2. LITERATURE REVIEW

Cast iron is a Fe-C-Si alloy that usually contains minor alloying elements not exceeding 0.1%. But in excess of this limit, it is often used in the as cast condition or after heat treatment. The microstructural aspects of major importance in the cast iron are the chemical and morphological forms taken by carbon and the matrix structure. Important matrix structure includes ferrite, pearlite, bainite, austenite or martensite. Cast irons may be divided into two main groups- gray cast irons and white cast irons including alloy cast irons. Gray cast irons are extensively used as general - purpose engineering materials because of their high thermal conductivity, low modulus of elasticity, ability to withstand high thermal shocks and good machinability. The special purpose white and alloy cast irons are used for applications involving extremes of heat, corrosion and abrasion [Elliott, 1995]. The general purpose gray cast irons from the largest group of casting alloys are generally classified according to the graphite morphology into flake, malleable, spheroidal / ductile and compacted / vermicular types.

2.1 Ductile cast iron - evolution

Continued search by foundry men for an ideal cast iron with mechanical properties equal or superior to malleable iron, led to the invention of DI in which as cast structure contained graphite in the form of almost perfect spheres. It was first made in 1943 at International Nickel Company Research Laboratory, by a ladle addition of magnesium to cast iron. In the 1948 AFS Convention, British Cast Iron Research Association announced production of spheroidal graphite in cast iron by addition of small amounts of cerium as graphite spherodizer. International Nickel Company (INCO) also referred it as spheroidal graphite cast iron/DI consisting of graphite spheroids dispersed in a matrix similar to that of steel. The significant difference between gray cast iron and DI is in the shape of the graphite phase; the matrix can be similar. Ductile cast iron differs from malleable cast iron, as it is usually obtained because of solidification and does not require any heat treatment. The spheroids of DI are more rounded than the irregular aggregate of temper carbon found in malleable cast iron. DIs are inferior to flake irons with respect to physical properties such as thermal conductivity and damping capacity, but have principal advantages of gray cast iron like low melting point, good fluidity and castability, excellent machinability and good wear resistance. It has engineering advantages as of steel like strength, toughness, ductility, hot workability and hardenability. It is
generally used for casting, requiring shock and impact resistance along with good
machinability such as navy shipboard, gears, crank shaft, cams, guides, valves and pump
bodies, shock resisting parts, heavy duty machinery. The matrix structure has great effect on
the properties of DI. According to matrix, DI can be classified as:

- DI with matrix having less than 10% pearlite and rest of matrix ferrite, is called Ferrite
  DI. The amount of ferrite depends upon composition and rate of cooling. This structure
gives maximum ductility, toughness and machinability.
- DI with a matrix, which is largely pearlite, is called Pearlite DI and can be produced by
  normalizing. It is stronger but less ductile than ferrite ductile cast iron.
- The matrix of Ferritic Pearlite DI consists of about 50% ferrite and 50% pearlite.

After inception in 1948, a rapid development in the 1950’s followed increased usage of DI in
the 1960’s. Figure 2.1 shows that its production has been maintained steadily in 1975-85,
despite a fall in the overall iron production in that period. DI has several advantages as an
engineering material including its versatility and high performance at low cost. As illustrated
in Fig. 2.2, other ferrous materials may have superior individual properties suitable for certain
applications but none has the versatility of DI, as it often provides the designer with the best
combination of overall properties [Ductile Iron data, 1990]. It has the castability, surface
hardenability and corrosion resistance similar to those of gray iron with mechanical properties
comparable with steels. They make suitable replacement for steel and flake iron, for example
in pipes [Else, 1993] and are complementary to malleable irons [Jacobs, 1975]. They are less
section sensitive than flake irons but are very sensitive to minor and trace elements, an
influence that can be magnified by segregation during solidification. Close control must be
exercised over composition during production in order to ensure cast structure with acceptable
graphite shape and desired matrix structure, which may range from fully ferrite to pearlite or
bainitic. Figure 2.3 [Elliott, 1995] shows that a much wider range of properties can be
achieved by their heat treatment. Although this adds to the cost, some foundries use heat
treatment to provide a wide range of properties, from an iron of fixed composition. A
comparison of typical microstructures and tensile strengths of different types of DIs is given in
Fig. 2.4. Development of ADI demonstrates yet another direction for the improvement in
properties, which could be achieved by heat treatment of DI.
Fig. 2.1: Cast iron production in the United Kingdom over the period 1960-1985 [Elliot, 1995].

Fig. 2.2: Comparison of characteristics of various types of irons [Elliot, 1995]. (Scale: 1 = worst and 5 = best).

Fig. 2.3: Strength and ductility that can be achieved in DIs in the following conditions: (a) austempered (b) quenched and tempered (c) normalized (d) as-cast bulls eye ferritic (e) as-cast or annealed [Elliot, 1995].
Fig. 2.4: Microstructure and tensile strengths of various types of DIs. [Ductile Iron data for design Engineers, 1990].

2.1.1 Advent of austempered ductile iron

Development of ADI is due to the superiority of the material to compete with existing materials. As compared to the conventional grades of Cast irons, it exhibits better mechanical strength, wear resistance and fatigue strength. [Moore, 1985; Rouns, 1984; Tartera, 1989; Gundlach, 1986; Shah, 1986; Bartosiewicz, 1993; Shanmugam, 1994; Imasogie, 2000; Abdullah, 2011]. It offered relatively low density, lower energy consumption and lower cost involved per unit weight produced. Besides offerings the manufacturing flexibility of the cast product, ADI offers wear resistance strength and damping capacity greater than its competitive materials [Gundlach, 1986]. The first tentative step [Design Award winner, 1978] to apply austempering heat treatment to DIs was taken in 1950s, yet it was not until mid-1970s that ADI components started to be produced in large number. Considerable interest was aroused by the announcement that General Motor has developed a lower bainite grade of ADI which has sufficiently good properties that it could replace the case carburized forged steel for the manufacture of rear axle pinion and ring gear for passenger car [Design Award winner, 1978; Dodd, 1978]. This interest was reinforced by the announcement in China [Yicheng, 1979], that ADI is being used as gear material [Johansson, 1978] and for the production of a wide range of engineering components [Johansson, 1977; Johansson, 1978].

Table 2.1 lists the applications of ADI in different components reported by different researchers time to time.
Table 2.1: Application of ADI [Batra, 2003].

<table>
<thead>
<tr>
<th>Component</th>
<th>Application reported</th>
</tr>
</thead>
<tbody>
<tr>
<td>Crankshaft</td>
<td>Diesel Engine [Watmough, 1984; Forrest, 1985]</td>
</tr>
<tr>
<td>Agricultural Components</td>
<td>Plogh shares and tips [Dodd, 1984]</td>
</tr>
<tr>
<td>Railway Component</td>
<td>Wagon Wheels [Johansson, 1978]</td>
</tr>
<tr>
<td>Sprockets</td>
<td>Heavy Duty chain sprocket [Dodd, 1978; Johansson, 1978]</td>
</tr>
<tr>
<td>Rolls</td>
<td>Wire Mill Rolls [Johansson, 1978]</td>
</tr>
<tr>
<td>Engine</td>
<td>Connecting Rods [Johansson, 1978]; Camshaft</td>
</tr>
</tbody>
</table>

2.2 Development of ADI

Development of Austempered ductile iron (ADI) involves two steps. Firstly, austenitization at a fixed temperature (T_a) for a predetermined time (t_a) followed by austempering at a temperature (T_1) for a fixed time (t_1). The temperature range for austenitization of DI is between 850°C to 950°C for sufficient time extending over 1 to 2 hrs, followed by rapid cooling to a temperature range of 250°C to 450°C and holding it for austempering before water quenching or air cooling. Schematic diagram of typical austempering heat treatment cycle for cast iron is shown in Fig. 2.5. This technique develops a unique material without impairing mechanical properties. The microstructure of ADIs consists of nodules of graphite dispersed in a duplex matrix of ausferrite i.e., ferrite needles and high carbon austenite. ADI can be twice as strong as the standard spheroidal iron at the same level of toughness. It also responds to work hardening effects and exhibits excellent fatigue and wear resistance properties [Zammit, 2012; Azevedo, 1993; Cooper, 1987; Hardening, 1992; Hardening, 1986; Faucher, 1987; Lee, 1988; Palmer, 1986; Robinson, 1984; Sharma, 1984]. During austenitization, the as cast microstructure transforms to austenite, γ° which transforms further during austempering process. In general, austempering occurs in two stages:
Stage I:
Austenite, $\gamma^o \ (C_{y^o}) \rightarrow$ Acicular ferrite, $\alpha$ + High carbon austenite, $\gamma \ (C_{\gamma})$  \( \ldots \) Eq 2.1

Stage II:
High carbon austenite, $\gamma \ (C_{\gamma}) \rightarrow$ Bainitic ferrite + Carbide.  \( \ldots \) Eq 2.2

These transformations have been shown schematically in Fig. 2.6. In Stage I, the matrix austenite, $\gamma^o$ with carbon content $C_{y^o}$ transforms isothermally to acicular ferrite and carbon enriched stabilized austenite, $\gamma$ with carbon content, $C_{\gamma}$. In Stage II, the stabilized austenite, $\gamma$, decomposes to ferrite and carbide. Investigations [Zhukov, 1997; Zhukov, 1971] have shown that there is a preliminary stage prior to Stage I reaction, during which a spinodal decay of austenite takes place. In this, austenite without any incubation period gradually decomposes.

![IT diagram of a processing sequence for austempering](asm-handbook-vol-4-heat-treating-1991)
into two austenites, $\gamma_1$ and $\gamma_2$, one depleted in carbon and the other enriched with carbon shown in Fig. 2.7. This hypothesis was based on thermodynamic computations and has been supported by the findings by Zhukov, 1979. The $\gamma_2$ austenite is stabilized by its high carbon content that has been transferred from $\gamma_1$ austenite and this phase is the genuine retained austenite in the system. The $\gamma_1$ austenite becomes destabilized because of carbon impoverishment and hence during austempering the formation of bainitic ferrite formation starts, which is taken as Stage I reaction. In this stage, fast growth of ferrite plates with lower ausferrite occurs. In between the feathery or acicular structure of bainitic ferrite, austenite becomes progressively more enriched with carbon, hence more stable at low temperatures, especially in the presence of high silicon content which restricts the formation of carbides.
[Riposan, 1989]. The latter is not spinodal carbon rich austenite, but that enriched by the additional carbon obtained from \( \gamma \) in Stage I transformation. This carbon – stabilized austenite is, therefore, not retained austenite like \( \gamma_2 \), but may be termed as the ‘transformation stabilized austenite’. Still both the austenite are considered together [Zhukov, 1992] as these cannot be distinguished separately. Hence Stage I reaction produces acicular ferrite and high carbon stabilized austenite. In an ideal austempering process, Stage I and Stage II would be separated in time so that carbon austenite forming in Stage I reaches a plateau, i.e. maximum and constant value at an austempering temperature for a stipulated interval of time, with negligible or no martensite. The mechanical properties of ADI reach their optimum values during this time “processing window” shown in Fig. 2.6. It is reported that the carbon content of austenite formed in this period can be as high as 2% and this high carbon content of austenite makes it enable even at temperatures as low as -120°C. Stage II reaction begins on prolonged aging with the precipitation of carbides from the carbon stabilized austenite followed by sidewise growth of ferrite plates. The carbides precipitated in Stage II reaction in an unalloyed ADI has been identified as \( \eta \) - carbides and \( \varepsilon \) - carbides [Sidjamin, 1992]. Thus the maximum amount of high carbon stabilized austenite with the minimum amount of the constituents like martensite and carbides, formed in Stages I and II respectively, is desired in this transformation.

