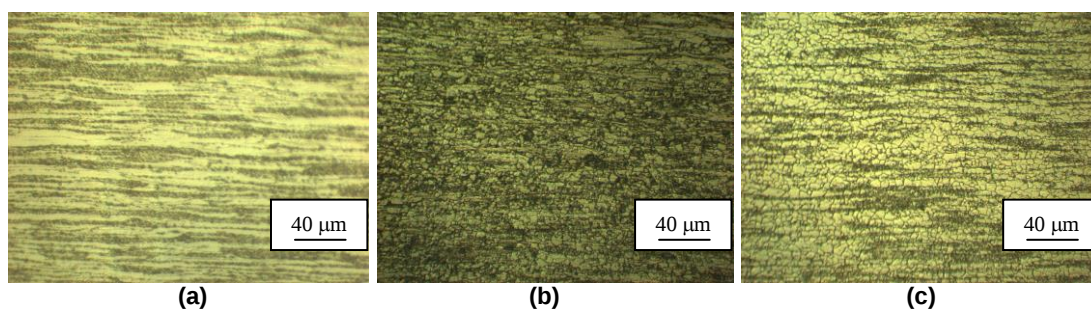


Figure D.7: 20NiCrMo3 79% Red. annealed at 650°C (a) As received (b) 25 min (c) 45 min (400x).

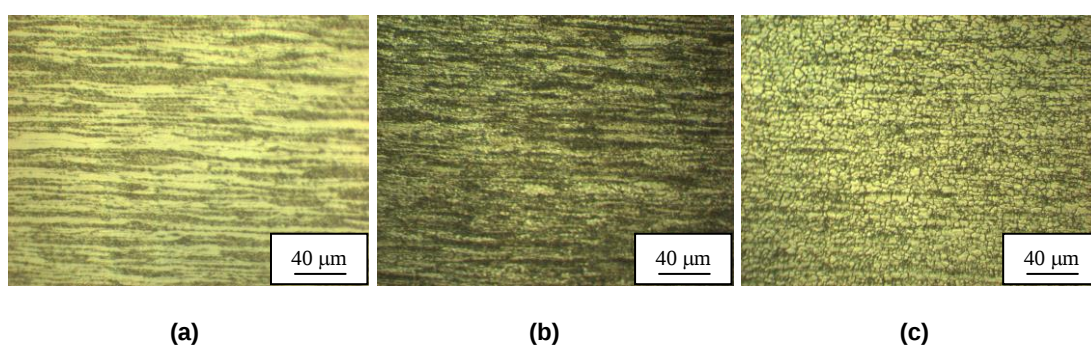


Figure D.8: 20NiCrMo3 79% Red. annealed at 700°C (a) As received (b) 1 min (c) 3 min (400x).

BIOGRAPHY

Çağdaş Uyar was born in İstanbul, Turkey in 1978. He graduated from Özel Ayazağa Işık Lisesi in 1996 and attended Metallurgical Engineering Department of İstanbul Technical University in the same year. He got the degree of B.Sc. on Metallurgical Engineering in 2000 and he started his graduate study at the Material Science program of İstanbul Technical University in 2001. He has completed his military service in May 2004 and he is working in the iron and steel sector at the company Uyar Çelik since the year of 2000.