The low carbon austenite is the main constituent transforming during Stage I reaction and it

\[
\begin{align*}
T_{sd} &< T \\
T &< T_{sd}
\end{align*}
\]

\[
\text{CARBON CONTENT}
\]

Fig. 2.7: Free energy diagram with respect to carbon content for spinodal decomposition of austenite into two austenites \( \gamma_1 \) and \( \gamma_2 \) [Zhukov, 1979].
remains untransformed and on subsequent air-cooling transforms into martensite, it results in reducing the ductility. This constituent is exaggerated by the alloying addition which delay the Stage I reaction and their segregation during solidification. In Stage II reaction, high carbon austenite disappears as a consequence of carbide formation and there is reduction in ductility [Harding, 1986]. The properties of the products of isothermal transformation during Stage I in unalloyed DI have been investigated by Dymski, 1996. They observed that during austempering, the carbon content of austenite becomes approximately twice that of the equilibrium carbon content at the austenitization temperature. The lattice parameter of the ferrite phase decreases with increasing cooling rates, but increases when formed in upper bainite range. The crystal structure of the austenite phase formed in the upper bainitic region is more perfect than that formed in the lower bainitic region.

2.2.1 Austenitization process

Austenitization is an important step in the development of ADI. In order to achieve better mechanical properties of ADI, the optimization of this heat treatment is important and it requires an understanding of the details of transformation occurring.

2.2.1.1 Austenitization parameters

Austenitization involves holding the DI in a bed of cast iron chips in a furnace maintained at a suitable temperature in the austenite range (generally 850°C - 950°C) to obtain complete transformation of the matrix of DI to austenite and its homogenization in respect of carbon and alloying elements. The alloy is held for a definite minimum time at the austenitization temperature. Austenitization occurs by nucleation and growth process. Austenite can nucleate at ferrite/cementite or ferrite/graphite interfaces. The influence of austenitization temperature [Murakami, 1984], silicon content of the alloy and the prior microstructure (i.e. percentage pearlite and nodule count) have been investigated [Murakami et.al, 1984]. Austenitization of a DI with ferrite matrix have also been investigated [Solenstev, 1974; Ueda, 1977]. Austenitization in such material is controlled by diffusion of carbon from graphite to surrounding austenite. Hence carbon distributes nonuniformly. The influence of percentage pearlite on the time required for initiation of austenite formation has been investigated by Ogi, 1988 and it indicates that nucleation of austenite at the ferrite-cementite interface is more favoured over that the ferrite - graphite interface [Ogi, 1988]. This has been attributed to the lower temperature of metastable transformation as well as kinetic factors relating to the microstructure. Growth of austenite depends upon the diffusion of carbon to austenite, which
may take place from the graphite nodule and the cementite of the matrix pearlite, if any, through graphite/austenite and cementite/austenite interfaces respectively. The diffusion distance is of the order of the interlamellar spacing when austenite is formed in matrix pearlite, but it is of the order of the dimension of eutectic cell in ferritized sample. Thus, the effect of increased pearlite contents on reducing the austenitization time is apparent [Chakrabarti, 1987; Krohn, 1984; Grech, 1990]. The diffusion time required for austenitization depends on temperature, nodule count, matrix structure, diffusivity of carbon, alloy content and its segregation in iron. The mean carbon content of austenite obtained on austenitization of ferritized and other DIs is shown in Fig. 2.8. [Ogi, 1988]. Darwish has also reported similar results (Part II, 1993) as shown in Fig. 2.9. The higher nodule count decreases the size of eutectic cell and thus, the diffusion distance decreases and faster homogenization of carbon in austenite may occur. Voigt and Loper, in 1984 have reported an equation, which has been used frequently, to calculate the equilibrium carbon content of austenite. This equation is given below:

\[ C_y = \left( \frac{T}{420} \right) - (0.17 \times \% \text{Si}) - 0.95 \]  ...Eqn 2.3

where \( T \) is austenitization temperature in degree celsius.

Fig. 2.8: Mean carbon content of austenite affected by austenitization time at 900°C for as-cast (4A) and ferritized (1F) DIs. [Ogi, 1988].
Fig. 2.9: Variation of carbon content of matrix austenite, C_{\gamma}^0 with austenitization time for various austenitization temperatures. [Darwish, Part III, 1993].

The experimental data of Darwish, Part II, 1993 for low manganese DI has shown good agreement with this equation and led him to conclude that alloying elements have little effect on the equilibrium carbon content.

### 2.2.1.2 Austenitization kinetics

Austenitization kinetics has been monitored through the changes in hardness on samples austenitized for various time periods and water quenched subsequently. The change in hardness indicates the volume fraction of martensite formed by quenching of austenite. Carbon content of martensite has been determined by X-ray diffraction study and is used to follow austenitization kinetics. The results are shown in Fig. 2.10 [Darwish, Part II, 1993]. It may be seen that austenite formation follows a sigmoidal curve characteristics of nucleation and growth process. The information has been quantified by observing time $t_j$, at which
austenite formation starts as indicated by increase in hardness and the time, $t_f$, at which austenitization is complete and hardness reaches a maximum. The period $t_f - t_i$ may be plotted in terms of $T$. The carbon content of austenite formed during austenitization affects austemperability and determines the driving force for austempering process. The homogeneity of carbon in austenite results in uniform austempered microstructure. Thus, the austenitization kinetics is important in the context of austempering and must be given special attention. The uniformity of carbon in austenite should also be taken into consideration. Many workers studied austempering process and its effect on mechanical properties [Shamanth, 2012; Shih, 1990; Hayrynen, 1990] without considering the initial carbon content and its distribution in austenite. The optimization of austenitization temperature and time has not received due consideration and no attempt has been made to model austenitization of DI for this purpose.

![Graphs showing variation of volume fraction of martensite and hardness with austenitization time and temperature](image)

**Fig. 2.10:** Variation of volume fraction of martensite (austenite) and hardness with austenitization time and temperature in (a) and (b) Cu-iron (c) and (d) Ni-Cu iron. [Darwish, Part II, 1993].
2.2.1.3 Microstructure control during austenitization

In steels, it has been reported that prior microstructure and holding time has no effect on grain size of austenite at temperatures below 900°C, while grain coarsening does occur beyond 950°C. The resulting austenite grain size at these temperatures depends on the temperature and holding time. Higher austenitization temperature and time lead to coarse austenite. Austenitization time dictates the uniformity of carbon saturation. For example, if $D_c$ is the diffusion coefficient of carbon at austenitization temperature, $T_y$, the diffusion distance, $X_{rms}$ may be estimated as $(2D_c x t_y)^{1/2}$ for a given austenitization time of $t_y$ [Achary, 1995]. Hence, the austenite grains smaller than $2X_{rms}$ should have reached the saturation carbon content, $C_{y^s}$, predicted by the phase diagram. Core of austenite grains larger than $2X_{rms}$ would contain lower carbon and may therefore be unstable. Hence, the time should be large enough to allow complete austenitization and saturation over larger distances of graphite nodule separation.

During austenitization, it is important to control the carbon potential of the furnace atmosphere (or bath) at the equilibrium carbon level in the austenite. If the carbon potential in the furnace is too low, a carbon-depleted layer will be formed at the surface. This layer will have lower strength, hardness and fatigue resistance than the bulk material. Thus a protective atmosphere has to be employed to minimize scaling and decarburization from the surface.

2.2.2 Austempering process

2.2.2.1 Austempering parameters

Austempering process in development of ADI is very critical and requires a precise control over time and temperature of austempering. It is the process which classifies ADI into two broad categories of microstructure (a) fine ausferrite (b) coarse ausferrite. This decides the mechanical behavior of ADI.

2.2.2.2 Austemperability

To obtain fully austempered microstructure, the base iron must possess sufficient ‘austemperability’. Austemperability is a concept similar to hardenability. Both can be expressed in terms of the same critical cooling rate that must be exceeded to avoid ferrite and pearlite formation during cooling to austempering temperature. An alloy is said to have sufficient austemperability for an application if quenching from austenitization temperature to that of salt or oil austempering bath produces bainitic products isothermally, without formation of pearlite or other products of higher temperature transformation. Austemperability or the maximum section size that can be successfully austempered can be determined from the
austenite transformation characteristics of the alloy and quench severity of the austempering bath. There have been several experimental studies relating the austemperability (maximum diameter or section thickness that can be successfully quenched into an ideal quenchant of quench severity (H = ∞) data to the critical austempered diameter for different austempering baths and quench severities [Voigt, 1991; Dorazil, 1987; Harding, 1987]. All these studies [Darwish & Elliot, 1993; Voigt, 1991; Dorazil, 1982] predict a linear relationship between critical austempered diameter and austemperability, but with varying slopes. Dorazil (1982) investigated austemperability of multiple alloyed DIIs. Analyzing their work, Voigt and Loper (1984) have developed a mathematical model relating composition and austempering temperature to critical bar diameter that could be austempered pearlite-free. This relationship is:

\[ d_c = 124 \, C_r^0 + 27 \, (\%Si) + 22(\%Mn) + 16(\%Ni) + 25(\%Mo) - 1.68 \times 10^{-4} \, T_a^2 + 12(\%Cu) (\%Ni) + 62(\%Cu)(\%Mn)+ 88(\%Ni)(\%Mo)+ 11(\%Mn)(\%Cu)+ 1279(\%Mn)(\%Mo)-20(\%Mn)(\%Ni)-137. \]

where, \( C_r^0 \) is the carbon content of the matrix austenite at the austenitization temperature and \( T_a \) is the austempering temperature. This model takes into account the synergistic effects of the commonly used alloying elements. It can be deduced from this expression that the maximum thickness of an unalloyed DI that can be quenched without formation of pearlite is only a few millimeters. Hence, majority of ADI components need to be alloyed for satisfactory austemperability. The presence of alloying elements like nickel, chromium and molybdenum increases the hardenability in steels by delaying the austenite to pearlite and / or bainite reactions and thus permitting more martensite to form for a particular cooling rate, resulting in greater hardness. Elements like vanadium and chromium used to increase the hardenability of steels, form stable carbides in cast irons that not only have an embrittling effect at the cell boundaries but also deplete the matrix of carbon and reduce austemperability. The effects of alloying element on critical austempering diameter for different austempering temperatures are shown in Fig. 2.11.

In general, molybdenum is the most potent element affecting austemperability. It is useful as an austemperability agent for large castings. It works in two ways by itself and by increasing carbon solubility. Manganese, the second most potent austemperability agent, affects austemperability in the same manner as manganese [Morgen, 1987]. Nickel is frequently added in ADI to increase austemperability. According to Rouns and Rundman work on kinetics of austempering, after austenitizing at 927°C followed by austempering at 371°C, the presence of Ni in base iron increase the driving force \( C_r^0 - C_r \) upto 0.6 Wt. % of Ni and further addition of
Ni decreases this driving force. The third most potent austemperability agent is copper, with similar potency for improving austemperability as of nickel and is the least expensive austemperability agent. Its effectiveness as austemperability agent diminishes when its concentration is beyond 0.8% [Shih, 1990].

Fig. 2.11: Effects of alloying elements on the critical austempered diameter at austempering temperatures of 300°C and 400°C [Voigt, 1989].

Copper can be used in combination with molybdenum to improve the austemperability more effectively than by adding them separately [Dodd, 1984; Park, 1984].

2.2.2.3 Austempering thermodynamics

The thermodynamics of austempering depends on the chemical composition of the iron and can be analyzed in the context of the appropriate phase diagram. At the austenitization temperature, the matrix carbon content $C_\gamma$ is obtained from $\gamma / (\gamma + \text{graphite})$ phase boundary at that temperature. Since the mixture of phases in ausferrite at the austempering temperature are in metastable state, the composition of these phases could be obtained from project of $(\alpha + \gamma)$ regions above the eutectoid temperature to the austempering temperature as shown in Fig. 2.12 [Gundlach, 1984].
The driving force for the reaction occurring at the austempering temperature may be identified from a free energy - composition diagram of the iron at that temperature. A schematic free energy - composition diagram along with the corresponding phase diagram is shown in Fig. 2.13 [Hayrynen, 1993]. The diagram shows the driving forces (free energy reduction) for both the Stage I and II reactions. Any change of $C_\gamma$ or the austenitization temperature changes the relative driving force for Stage I and Stage II respectively, for example lowering of austenitization temperature increases the driving force for Stage I of austempering reaction.

To understand the reaction at different temperatures, it is necessary to compare the free energy - composition diagrams at these temperatures. Fig. 2.14 [Achary, 1995] compares the situation at decreasing austempering temperature. A typical ausferrite microstructure is represented by the point G₂ in Fig. 2.14(a) [Achary, 1995] illustrating a purely $\alpha/\gamma$ equilibrium.
If the austempering temperature is low enough to make the carbon diffusion from ferrite difficult, ferrite may become supersaturated with carbon and transition carbide are formed within ferrite. The microstructure will then contain austenite, ferrite and transition carbides within ferrite and can be called austempered bainitic structure. This is represented by the point $G_4$ in Fig. 2.14(b).

If in addition to the kinetic barrier, the austempering temperature, $T_A$, is sufficiently low and the point $T_o$ in Fig. 2.14(c) lies to right of $C_{f0}$, then the conditions are conducive for transformation, based on short range diffusion without major change in chemical composition or a martensitic type transformation, which needs large activation energy. This energy could be supplied at room temperature by applied stress leading to the transformation mechanism.
called SITRAM (strain induced transformation of austenite to martensite). The specific transformation mechanism is very important in analyzing the in-service performance of ADI.

The effect of austempering time is also important for obtaining optimum properties. The proper time should produce a structure free from martensite (too short $t_a$) and carbides (too long $t_a$). At too long austempering time, the enriched austenite has the chance to reduce its free energy by rejecting carbon and thus, inducing precipitation of carbides within austenite. The structure then contains ferrite, austenite and carbides in austenite. This has lower free energy than the original structure, represented by point $G_3$ in Fig. 2.14 (a). If the holding times are extremely long, austenite tends to disappear and ferrite and carbides form the most stable structure with free energy given by $G_5$ in Fig. 2.14 (a).

A schematic isothermal transformation (IT) diagram and the continuous cooling transformation (CCT) diagram of DI are given in Figs. 2.5 and 2.15 [ASM Handbook, Vol. 4 Heat Treating, 1991]. The IT diagram dictates the processing window for successful austempering with respect to $T_A$ and $t_A$.

![Fig. 2.14: Qualitative free energy diagrams at different austempering temperatures (Achary, 1995).](image-url)
2.2.2.4 Austempering transformations and processing window

As dictated by thermodynamics, isothermal transformation takes place at austempering temperature in two stages as described in section 2.2 by equation 2.1 and 2.2.

**Stage I:** In Stage I, as mentioned earlier, the matrix austenite, \( \gamma_0 \), with carbon content, \( C_{\gamma_0} \), transforms isothermally to acicular ferrite and carbon - enriched stabilized austenite, \( \gamma \), with carbon content, \( C_{\gamma} \). During the austempering process ferrite nucleates at or near graphite nodules or as matrix austenite grain boundaries. As these ferrite plates grow, carbon is rejected into the surroundings austenite increasing its carbon content. Thus, in Stage I of austempering, enriched austenite and ferrite grow at the expense of matrix austenite.

The carbon content in the reacted austenite and ferrite in this non equilibrium mixture may be determined from the extended ferrite + austenite (\( \alpha + \gamma \)) field above the eutectoid temperature to the austempering temperature in the equilibrium diagram of the iron as shown schematically in Fig. 2.12 [Gundlach, 1984]. This suggests that reacted austenite can have more than 2% carbon while the ferrite phase may dissolve as much as 0.2% carbon. As ferrite with such high carbon levels cannot exist for long, carbides are expected to precipitate within ferrite, especially at lower austempering temperatures. At higher austempering temperatures, however, carbides are not always present, probably because carbon may diffuse out of the growing ferrite into surroundings austenite. When the austempering temperature decreases, both ferrite
Stage II: In Stage II, when stabilized austenite, \(\gamma\), decomposes to ferrite and carbide, there is a decrease in volume fraction of austenite, \(X_\gamma\), and hence ductility decreases. Because of the very fine scale of this transformation, it can be studied only with high resolution electron microscopy. Stage II is marked by the decrease in the amount of stabilized austenite with time. Thus, from the start of austempering reaction, the volume fraction of stabilized austenite, \(X_\gamma\), increases, goes through a maximum and then decreases.

In an ideal austempering process, Stage I and II would be separated in time so that the volume fraction of stabilized austenite, \(X_\gamma\), would appear as schematically shown in Fig. 2.16 [Moore, 1986]. Stage I goes to completion at time \(t_i\) creating a somewhat stable combination of \(\gamma\) and acicular ferrite for a time period \(t_2 - t_i\), called the processing window. Eventually at time \(t_2\), \(\gamma\) would begin to transform to its more stable products of ferrite and carbide, leading to a steady decrease in \(\gamma\) to zero. In this ideal case, ductility will follow the same pattern as \(X_\gamma\) with well-defined time period \(t_2 - t_i\), where ductility would be maximum and constant. Moore, 1985 suggests that \(X_\gamma\) follows a pattern shown in Fig. 2.17, indicating an overlap of Stage I and Stage II. The situation shown in Fig. 2.16 because both the Stage I and Stage II being nucleation and growth events, Stage II nucleation can be expected as soon as some \(\gamma\) is formed in Stage I, but at a relatively slower rate, depending primarily on \(T_A\) and alloy content. Rundman, 1986 also reported that the segregation of certain alloying element have the effect of stringing out Stage I and Stage II because of different rates at different regions. The processing windows for irons, containing various amounts of manganese in the range between 0.15 - 0.97%, austenitized at 871°C or 927°C and austempered at 371°C narrow down and closes with increasing manganese content. Reducing the austenitization temperatures can be seen to shorten heat treatment times and delay window closure until addition of a higher level of manganese. It has been reported that molybdenum addition do not increase \(t_i\). Increase in silicon level broadens the range of \(t_2 - t_i\) as its presence delays the formation of iron carbides [Dorazil, 1982; Rouns, 1984; Shiokawa, 1978; Dubensky, 1985; Gagne, 1985]. Previous work [Rouns, 1984] has shown that mechanical deformation at the austempering temperature prior to austempering greatly increases the rate of Stage I, reducing \(t_i\). The effect of deformation on time \(t_2\) is not known.
2.2.2.5 Austempering kinetics

The rate at which various transformations occur during austempering and the time selected determines the extent of various reactions. The kinetics of Stage I and Stage II reactions have been followed using hardness measurement. The volume fraction of austenite and the austenite carbon content have been determined by X-ray measurements [Darwish, PartII, 1993; Rouns, 1987]. Stage I kinetics have been studied using measurements of the untransformed austenite. Measurements so far have shown that Stage I kinetics is influenced by austempering temperature, austenitization temperature, amount and type of alloying element. Limited analysis of Stage II kinetics suggests that this is influenced primarily by alloying elements. The procedure adopted for kinetics study of austempering process in Stages I and II and the influence of austenitization temperature and time and austempering temperature have been discussed in following sections for high and low austempering temperatures separately.
(a) Upper bainite reaction: As the Stage I of austempering begins, the hardness starts to decrease and reaches a plateau. Previous studies by Darwish (Part II, 1993) and Rouns (1984) have used the time for the hardness to fall to 100 units above the plateau values as being representative of the time to complete 60 to 80% of the Stage I reaction. In Fig. 2.18, the hardness plateau at an austempering temperature of 316°C is approximately equal to 410 Hv10. The time needed to attain 510 Hv10 or a point 100 Hv10 above plateau is about 5 min. Assuming that the hardness difference between the as quenched matrix and the plateau is proportional to the percentage bainitic reaction and realizing that the total change in the hardness during the first stage is between 250 and 400 Hv10, it is clear that the 100 Hv10 above plateau represents a time required ~ 60 – 80% deformation. Darwish, 1993 who worked on low manganese DI has noted that the alloying of iron with nickel and copper delays the Stage I reaction. He has also observed that the increase in austenitization temperature delays the Stage I of austempering. Increasing austenitization temperature from 900°C to 1000°C increases the carbon solubility in austenite by approximately 0.3%. This change increases the time to complete 60 - 80% of the transformation from 2 - 9 min. Thus a relatively small increase in carbon content slows the Stage I process significantly. Also Darwish, 1993 has reported that changing combined Ni and Cu content from 0.3 to 1.9% only delays the transformations from 2 to 5.5 minutes. Thus the most potent element for delaying the Stage I process is carbon and the austenitization temperature may control this. The work by Rouns (1984) on various iron compositions containing manganese varying from 0.15 – 0.63 Wt. %, has concluded that

![Fig. 2.18: Vickers hardness vs. austempering time for specimens of 0.3% Mn DI austenitized at 927°C and austempered at 316°C [Rouns, 1984].](image)
addition of 0.13 Wt. % of Mn or Mo or Ni increase the time to complete 60 – 80 % of transformation by only 1.53, 1.18 and 0.89 minutes respectively. Yusuf, 2011 in his experiments, stated that the most important matrix elements affecting the rate of Stage I process is carbon, determined by the control of austenitization temperature, $T_y$, and it has been observed to slow down the reaction by a factor 3 - 6 times more than that possible by the addition of Mn, Mo or Ni.

The delaying effect of alloying element addition on the bainite reaction in DI is consistent with the observation made in other ferrous materials. Explanation for these effects include reduced carbon diffusion to the presence of alloying elements; segregation of alloying elements to moving ferrite/austenite boundary by exerting a solute drag effect as well as solute atom partitioning by acquiring solute atom diffusion.

A logical explanation for more rapid Stage I process in a matrix containing lower carbon can be obtained from the schematic free energy composition diagram of Fig. 2.14. It is apparent from this figure that when austenitization temperature $T_y$ is reduced, there is a corresponding reduction in austenite carbon content which results in a greater driving force for the transformation to the metastable product phases, $\alpha + \gamma$, than that for higher $T_y$. The resulting greater driving force may affect the rate Stage I in two ways – firstly, the number ferrite nuclei increases directly and secondly, the activity gradient driving carbon diffusion in austenite in advance of growing ferrite is also increased. Thus, when carbon content of the matrix decreases due to decreasing austenitization temperature, $T_y$, the transformation of bainitic product occurs more rapidly because of a combination of more ferrite nuclei and more rapid ferrite plate growth. The importance of this result is manifested not so much because of the obvious reduction in the needed austempering time but due to the fact that a much more uniform bainitic product results. If the driving force for the transformation increases on the average, it will also increase locally within the segregated volumes in the casting and thereby increase the likelihood of transformation to bainitic ferrite even within the segregated volume where higher alloy content would naturally slow down the transformation. Rouns, 1984 has reported that for an alloy with 0.6 Wt.% manganese there is very little martensite containing inter-dendritic volume in the specimen austenitized at 871°C compared to similar specimen austenitized at 927°C, which contains much larger and more obvious volume of martensite containing inter-dendritic volume. This is because, the decrease of austenitization temperature reduces the carbon content of austenite matrix, which increases the driving force for Stage I process and thus, increases the rate of the Stage I process.
Further insight into the Stage I kinetics may be gained by using the volume fraction of martensite, $X_m$, particularly for the early part of the reaction. This parameter cannot be measured directly but it may be determined by using the mass balance for carbon [LeHouilier, 1971] as follows:

$$C_y = X_a C_\alpha + X_y C_\gamma + X'_{a'} C_{\alpha'} + X_{cm} C_{CM}$$

...Eqn 2.5

where C’s and X’s represent respectively the Wt. % carbon and the weight fraction of the indicated phases of ferrite denoted by subscript $\alpha$, austenite denoted by $\gamma$, martensite denoted by $\alpha'$ and cementite denoted by $C_{CM}$. For the Stage I reactions involving upper bainite, the following assumptions are made:

(i) $C_\alpha = 0$, ii) $X_{cm} = 0$ and most significantly iii) $C_{\alpha'} = C_y$.

At any time during the reaction, the values of $X_y$ and $C_y$ are known and this reduces the number of unknown in the mass balance Eqn (2.5) to just $X_{a'}$, as given below

$$C_y = X_y C_\gamma + X_{a'} C_{\gamma'}.$$ ...

...Eqn 2.6

With the assumption that there is no carbide present, the matrix is then composed of only ferrite, martensite and austenite and thus, $X_{a'}$ could be determined. The time to reduce the volume fraction of martensite to 50%, $t_{0.5}$, has been used to determine the rate at which bainitic ferrite needlessly grow into austenite in the early part of Stage I. The calculations are restricted to upper bainitic structures only, when there is no carbide as discussed in the earlier part of this section. The early stages of transformation depend on driving force. This is relatively independent of alloying additions except through their influence on $C_y$ and $C_\gamma$. It has been shown [Darwish, Part II, 1993] that a decrease in the austenitization temperature or a decrease in the carbon content of matrix austenite increases the driving force for austempering and decreases $t_{0.5}$. Completion of Stage I reaction in the upper bainite reaction can be measured by the decrease in the unreacted austenite volume fraction, $X_{au}$. In alloys with more manganese, there is sharp decrease in $X_{au}$ followed by more gradual decline to a residual value. This has been reported for high manganese iron [Moore, 1987] and this is shown in Fig. 2.19. For low manganese DIs [Darwish, Part III, 1993] there is a continuous decrease in $X_{au}$.

Stage II of austempering involves the dissociation of high carbon austenite to more stable products of ferrite + carbide. The Stage II process can be monitored by the decrease in volume fraction of retained austenite $X_r$. Further information on the kinetics may be gained by normalizing the results of $X_r$ by $X_r^{max}$, the maximum volume fraction of retained austenite. Ratio $X_r / X_r^{max}$ falls from 1 to 0 as the reaction in Stage II of austempering proceeds. It has
been observed that as the austempering temperature increases, the Stage II reaction also increases significantly, but alloy additions may delay the transformation. There does not appear to be any systematic variation in Stage II kinetics with austenitization temperature, as it has been observed for Stage I kinetics.

(b) Lower bainite reactions: Stage I at lower austempering temperature involves the transformation of austenite to a mixture of lower bainite and high carbon austenite. As the diffusion of carbon is restricted due to lower transformation temperature, the morphology of lower bainite develops as carbide precipitation within the ferrite needles.

The early part of Stage I transformation in the lower bainitic reaction cannot be followed using time to reduce the volume fraction of martensite to 50% (i.e., \(X_{\alpha} = 0.5\)) as is used in case of upper bainitic reactions. The carbide formation in lower bainite invalidates the estimate of volume fraction martensite by using equation 2.6. This problem can be overcome by using the time for the austenite carbon content to reach the mid-point value between \(C^*_\gamma\) and \(C^\text{max}_\gamma\) as a measure of initial rate of Stage I transformation. \(C^\text{max}_\gamma\) may be determined from the relationship of carbon in austenite with austempering time. Rouns, 1987 has used this method to study kinetics of Stage I in the lower bainite reaction for a number of irons at austempering temperature of 316°C. Darwish, Part III, 1993 has studied the relationship of this time as a function of driving force and has found its strong dependence on the driving force. He has

Fig. 2.19: Variation of unreacted austenite volume fraction with austempering time for DI with varying amounts of Mn in base iron (\(T_\gamma = 927°C, T_A = 371°C\)).

The early part of Stage I transformation in the lower bainitic reaction cannot be followed using time to reduce the volume fraction of martensite to 50% (i.e., \(X_{\alpha} = 0.5\)) as is used in case of upper bainitic reactions. The carbide formation in lower bainite invalidates the estimate of volume fraction martensite by using equation 2.6. This problem can be overcome by using the time for the austenite carbon content to reach the mid-point value between \(C^*_\gamma\) and \(C^\text{max}_\gamma\) as a measure of initial rate of Stage I transformation. \(C^\text{max}_\gamma\) may be determined from the relationship of carbon in austenite with austempering time. Rouns, 1987 has used this method to study kinetics of Stage I in the lower bainite reaction for a number of irons at austempering temperature of 316°C. Darwish, Part III, 1993 has studied the relationship of this time as a function of driving force and has found its strong dependence on the driving force. He has
observed in his studies on low manganese DI, that this time is relatively independent of alloying additions.

On comparing the Stage II transformation in upper and lower bainitic reaction [Darwish, Part III, 1993; Moore, 1986], it has been observed that the Stage II transformation is delayed for lower bainitic reaction as compared to that for upper bainitic reaction. Previous study [Moore, 1986] has shown that the volume fraction of retained austenite at $10^4$ minutes is strongly dependent on alloying additions, suggesting that Stage II kinetics in the lower bainitic reaction is also strongly dependent on alloying additions. The kinetics study for time period up to 1440 min for low manganese irons by Darwish (Part III, 1993) has shown that the rate of decrease of the volume fraction of stabilized austenite, $X_y$, at 300°C is much slower in copper-nickel alloyed iron than that in the copper-iron, indicating a significant dependence of Stage II reaction on alloying additions.

2.2.3 Role of alloying elements

2.2.3.1 Influence on austenitization

The basic aim of austenitization process in development of ADI is to obtain the transformation of the matrix of DI in to completely saturated and uniform austenite. The matrix is required to homogenize with respect to the carbon and the alloying elements. The presence of carbide forming elements like molybdenum, vanadium, chromium etc. leads to reduction in diffusivity of carbon and hence retards the process of austenitization [Ogi, 1988]. Chakrabarty, 1987 and Krohn, 1984 have reported that increase in pearlite content reduces the time for homogenization of carbon and hence reduces austenitization time. Hence it can be fairly concluded that not only the alloying elements but the matrix of DI also influences the process of austenitization. The elements like nickel, copper, manganese increases the austenitization time [Cooper, 1986; Dorazil, 1982].

2.2.3.2 Influence on austempering process

It is well known that alloying element additions serve to delay the transformation of austenite in DI [Dorazil, 1982; Johansson, 1977; Rouns, 1984]. The effect of alloying additions would depend upon their relative effects on two reactions of Stage I and Stage II of austempering. It is known that the different elements will have different relative effects [Dorazil, 1982]. The influence of silicon and the individual or combined addition of copper, molybdenum, nickel, niobium and manganese on the transformation characteristic have been examined by many
investigators and their results are summarized below [Zammit, 2012; Abdullah, 2011; Dorazil, 1982; Dodd, 1978; Rundman, 1982; Viau, 1987].

Silicon levels in DI in excess of 2.0% are essential to delay Stage II reaction of austempering, sufficiently for developing high toughness ferrite and carbon stabilized austenite in the microstructures [Copper, 1986]. Increasing silicon level from 2.5 - 3.1% has been observed to delay the beginning of Stage II reaction from 1.5 to 4.5 hrs [Park, 1987]. Therefore, control of silicon is very important for the control of austempering kinetics [Rouns, 1984; Takita, 1988, Voigt, 1991; Gagne, 1987; Velez, 1996; Ductile Iron data 1990]. The effect of silicon on the stability of austenite is shown in Fig. 2.20 [Stenfors, 1986]. Excessive silicon content results in an effective decrease in the carbon content of austenite at a constant austenitization temperature [Rouns, 1984; Takita, 1988; Voigt 1991; Gagne, 1987; Velez, 1996; Kurt, 1986; Park, 1987]. The investigations carried out by Velez, 1996 indicate that increasing the silicon content from 1.55-3.48 Wt. % greatly increased the time for completion of Stage I reaction. The volume fraction of bainitic ferrite decreased with silicon content, whereas there is a slight increase in volume fraction of retained austenite.

![Fig. 2.20: Variation of austenite stability range with silicon content in DI [Stenfors, 1986].](image)

However the morphology of bainitic ferrite did not change significantly. The ‘processing window’ gets closed for the silicon contents.
Manganese imparts higher stability to the austenite formed during austenitization and slows down its transformation in Stage I of austempering process. It delays the beginning of Stage I and Stage II reactions and narrows the processing window. This undesirable effect of manganese is particularly observed with treatments made at high austenitization and austempering temperatures [Palmer, 1986]. The reduced processing window for austempering of manganese-alloyed irons is due to its strong tendency to segregate at the inter-cellular region of the DI structures. Decreasing of the austenitization temperature can reduce this effect [Juneja, 1989]. Many researches have suggested an acceptable range for manganese content between 0.25-0.6% for the successful production of ADI [Rouns, 1984; Elliot, 1995; Dodd, 1978; Voigt, 1989; DI Ductile Iron data, 1990; Rundman, 1988; Jenking, 1986; Kovacs, 1986; Mayer, 1986; Dubensky, 1985]. Some studies [Zammit, 2012; Darwish, 1993; Ali, 1994; Bayati, 1995] on austempering characteristics of low manganese (< 0.05 %) and high manganese (0.67%) DI s alloyed with Cu, Ni-Cu, Mo-Cu have been carried out by investigators, which include the variation of mechanical properties as a function of austenitization temperature and the time and temperature of austempering apart from the general relationship between the mechanical properties and the microstructure after austempering. The variation in the amount of retained austenite, austenite carbon content and the unreacted austenite content with austempering temperature has also been investigated. Recent studies [Kazerooni, 1997; Ali, 1994; Ali, 1996] manifest the influence of austenitization and austempering temperature on the austempering kinetics and mechanical properties of austempered Mn-Mo-Cu alloyed DI s.

Molybdenum reduces the rate of diffusion of carbon in austenite and delays the Stage I of austempering but not as much as done by manganese. It segregates strongly during solidification and forms carbides, its much reduced effects on Stage I kinetics makes it a better alloying addition than manganese for controlling austemperability. However, addition must be restricted to about 0.3% because of its strong segregation and carbide forming tendencies during solidification. Published results indicate that molybdenum alloyed DI s with manganese greater than 0.3%, should be considered as Mn + Mo alloyed irons because both the elements contribute synergistically to the resultant austempering response. Molybdenum delays the Stage II reaction. Copper does not affect the carbon diffusion in austenite and the stability of austenite. However, it has been reported that copper suppresses carbide formation in lower ausferrite [Elliot, 1995]. Experiments carried out by Sastry, 1993 have shown that there is slight increase in retained austenite content of ADI alloyed with 1.0% copper. Nickel acts in a similar way as copper and suppresses carbide formation in the lower bainitic structures. Nickel
addition has a pronounced effect on the morphology and fineness of bainite formed at higher temperatures. Up to 1.0% nickel, the structure shows no detectable variation, but at 1.5% and 2.0% nickel, a change in the morphology is clearly seen. Because nickel shifts down the transformation temperature range [Gerlach, 1965], it results in the transformation taking place corresponding to high temperature transformation products. Also if present in excess of 0.5%, slows down the bainitic reaction, thus cause the formation of martensite at cell boundaries [Johanson, 1976]. Segregation of nickel may result in martensite formation during air cooling after austempering treatment and thus reducing the ductility of ADI [Sastry, 1993; Dorazil, 1982]. Antimony and Tin are very effective carbon diffusion barriers [Archary, 1995] and make saturation of austenite with carbon difficult thereby increasing the austenitization time. Thus, these elements may result in large volume of low carbon unstable austenite, which may transform to martensite in the ADI. Buhr et. al, 1989 have studied the effect of residual elements like chromium, vanadium, titanium and antimony, all at levels of < 0.1% either singly or in combinations, on the microstructures and mechanical properties of ADI alloyed with nickel, copper and molybdenum. These residual elements have minor effects on the mechanical properties of ADI.

2.2.3.3 Influence of segregation of alloying elements

In the previous section, the effects of various alloying elements on Stage I and Stage II of austempering have been discussed. Their segregation occurring within the microstructure further complicated the relatively simple picture of their effect on the austempering process. Various workers have discussed this aspect [Dorazil, 1982]. An estimate of the extent of inter-dendritic segregation of alloying elements may be obtained with the aid of the Scheil equation [Moore, 1985]. During the solidification of DI, austenite is formed around the graphite nodules. With advancing solid-liquid interface these alloying elements are rejected by the austenite into the remaining liquid. The distribution of alloying elements takes place as per Scheil equation, predicting the alloy concentration in dendrite, \( C_m \), given by:

\[
C_m = k'C_0'(1-f)^{k-1}
\]

...Eqn 2.7

where \( k' \) is the equilibrium distribution coefficient, \( f \) is fractional distance between dendrite arm center (\( f = 0 \)) and mid-point between adjacent dendritic arm (\( f = 1 \)), \( C_0' \) is the average concentration of alloying element in the alloy. It is assumed that there is no significant diffusion of the alloying element in the solid austenite and there is sufficient diffusion in the liquid to keep the inter-dendritic liquid fairly uniform in alloying composition. Inter-dendritic region will get richer in alloy content as the dendrite arms fatten until eutectic solidification.
finally occurs in the last volume to solidify. Thus concentration of alloy content will increase from the dendrite to the edge with its maximum concentration in the last freezing liquid. The degree of segregation is determined by the equilibrium distribution coefficient ‘k’ defined as the equilibrium ratio of alloy content in the solid to that in the liquid in the range of temperature over which solidification occurs. Fig. 2.21 shows a schematic variation of the concentration of the alloying elements with distance across a dendrite arm when k= 0.5. There is significant variation in the concentration of the alloy element from the center to the edge of the dendrite. Under these conditions, the carbide forming elements like manganese, molybdenum, chromium and magnesium segregate to the last freezing liquid, while non-carbide formers like silicon, nickel, copper, antimony and tin segregate inversely, with their highest concentration in the austenite forming initially at the graphite/matrix interface [Janowak, 1983; Merchant, 1961; Janowak, 1982]. Typical solute distribution tendencies of the elements are illustrated schematically in Fig.2.22 [Kovacs, 1981].

Fig. 2.21: Schematic plot of variation of alloying element with distance across a dendritic arm (Scheil Prediction) [Moore, 1985].

The segregation of alloying elements in ADI is important for two reasons. Due to the variation of their concentration, these elements change the equilibrium position of the phase diagram from location to location in the casting. The required austenitization temperature and time may vary significantly within the casting. Also, the segregation may cause non-uniform carbon solubility in the matrix. Varying carbon content in turn causes a non-uniform rate of austempering reactions in different locations over microscopic scale. Severe segregation may create such conditions in the casting, that in certain locations, the austempering reaction is overdue while in certain others, the reactions may not have even started. Segregation of alloying elements in ADI casting is strongly influenced by the nodule count which in turn is
controlled by section size and cooling rate. A large casting, which solidifies slower than a small casting, usually has a smaller nodule count and larger cell size. However, if the nodules counts can be increased by special inoculation or mold cooling technique, the extent of solute segregation may be reduced significantly.

2.3 Microstructural evolution in ADI

The following parameters of microstructure of DI have important bearing on the development of ADI. It is established that a predominantly pearlitic matrix is more easily austenitized. A matrix free of proeutectic or eutectic carbide with at least minimum prescribed pearlite/ferrite ratio should be maintained in the base iron for effective austenitization. Nodule size and nodule count are important, since one of the means of minimizing segregation of alloying elements in DI is by increasing the nodule count and reducing the nodule size with improved inoculation or alloying with rare earth elements. At high nodule count, higher manganese level may be tolerated. High nodule count also reduces silicon segregation leading to higher silicon content in the cell boundary enabling austempering reaction to be more uniform and accelerated [Kovacs, 1986]. Graphite morphology is important in determining the microstructure resulting in ADI. Better than 80% nodularity of nodules is required for ADI. A uniform distribution of small and spherical graphite nodule and high nodule count is essential for uniform austempering process. Any transition of graphite morphology from nodular to vemicular or coral results in increased graphite/austenite contact area, increased ferrite content in as cast structure, increased carbon content of austenite during austenitization and decrease in the ratio of reacted austenite to unreacted austenite after austempering treatment [Riposan, 1990].
Distribution and morphology of graphite can be determined metallographically by using simple comparator charts, magnetic techniques and ultrasonic velocity measurements. The solidification rate of DI influences its austemperability through its effect on the nodule size, the dendrite size, solute segregation and the matrix structure. For castings of complex shape with significant thickness variation, mechanical properties may vary significantly from section to section. After the austempering treatment, some characteristics of as cast structure remain as such, since the graphite morphology, the dendrite size and the solute segregation patterns do not change significantly [Kovacs, 1986]. In addition, it has been reported [Ueda, 1986] that for identical compositions, bainitic zone (both B_s and B_f temperatures) in the TTT diagram is greatly shifted to the left in relation to the pearlite zone in irons with faster solidification rate (cast in sand mold). This makes it easier to avoid formation of pearlite during austempering to that extent. Also, the degree of segregation decreases with the increase in solidification rate of the metal.

2.3.1 Effect of austenitization temperature

Decreasing the austenitization temperature influences the austempered microstructure by:

(i) refining austempered structure so that the length of bainitic ferrite needles, which corresponds closely to the parent austenite grain size, is decreased and their number and uniformity of distribution increases,

(ii) decreasing the volume fraction of retained austenite,

(iii) reducing the amount of large blocky structure,

(iv) reducing the amount of martensite formed, especially in the cores of blocky austenite and inter-cellular regions.

These general observations on microstructural features made by many workers [Darwish, part II, 1993] are consistent with the kinetics described above. Since lowering of the austenitization temperature increases the driving force for Stage I, more bainitic ferrite nuclei may form leading to a more rapid transformation. Consequently, at a given austempering temperature and time, the transformation is relatively more rapid in an iron austenitized at lower temperature and there will be less unreacted austenite. The faster the transformation, the more is the carbon rejected from the growing needles and more stable is the retained austenite. This austenite is less likely to transform to martensite on cooling or on the application of stress. It has been observed that the $X_{\text{uq}}$ decreases as the austenitization temperature decreases [Darwish, Part II, 1993]. However, at higher austenitization temperature, the parent austenite contains more carbon. Hence, the austenite present at any stage of the austempering process may be stabilized.
even if a lower volume of bainitic ferrite has formed. But some amount of martensite is still present in the austempered microstructure. The reason for this apparent contradiction is that the spacing of the bainitic ferrite is too large to complete carbon saturation of the blocky austenite areas in a given austempering time. If it is assumed that the diffusion of carbon in austenite is described by:

\[ D = D_0 \exp^\left(\frac{Q}{RT}\right) \]  

...Eq 2.8

where \( D_0 \) is \( 1.4 \times 10^7 \mu m^2s^{-1} \), the activation energy \( Q \) is 133.5 KJ mole\(^{-1}\), \( R \) is the gas constant and \( T \) is the absolute temperature, the diffusion distance can be calculated using the \( \sqrt{2Dt} \). Consequently it should not be surprising that most of the area contains martensite when temperature of austenitization is high. The formation of martensite, which is detrimental to mechanical properties, can be prevented by alloying sufficient time for carbon diffusion or by Stage II reaction.

2.3.2 Effect of austenitization time

Previous study has shown that provided the austenitization time is sufficient to produce an austenite structure with an average carbon content, \( C_r^o \), close to the equilibrium value, austenitization time has no influence on austempering kinetics, if there is no significant growth in austenite. Too low an austenitization time may cause the entire matrix to transform to austenite but the carbon distribution may not be homogenized. Consequence of these would be the non-uniform austempering transformation at different regions of the matrix. Long austenitization times would be expected to result in larger austenite grains, which may retard the austempering kinetics and produce larger bainitic ferrite needles resulting in coarser austempered structure. Such observations have been reported [Darwish, Part II, 1993] for low manganese DI and for an iron with composition as 3.3% carbon, 2.65% silicon, 0.35% manganese, 0.0085% sulphur, 0.01% phosphorous, 0.04% magnesium, 1.6% nickel and 1.6% copper [Delia, 1998].

2.3.3 Effect of austempering temperature

At higher austempering temperature, upper bainitic ferrite needles form first close to graphite nodules and then in the inter-cellular areas during Stage I transformation. This observation has been made for low manganese DI [Darwish, part III, 1993], for various DI with varying manganese content from 0.15 to 0.6 Wt. % by Moore (1986) and in nickel-molybdenum alloyed DI by Shih (1990) also. It has been suggested that delay of bainitic transformation in inter-cellular regions is due to the presence of segregated alloying as \( (C_r^{max} - C_r^o) \) decreases. At
the highest austempering temperature, the centers of these austenite areas transform to martensite on cooling to room temperature. At lower austempering temperature, bainitic ferrite nucleation is more rapid for the lower bainite transformation, resulting in a more acicular structure of finer needles distributed more uniformly throughout the structure. Lower bainite structure resembles tempered martensite. The amount of retained austenite is reduced and appears as silvers between the needles and is more difficult to locate in the microstructure.

2.3.4 Effect of austempering time

As described earlier, the stabilized austenite volume fraction, $X_Y$, is strongly dependent on austempering time. A typical variation of $X_Y$ with austempering time shows that $X_Y$ increases initially with increasing austempering time, reaches a maximum and then decreases with further increase in time of austempering. The position of the maximum depends, among other factors, on austenitization and austempering temperature, $T_y$ and $T_A$, respectively. Optimum properties are achieved with around 60 – 80 % bainitic ferrite and 20 – 40 % retained austenite in the microstructure. Such structural condition is generally achieved using austempering time within processing window. Onset of Stage II that involves dissociation of high carbon austenite to ferrite and carbides, results in rapid deterioration of mechanical properties. Because of submicroscopic nature of the Stage II, determination of maximum austempering time is difficult. However, the start of decrease of $X_Y$ or decrease in difficulty can be taken as start of Stage II. Previous study has emphasized the need to select the correct austempering time in order to optimize the mechanical properties whatever the austenitization and austempering temperatures may be.

2.4 Mechanical properties

Mechanical properties of ADI, or any metal/alloy for that matter, strongly depend on its microstructural features and may be changes to a large extent by changes in microstructure. The purpose of austempering of DI is to achieve the desired combination of mechanical properties by subjective it to an appropriate temperature – time cycle, designed to obtain specific microstructural features. Thus an understanding of the relations between mechanical properties and microstructure of ADI is essential to fully appreciate the process to which DIs are normally subjective. A brief discussion on 0.2% proof stress, ultimate tensile strength, % elongation, as a function of alloying elements, optimization of time and temperature during austenitization and austempering temperature and time has been included in the following
subsections. Dependence of mechanical properties of ADIs on austempering temperatures has been studied as shown in Fig. 2.23 [Kovacs, 1986].

2.4.1 0.2% Proof stress

The yield strength in the present context is the offset yield strength determined by the stress corresponding to the inter-section of the stress strain curve and a line parallel to the elastic part of the curve offset by a specified strain [Dieter, 1989]. In India, strain offset of 0.2% is generally used and so it is called 0.2% proof stress. The yield strength obtained by an offset method is commonly used for design specification because it avoids the practical difficulties of measuring the elastic or proportional limit. Significant changes have been observed in strength with changing austempering temperatures, which are accompanied by significant microstructural changes, namely austenite volume fraction (and therefore ferrite volume fraction) and mean particle size of ferrite and austenite phase [Hayrynen, 1990]. A fundamental analysis of microstructure and properties requires describing the yield strength as a function of the measurable variables because the yield stress defines the onset of plastic deformation [Olofsson, 2011].

Hayrynen, 1990 carried out an analysis of the yield strength, $\sigma_y$, in terms of the particle size of the ferrite and volume fraction of austenite, determined by X-ray diffraction. The dependence of yield stress, $\sigma_y$, on the volume fraction of austenite, $X_p$, is given by:
\( \sigma_y = \sigma_0 + B' X_y \quad \ldots \text{Eqn} \ 2.9 \)

For the specimen with identical particle size but varying \( X_y \), \( B' \) is a constant. This equation manifests a rule of mixtures [Riposan, 1990]. The famous Hall-Petch relation equation 2.10 gives the yield strength for the specimens with identical \( X_y \) but varying particle size, \( d_a \):

\[ \sigma_y = \sigma_0 + A d_a^{-1/2} \quad \ldots \text{Eqn} \ 2.10 \]

where, \( \sigma_0 \) and \( A \) are constants. In ADI, both \( d_a \) and \( X_y \) are expected to contribute to a changing \( \sigma_y \). Combining equations 2.9 and 2.10 we get,

\[ \sigma_y = \sigma_0 + A d_a^{-1/2} + B' X_y \quad \ldots \text{Eqn} \ 2.11 \]

Hayrynen, 1990 and Hamid Ali, 1994 have found that the coefficient \( A \) is much larger than the volume fraction coefficient \( B' \), implying that particle size changes are much more important in determining yield stress than a change in austenite volume fraction. Comparison of the \( \sigma_y \) values at two austempering temperatures of 371°C and 410°C demonstrates the small effect of \( X_y \) on \( \sigma_y \) [Hayrynen, 1990]. It has been reported that despite large differences in \( \sigma_y \), \( X_y \) changes very little, while the value of ‘\( d_a \)’ changes by a large amount. Thus, it is clear that the scale of microstructure is very important in determining the strength. Hall – Petch plot of yield strength, \( \sigma_y \) vs. \( d_a^{-1/2} \) for the clean alloy studied by Heyrynen (1990) is shown in Fig. 2.24, in which the yield strength is observed to increase dramatically with decreasing particle size.

![Fig. 2.24: Hall Petch plot of yield strength vs. \( d_a^{-1/2} \) (ferrite (211) particle size) for data from clean alloy [Hayrynen, 1990].](image)

During the transformation of austenite to the austempered product, the size of product is dictated in large part by the rate at which carbon is able to diffuse away from the growing ferrite. At lower austempering temperature, where the diffusion of carbon is slower, the system generates more ferrite plates so that the reaction may progress. Thus, the size of the product becomes finer and more obstacles for dislocation motion are created resulting in an increase in
the yield strength. As austempering temperature decreases, another potential factor contributing to both decreasing $d_a$ and increasing $\sigma_y$, is the presence of increasing amount of iron carbide. Earlier works [Rundman, 1988; Gutierrez, 1995] have documented the presence of increasing amount of carbide as the austempering temperature decreases. Direct observations of iron carbides under transmission electron microscopy and indirect result from x-ray diffraction show that $X_r$ decreases with decreasing temperature in fully austempered structure. The effect of this factor has not been investigated separately and it is understood that it is included in the effect of particle size in equation 2.11 The change in strain energy with austempering temperature and its effect on yield strength has not yet been evaluated.

2.4.2 Ultimate tensile strength (UTS)

Tensile strength is easy to determine and is a reproducible property. It is useful for the purpose of specification and for quality control of a product. Extensive empirical correlations between UTS and other properties are available for steels [Sharma, 1996]. However, Hayrynen, 1990 studied the properties of a clean austempered ductile iron and compared the results of this alloy with similarly treated alloys with higher alloy content (specifically manganese). They reported that increasing the purity of DI resulted in a significant improvement of tensile strength. Fig. 2.25 [Hayrynen, 1990] shows the variation of tensile strength with manganese content for alloys austenitized at 927°C and austempered at 371°C. The pure material studied by Hayrynen, 1990 has 0.03% manganese, while the other results were taken from earlier work of Moore, 1987. Figure 2.25 indicates that elimination of manganese from the alloy enhances strength. Such improvement is probably due to the absence of the negative effects of

![Fig. 2.25: Variation of UTS and % high carbon austenite, RA, with Wt. % Mn in specimens austenitized at 927°C followed by austempering at 316°C [Hayrynen, 1990].](image-url)
manganese segregation in the last freezing liquid of the casting. However, the clean alloy has poor austemperability. Krishnaraj, 1992 has studied the effect of nickel, molybdenum and copper, individually and in combination on UTS of ADI. Figure 2.26 [Krishnaraj, 1992] shows the effect of nickel content on UTS. The increase in nickel decreases the strength at low austempering temperature. It has been reported that the presence of molybdenum along with nickel does not have pronounced effect on UTS; rather molybdenum tends to neutralize the adverse effect of nickel to the extent that the reduction in UTS at high nickel content is only marginal in the presence of molybdenum at low austempering temperature. But high level of addition of molybdenum by itself may be to lowering of strength. Therefore, a balanced addition of nickel and molybdenum has been recommended. Copper addition does not affect UTS appreciably and its effect is only cumulative to that of nickel.

![Fig. 2.26: Variation of peak UTS value with nickel content [Krishna Raj, 1992].](image)

The UTS of three ADI with (i) 0.09% nickel (ii) 0.3% Molybdenum and (iii) 0.9% nickel, 0.6% copper, 0.3% molybdenum have been compared [Shih, 1991]. In general nickel – copper – molybdenum alloyed iron exhibited poor UTS, a fact attributed to micro shrinkage and intercellular compounds observed in the inter-cellular regions of this alloy. UTS of low manganese (0.037%) – copper and low manganese (0.016%) – nickel – copper irons austempered under various conditions have been studied [Hamid Ali, 1994] as a function of austempering time and temperature as well as austenitization temperature. The UTS shows optimum value corresponding to processing window. He has emphasized on the need to select appropriate austemperering time. The results indicate that the austempering temperature has a
significant effect on UTS. It has been noted that a lower austenitization temperature may also require a shorter austempering time. Ishihara et al., 1987 studied the tensile strengths of ADI at elevated and low temperatures. They noted that austempering at temperatures up to 300°C does not change the tensile strengths. When austempering temperature is increased to 400°C, there is a sudden drop in tensile strength and it becomes only half of its original value for austempering at 500°C.

Hayrynen, 1990 studied the austempering of ADI castings with larger sections size castings and has reported that most of these castings contain an acicular bainitic ferrite like constituent and pearlite at the centre of the castings, where there are high concentrations of silicon, nickel and copper. Hardness and ultimate strength are lower in ADI castings with larger sections as compared to those of the tensile bars austempered after removing from the large casting. This is attributed to higher transformation temperature of ADI castings with heavy sections (due to segregation) than those in castings having relatively smaller sections size.

Another study on ADI castings of large section size, made by Faubert et al., 1991, reveals that as the solidification rate decreases with increasing section thickness of castings, the corresponding tensile strength decreases independently of the matrix microstructures.

The results on mechanical properties of ADI indicate that there is deterioration in mechanical properties due to the formation of pearlite because of low austemperability. The presence of pearlite reduces the strength of the lower bainite bainitic irons.

2.4.3 Impact strength

The impact strength of ADI is measured generally by Charpy impact tests, which indicates the energy absorbed in fracturing the specimen. It is important to realize that fracture energy measured by the Charpy test is only a relative energy and cannot be used directly in design equations. The fracture surfaces of Charpy test specimens are examined to determine the type of fracture.

Dakwood, 1986 observed that the notch toughness of ADI is less than that of most wrought steels. Shih, 1991 reported that ADI alloyed with nickel and molybdenum shows significant improvement in the impact strength.

Lin, 1995 have studied the effect of segregation of alloying elements on the Charpy impact energy of ADIs. Four alloys containing either 0.4% manganese or 1% copper or 1.5% nickel or 0.4% molybdenum have been melted separately, poured into casting bars of three different sizes of 3, 15 and 75 mm diameter. The distribution and the degree of segregation of certain
elements have been quantitatively analyzed under electron microscope. The segregation of alloying elements with increasing diameter of the casting bars has been discussed in section 2.2.3.3. The irons containing either nickel or copper or manganese that have been austempered to complete the first stage of bainite formation, show approximately the same impact values for samples of similar diameters. The iron containing molybdenum having no bainite but untransformed retained austenite in the inter-cellular regions has shown lower impact values as compared to those containing either nickel or copper or manganese. Also it has been reported that the impact strength for these specimens with less segregation is greater than those with greater segregations.

Impact test has been used to study the influence of heat treatment parameters on toughness in ADI [Abdullah, 2011; Darwish, Part III, 1993, Moore 1987]. Previous studies [Shih, 1991; Hamid Ali, 1994] have shown that the contour maps of volume fraction of retained austenite and impact energy as a function of austempering conditions coincide. At short austempering time, austenite phase formed is not sufficiently enriched in carbon to prevent martensite formation during cooling to room temperature. At long austempering times, the Stage II reactions results in ferrite and carbide, under both these conditions impact strength would be low. Corresponding to austempering times within the processing window, the impact energy is maximum, because of the presence of maximum volume fraction of high carbon retained austenite. Increasing the austenitization temperature is known to increase the volume fraction of retained austenite in ADI and correspondingly impact energy also increases at lower austempering temperatures. But at higher austempering temperatures, there is decrease in the impact energy even for higher austenitization temperatures. Because the austenite forms large blocky areas, the centers are insufficiently saturated with carbon to prevent the formation of martensite. The above discussion emphasizes that optimum impact energy is obtained when irons are treated within the processing window and that a critical volume fraction of retained austenite exists. The volume fraction of austenite formed at lower austempering temperature is lower as compared to that formed at higher austempering temperature and it undergoes transformation to martensite on plastic deformation. Consequently, the impact energy should increase with increase in the amount of retained austenite [Aranzabal, 1994]. However, evidence was provided for the operation of the TRIP mechanism (transformation of austenite to martensite) at higher austempering temperature resulting in increasing impact energy with increasing volume fraction of retained austenite.
2.4.4 Percentage elongation

The measurement of ductility is of interest in two ways. Firstly, the designer may assess from it in a general way, the ability of the metal to flow plastically before fracture. A high ductility indicates that the material is “forgiving” and is likely to deform locally without fracture should the designer err in the stress calculation or in the prediction of severe loading circumstances. Secondly, the ductility serves as an indicator of changes in processing conditions or impurity level. Ductility measurements may be specified to assess material, “quality” even though no direct relationship exists between the ductility measurements and performance in service.

It has been reported that the purity of iron results in an increase in ductility [Hayrynen, 1990]. Figure 2.25 shows that the % reduction in area decreases, with increasing manganese content for alloys austenitized at 927°C and austempered at 371°C. The likely reason for this behavior is the absence of negative effects of manganese segregation in the last freezing region in the casting.

Krishnaraj, 1992 has studied the effect of addition of nickel individually in the range of 0.5 - 2.0% and in combination with (a) molybdenum at levels of 0.05 and 0.1 % (b) copper at the levels of 0.5 and 1.0%, on the % elongation of ADI. It has been observed that the ductility is affected considerably by nickel additions. At a lower austempering temperature of 350 °C, the elongation increases from 6 % in unalloyed iron to 10 % in an iron alloyed with 1.0% nickel due to the effect of nickel limiting the bainitic carbide formation. Further increase of nickel to 2% brings about reduction in elongation. High nickel content causes differences in the rates of bainitic transformation at different locations of an austenite cell due to segregation.

Darwish (Part III), 1993 in his study on low manganese DI has concluded that reducing the austenitization temperature at higher austempering temperature enhances the ductility which is because of faster austempering process and less amount of martensite formation. The percentage elongation increases with increase in austempering temperatures, which is as a result of the change in morphology of bainite from lower bainite obtained at lower austempering temperatures to upper bainite obtained at higher austempering temperatures [Johansson, 1978]. The relationship between % elongation and austempering time has been studied [Darwish, Part III, 1993; Johansson, 1977]. The maximum ductility is observed for austempering time corresponding to the processing window. For austempering time lower than that corresponding to the start of processing window, the ductility is low because of the presence of martensite in the structure. For larger austempering times, longer than the end of the processing window, the ductility is low because of Stage II reaction leading to formation of
carbide by dissociation of retained austenite of Stage I. Previous studies [Yan, 1995; Shih, 1991] have shown that high carbon retained austenite promotes ductility in ADIs.

Bayati, 1995; Bayati, 1997 carried out austenitization of 0.67 % Mn, 0.25 % Mo and 0.25 % Cu at 920°C for 120 minutes and austempering in a range of temperatures for different times in order to study the mechanical properties of DI. All these satisfy the high strength grades of standard ASTM A897M: 1990, but fail to satisfy the higher ductility grades. This is attributed to the overlapping of the Stage I and Stage II reactions and the occurrence of the transformation induced plasticity mechanism during the deformation, particularly in irons austempered at higher temperatures.

Bahami, 1994 on the basis of their studies involving the effect of pearlite content on the mechanical properties of ADI concluded that the ultimate tensile strength, 0.2% proof strength and elongation decrease with increase pearlite content in the microstructure. However, a smaller amount of pearlite can be tolerated in an alloyed ADI compared to that in an unalloyed one while satisfying the standard specifications.

2.5 Machining

Machining is defined as a production process in which a harder metal cuts a softer metal and the metal is removed in the form of chips (swarf) by a plastic deformation process. The deformation temperature and the force significantly contribute to the quality of the process. Temperature affects the cutting tool material and the forces effect the power and strength needed to perform the process [Coromant, 1994]. There are two general ways to machine described so far by researchers - orthogonal and oblique cutting. Orthogonal cutting has cutting edge perpendicular to the direction of cut and oblique cutting involves cutting edge at an angle to the feed direction. Orthogonal cutting represents a two-dimensional type of cutting. The general principle in orthogonal cutting is the sum of rake angle, clearance angle and wedge angle is a right angle. In orthogonal cutting, the resultant force is normal to the machined surface of the work piece. Shaping is a machining operation which is used to generate flat surfaces on the work piece by reciprocating motion of the tool against a securely mounted work piece. Shaping is a perfect example of orthogonal cutting [Boothroyd, 1965].

2.5.1 Machining ductile cast iron

Cast iron machining has been noteworthy in establishing metal cutting theories by eminent researchers such as Boothroyd, 1965; Shaw, 1986; Armarego, 1969. The history of cast iron
Machining has been long and extensive with significant gains during the industrial revolution. Gray cast iron is widely used in the automotive industry for many years in manufacturing automotive components such as engine blocks, cylinder heads, differential housings, shafts, flywheels, brake drums and disks [Coromant, 1994]. The development of cast iron machining is associated with advancement in tooling and machinery. With the introduction of new cutting tool materials such as Polycrystalline Cubic Boron Nitride (PCBN), Ceramics and other metal matrix composites, the cutting tools are able to survive in adverse cutting conditions. The machinery has advanced significantly offering wide range of speeds, array of spindle options and multiple axis machining. The demand for higher productivity, lower manufacturing costs and better quality of products has led to development of high speed machining in cast iron.

Cast iron is an alloyed material where iron and carbon are the main elements which affect the material property of iron. Cast iron also has higher amounts of silicon. The fraction of carbon decides the hardness levels in iron.

According to the Fe - C phase diagram, Fe - C alloy having less than 2% carbon is classified as steel and more than 2% as cast iron. The rising demand for materials having high strength-weight ratio and customized material properties has led to evolution of new variety of iron such as Ductile Cast Iron (DCI) and Austempered Ductile Iron (ADI) [Polishetty, 2008]. As the name suggests, DCI and ADI have better ductility, higher modulus of elasticity and improved material properties when compared to age old gray cast iron [Coromant, 1994].

The common observation in DCI and ADI is the carbon which exists in its allotropic form “graphite” and is nodular shaped. Graphite exists as flakes in gray cast iron. Silicon is used to control the size and shape of graphite in DCI and cast iron. Due to the presence of higher amount of carbon in DCI and ADI, these materials have an advantage in lubrication and heat dissipation during machining compared to normal cast iron. Graphite present in cast iron helps to dispose heat from the tool / work piece interface and reduces the friction on the chip/tool interface during machining [Rourke, 2003]. The common challenges faced in DCI machining are rapid wear out of the cutting tool insert as DCI has higher amount of silicon which increases the abrasiveness. The presence of carbon in DCI and ADI is also disadvantages in certain machining circumstances as it causes hindrance in predicting the tool behaviour and surface quality. Although PCBN cutting tools are popular and extensively used for gray cast iron machining, the use of PCBN for DCI especially for high speed machining is problematic as the reaction between the high ferrite content and PCBN due to chemical dissolution at high temperatures leads to rapid wear of the cutting tool [Coromant, 1994].
There exists different grades of DCI based on combinations of proportion of ferrite and pearlite i.e. all ferrite, all pearlite and ratios of ferrite and pearlite matrix. The higher the amount of pearlite, the better is mechanical properties such as strength and hardness, elongation but machinability is reduced drastically. There are noticeable variations in machinability of DCI and gray cast iron. Machinability of iron depends on factors such as graphite size and distribution, composition, ferrite/pearlite ratio, cooling rate from eutectic through eutectoid temperatures and presence of external and internal inclusions [Griffin, 2002]. Research studies on gray cast iron machining by Griffin, 2002 reveals that higher tool wear rates are associated with lower free carbon concentrations and higher combined carbon, higher manganese and tin content; and higher concentration of hard inclusions.

There is no correlation between the wear rates and graphite structure in case of cast iron machining [Griffin, 2002]. Work done by Bhattacharyya et. al, 1969 on cast iron machining state that there is no effect of coolant on cast iron machining as the flake shaped graphite present in the iron, lowers the shear strength and assists in lubricating the tool / work piece interface producing discontinuous chips. Powder metal residue along with the discontinuous chip is common during cast iron machining. The use of cutting fluids in cast iron machining leads to thermal cracking of the tool, especially for ceramic tools, as the sudden cooling of the tool from high temperatures during machining leads to thermal cracks and breakage of the cutting edge [Bhattacharyya, 1989]. Ductile cast iron produces a continuous chip during machining. Continuous chip with built-up edge is quite common in ductile cast iron during the final stages of machining. The chip formation undergoes a process known as adiabatic shear which was explored further through extensive research by Komanduri on high speed machining [Komanduri, 1995]. In an adiabatic shearing process, chip is produced as a result of thermal softening of the shear plane zone. According to Recht and Komanduri, initiation of cut takes place in a material at the onset of plastic deformation reaching a certain limit. If the rate of plastic deformation is low, then the potential for strain hardening along the shear zone is low and vice-versa at increased speeds. The strain hardening effect causes localized heat in the shear zone and contributes to the thermal softening of the work piece in the shear zone and thus causes the chip to slip rather than shear along the shear plane [Komanduri, 1995; Recht, 1964].

Based on the works of Nakayama, chip formation does not start from the tool / workpiece interface but initiates at the free surface of the work piece material. Nakayama also suggests the buildup of the stress normal to the rake face and crack formation which propagates along the shear plane reaching the tip of the tool [Nakayama, 1974]. Shaw termed such cracks as “gross cracks”. For a ductile material, micro cracks are developed around the tool/chip
interface and these micro cracks initiate the process of strain hardening that leads to adiabatic shearing process which has been discussed by Komanduri in his research. As a result of strain hardening, the gross crack extends from the free surface to a point in the shear plane where the rate of strain hardening is greater than crack propagation and leads to arrest of the crack formation process [Shaw, 1986]. Strain hardening through plastic deformation is a common phenomenon in DCI and ADI. During machining, plastic deformation results in cold working of the surface layer. The depth of cold worked layer depends on the ductility of the material [Astakhov, 2010].

Lattice crystal defects such as dislocations, which are irreversible deformation caused on a microscopic scale effect strain hardening. Variations in local stress fields cause dislocations leading to rearrangement of the lattice structure as the dislocations propagate [Degarmo, 2003]. As a result of increase in dislocation density of the material, strain hardening occurs in order to prevent nucleation of new dislocations. This resistance to dislocation can also be considered as resistance to plastic deformation. Plastic deformation is due to the work done on a material or energy imparted to the material. In case of machining, the energy applied is of high magnitude and in quick time, which not only moves the existing dislocations but also encourages new dislocation by jarring and working the material sufficiently enough [Degarmo, 2003]. Ghani, 2002 studies on machining DCI using ceramics say that the surface finish of the workpiece is not influenced by the tool wear but speed, feed and depth of cut play an important role to influence the surface finish [Ghani, 2002].

Most of ADI machining is done under dry conditions because of the graphite in ADI with its small coefficient of friction acts as a good lubricant and hence, increasing the tool life and reducing the tool wear [Matsuoka, 2003]. According to Kacal, 2011, the \(\gamma\)-pool (the massive austenite which exists on the boundary of the eutectic cell) average area in ADI structure, as a factor which affects the tool wear in drilling. The flank wear of the drill decreases with the decreases in \(\gamma\) pool average area and hence, it is important to decrease the \(\gamma\) - pool in order to improve the machinability. ADI tool wear phenomenon is characterized by extreme crater wear located very close to the cutting edge which destabilizes the cutting edge and leads to fracture of the crater lip. The unusual combination of abrasive and adhesive wear mechanisms is due the special ausferrite microstructure of ADI [Klocke, 2007].

2.5.2 Machinability factors

It is considered that the word machinability was first used in the 1920 referred to the speed/tool life relationship. Although extensive research was carried out on machining, researchers are
still elusive about an assignable material property to judge machinability [Armarego, 1969]. Machinability of a material is defined as the ability of a material, to allow machining efficiently consuming less power, producing good surface texture and without excessive tool wear or failures. Machinability of a material depends not only on its material properties and microstructure but also on proper selection and control of process variables.

Machinability factors play an important role in effective and efficient running of the metal cutting operation. Machinability factors include machining conditions such as cutting parameters - speed, feed and depth of cut; cutting tool material properties and wear, material properties of work piece, cutting fluids, surface texture, chip formation, rigidity of the machine tool, Metal Removal Rate (MRR) - power required to machine, manufacturing cost and productivity. There are four factors which determine the machinability of a material i.e. surface texture and integrity of the machined part, tool life, force and power requirements and chip control.

2.5.2.1 Cutting forces
Cutting forces play an important role in machining as they initiate the process of chip formation and control the flow of chip and thermal gradient involved in machining. The role of cutting forces is explained using a Hooke’s law, where stress is directly proportional to strain. The stress-strain curve indicates that up to a certain limit of stress the material is elastic and obeys proportionality. On increasing the stress, the material loses its elasticity and slips into the plastic zone undergoing plastic deformation and breaks away from the parent material. Similar is the case with chip formation, the cutting forces should reach a specific level to initiate the chip formation and the minimum force where the plastic deformation starts is known as the specific cutting force. The cutting forces acting along the direction of tool travel (Fc) and thrust force (Ft) are measured using a dynamometer supported by charged amplifier.

In any machining operation, the shear strength and cutting forces are related as

\[ F_c = t \times S_o \times \tau_s \times f \]  \hspace{2cm} \text{Eqn 2.12}

where, \( f = \zeta - \tan (\gamma_0) + 1 \)  \hspace{2cm} \text{Eqn 2.13}

where \( t \) is depth of cut in mm; \( S_o \) is the feed in mm and \( f \) is form factor, which is a constant. Form factor \( f \) is determined from the tool geometry and chip thickness using equation 2.13, where \( \zeta \) is the chip reduction coefficient and \( \gamma_0 \) is the orthogonal rake angle. Cutting force
analysis for a specific machining operation is required in order to have a balanced and positive cutting action with respect to the cutting edge strength and action [Boothroyd, 1965]. The cutting force mainly consist of shear force ($F_s$), axial cutting force ($F_c$) and frictional force ($F_f$). The cutting forces are mostly compressive and shear in nature and depend on the rake angle, cutting fluids, tool and workpiece material, cutting geometry and area of contact between the chip and rake face. Specific cutting force is the initial force required by the tool to initiate plastic deformation or chip formation [Coromant, 2004]. High cutting speeds generates high temperatures in the flow zone which leads to lower values of specific cutting force.

2.5.2.2 Cutting speed
The cutting speed to be chosen depends on factors such as machine tool used, stiffness, precision, damping capacity, numerical control response, contouring capabilities at high feed rates, chip disposal, coolant flow, dynamic balance in case of rotating cutters, part requirements, number of parts, type of machining operation, ratio of non-cutting time to cutting time, safety and economics [Coromant, 2004]. For constant depth of cut, longer tool life is obtained for the best possible combination of low feed rate and low speed.

2.5.2.3 Tool wear
There are two basic forms of tool wear: flank wear and crater wear. Flank wear is most common in machining hard and difficult to machine materials. Tool wear can also exist in different categories such as abrasion, adhesion, diffusion, fatigue, oxidation, deformation, creep and chemical wear high temperature and pressure involved at the cutting edge. The phenomenon generates an additional edge at the tip which is known as Built-Up-Edge (BUE) [Coromant, 2004]. For a bcc structure such as ductile material, the formation of BUE is a result of high plasticity region in front of the tool rake face within the contact length. According to Astakhov, a BUE is avoided by allowing the cutting process to take place at optimum temperature [Astakhov, 2010]. BUE in most of the cases is considered to be a negative factor in machining, as it changes the tool geometry - rake angle, reduces the cutting edge strength and leads to inefficient cutting. BUE formation indicates the possibility of more tool wear and therefore, less machinability.

Heavy built-up-edges leads to excessive heat generation and high frictional force against the flow of chip along the chip-tool interface resulting in high cutting force [Boothroyd, 1965]. Another way, in which BUE is detrimental to tool life, occurs during the cooling of the tool tip at the end of the cut. As the coefficient of linear expansion of steel is approximately twice that
of a carbide tool material, cracks may be introduced in the tip where the attached portion of built-up edge has contracted, during cooling, by a greater amount than the carbide [Armarego, 1969].

2.5.2.4 Surface integrity

Surface integrity is defined as the set of properties of an engineering surface that affect the performance of the component in service. The set of properties include surface finish, texture and profile; fatigue, corrosion and wear resistance; adhesion and diffusion properties and also in service properties such as frictional resistance, strain resistance, thermal emissivity, surface tension, optical properties etc. [Davim, 2010].

Surface integrity plays an important role in the performance, reliability and longevity of a machined component when subjected to stress and temperature, in service [Batzier, 1998]. Surface integrity involves study and control of two factors, surface roughness and surface metallurgy. The surface metallurgy entails investigation of the possible alteration in the surface layers after machining such as plastic deformation and residual stress distribution which are mainly generated in the final stages of machining. Work piece microstructure has significant effect on the surface quality of the machined surface. The grain size and its orientation influence surface roughness. Finer grains are always desirable [Coromant, 1994]. The surface integrity produced by different machining operations can significantly affect machinability. According to the work done by Bhattacharyya, 1989, on cast iron machining using ceramic cutting inserts, almost similar to ADI machining, the surface texture improves with an increase in cutting speeds and depends on contact between the tool and work piece.

At high cutting speed, sufficient heat is generated in the shear zone leading to thermal softening of the work piece which increases the tool / work piece contact resulting in reduced friction at the tool / work piece interface. Hence, better surface finish exists on the machined surface at higher cutting speeds. At lower cutting speeds, there is an intermittent contact between the tool and the work piece leading to trapping of hard abrasive particles at the tool / work piece interface, which at a later stage rub against the fresh machined surface [Bhattacharyya, 1989]. Surface alterations such as plastic deformation, micro cracking, phase transformations, micro hardness tears and laps related to built-up edge formation, residual stress distribution provide detail description of measuring methods available for surface integrity [Davim, 2010]. Micro cracking is a common surface defects while machining brittle materials whereas for ductile materials the high temperature and pressure during machining causes healing of these cracks. Material removal process causes structural changes to the work
piece surface in the form of plastic deformation due to the action of cutting forces and friction of tool flank [Astakhov, 2010]. Built-up edge on the tool is the primary cause of poor surface finish on the machined surface especially for a ductile material. Climb milling is avoided in the experiments as ADI being a cast material has a hard and abrasive top layer which, result in excessive tool wear and also tool failure. In face milling, the cutter is mounted on the machine spindle in such a way that the axis of rotation of the cutter is perpendicular to the work piece surface. The cutting action is due to combination of rotation of the tool at given rpm and movement of the work piece along a straight path at a linear speed.

2.5.3 Introduction to milling

Milling is a metal removal process where a rotating tool having multiple cutting edges removes material by moving along the axis of the work piece. The specialty with a milling cutter is that it produces multiple chips in one revolution of the tool depending on the number of cutting edges on the tool. Popular milling approach include: face milling and end milling. Face milling is employed in the second part of this research on ADI machinability involving study on phase transformations. Milling is an intermittent cutting process where each tooth produces a chip of variable thickness. In case of face milling, the undeformed chip thickness increases to the maximum value at the center of the travel and decreases towards the end of the tooth engagement. The initial and final size of the undeformed chip depends on the width of the work piece with respect to the cutter diameter and symmetry of the cutter axis with the work piece [Armarego, 1969]. In milling, the torque and power are not limited by the work piece strength, rather it is the strength of the cutter spindle which is important.

2.5.4 Earlier studies on ADI machinability

Machining ADI using machining techniques applicable to an as-cast or ductile cast iron, leads to failures in machining. A proper study into the material properties and metal matrix of ADI leads to conclusions of successfully machining ADI. There has been no comprehensive reasoning on machining ADI suggesting that it can only exhibit good machinability properties prior to the heat treatment. Machining post heat treatment leads to SIT caused by strain and thermal energy generated in the shear zone due to the cutting action. This phenomenon is a consequence of rapid change in microstructure from acicular ferrite to a completely difficult to machine microstructure, martensite. The fear of the material getting work hardened has made the machinist cut the material to complex, deep and long cuts prior to austempering. Some of the solution applicable to the above problem is to machine the material despite the formation of
hard martensite by using an ultra-hard cutting tool at low cutting speeds and increases feed rates or using different machining approaches or completely remove martensite from the material by heat treatment process or prevent the commencement of martensite formation. Many have tried to machine ADI as steels or as-cast DI and have been unsuccessful [Matsuoka, 2003]. Because of such failures, a myth has developed that ADI is not machinable. When the material properties and metal matrix of ADIs are taken into consideration, ADI can be successfully machined as shown in Fig.2.27.

![Figure 2.27: Showing relative machinability percentage of materials in comparison.](image)

Problems in machining the material have been brought about by inappropriate machining practices. Recently, a new variant of ADI is introduced known as Machinable Austempered Ductile Iron (MADI), which has high strength compared to DI and better fatigue properties compared to ADI (grade 900). MADI has unique properties exhibiting lower machining forces than DI and ADI [Druschitz, 2003]. Alloying elements also play a significant role in machinability of ADI. Carbide forming elements such as Molybdenum and Manganese significantly reduce tool life as the carbides hinder the machining process. Prior research in machining recommends using molybdenum as low as possible in the presence of higher levels of manganese.

When ADI is subjected to machining operations like drilling, turning and milling, the high normal loads acting at the periphery of the cut lead to localized phase transformations at the surface layer, in advance of the tool. The ductile and tough austenite undergoes a phase transformation forming martensite which relatively hard, brittle and difficult to cut. When machining ADI, the graphite particles determine the cutting forces and surface roughness and
the microstructure determines the tool life. When machining ADI with high ferrite content, the machinability is reduced as the ferrite sticks to the clearance face of the tool and causes poor surface finish [Seker, 2006].

Variation in the microstructure of work piece material due to cutting action has considerable effect on the machinability of the material and tool wear. Machinability definitions are connected to the factors effecting machining like the metallurgy, chemistry, heat treatments, additives, inclusions, surface skin, cutting edge, tool holding, operation and machining conditions [Boothroyd, 1956; Smith, 1995]. Macro inclusions are undesirable as they are abrasive and hard and lead to tool damage. Micro inclusions like FeO and MnO are also undesirable as they cause deflections. Silicates are desirable inclusion as they are softer at high temperatures and form a protective layer and retard tool wear. Sulphur is added to improve machinability as it increases the shear angle and chip curl, reduces the chip thickness and acts as a lubricant on the interface. Lead is also a good additive [Kesani, 2005].

ADI is not as easily machined as pearlitic or ferritic DI, but is comparable to RC30 hardened steel in metal removal rates. Because of the Austempering process, ADI has a different metal matrix than as-cast or conventional quench and tempered DI. The matrix consists of acicular ferrite and carbon-stabilized Austenite, also known as ausferrite.

Understanding the properties of the materials that make up this matrix is the key to understanding why it machines differently than other materials. When a high normal force is applied to ADI, a strain-induced phase transformation occurs on the surface of the part [Armarego, 1969]. The force exerted by the tool during cutting, drilling, or turning can cause a localized phase change in the material in front of the tool. Austenite on the surface undergoes
a transformation to martensite, which is harder and more brittle than the ausferrite structure. Therefore, while machining ADI, this transformation right in front of the tool face makes the material more difficult to machine, Fig. 2.28. Due to the hardness of the material, and the strain-induced transformation in front of the tool face, a deeper cut is required to improve tool life.

### 2.5.5 Effect of alloying elements on machinability of ADI

Some alloying elements, including copper, nickel, molybdenum and manganese increase the hardenability of ADI. Each behaves differently during the heat-treat process. Carbide-forming elements such as molybdenum and manganese tend to segregate toward cell boundaries during the casting process and cause carbide formation in the metal matrix. Carbides can be very detrimental to the machining process and will significantly reduce tool life. The formation of carbides can be reduced or eliminated by adjusting the use of alloying elements. Molybdenum should be reduced as much as possible, particularly in the presence of higher manganese levels. Copper and nickel don't seem to have a detrimental effect on machining, as these elements do not form carbides. Discontinuous chips are one of the benefits of machining ADI [Cemal, 2005]. Unlike steel, which creates a continuous coil like chip while it's being machined, ADI's discontinuous chips are easy to handle and recycle. This ease of handling proves especially beneficial for highly automated machining centers, because the small chips won't clog the equipment.

### 2.6 Electrical discharge machining- introduction

Now a days industries are facing problems in manufacturing of components due to several reasons such as the complexity of the job profile or may be due to surface requirements with higher accuracy and surface finish or due to the strength of the materials. Since it is very difficult to find tool material, which is sufficiently hard and strong to cut materials like stainless steel, austempered ductile iron, titanium and high strength temperature resistant alloys ceramics etc. In addition, manufacturing of complex shapes with better surface finish, precise tolerances and higher production rates in such materials by traditional methods is even more difficult. The solution to these problems lies in the fact that a process may be designed for machining of the materials where the physical and metallurgical properties of the materials do not pose any challenge to the machining process. In non-conventional machining methods, there is no direct contact between the tool and the work piece; hence, the tool need not be
harder than the work piece. Electrical discharge machining (EDM) is one such non-conventional method of machining of conductive materials. In this process, the material is removed by a succession of electrical discharges, which occur between the electrode and the work piece. There is no direct contact between the electrode tool and the work piece. These are submerged in a dielectric liquid such as kerosene or de-ionized water. Its unique feature of using thermal energy to machine electrically conductive parts regardless of hardness has been its distinctive advantage. EDM is a thermo-electric process in which material removal takes place through the process of controlled spark generation. The advantage with EDM is its ability to machine complex cavities, small parts with sharp internal or external radii and fragile materials or work-pieces requiring the creation and modification of very thin walls. However, the occurrence of tool electrode wear is unavoidable and is a very critical issue since tool shape degeneration directly affects the final shape of die cavity. The electrical discharge machining process is widely used in the aerospace, automobile, die manufacturing and moulds industries to machine hard metals and its alloy.

2.6.1 Basic principle of EDM process

The basic principle in EDM is the conversion of electrical energy into thermal energy through a series of discrete electrical discharges occurring between the electrode and work piece immersed in the dielectric fluid. The insulating effect of the dielectric is important in avoiding electrolysis of the electrodes during the EDM process. A spark is produced at the point of smallest inter-electrode gap by a high voltage, overcoming the strength dielectric breakdown strength of the small gap between the cathode and anode at a temperature in the range of 8000 to 12,000°C. Erosion of metal from both electrodes takes place there. Duration of each spark is very short. The entire cycle time is usually few microseconds (\(\mu s\)). The frequency of pulsating direct current supply is about 20,000-30,000 Hz is turned off. There is a sudden reduction in the temperature, which allows the circulating dielectric fluid to flush the molten material from the work piece in the form of microscopic debris. After each discharge, the capacitor is recharged from DC source through a resistor and the spark that follows is transferred to the next narrowest gap Fig. 2.29. The cumulative effect of a succession of sparks spread over the entire work piece surface leads to erosion, or machining to a shape, which is approximately complementary to that of the tool.
EDM spark erosion is the same as having an electrical short that burns a small hole in a piece of metal it contacts. With the EDM process, both the workpiece material and the electrode material must be conductors of electricity. The EDM process can be used in two different ways. Firstly a pre-shaped or formed electrode (tool), usually made from graphite or copper, is shaped to the form of the cavity. It is to reproduce the impression of electrode by feeding it vertically down and the reverse shape of the electrode is eroded (burned) into the solid workpiece, secondly, computer to follow a programmed path to erode or cut a narrow slot through the workpiece to produce the required shape [Wang, 1999].

2.6.2 Mechanism of material removal in EDM process

EDM is a non-traditional process that is used to remove metal through the action of an electrical discharge of short duration and high current intensity between the tool (electrode) and the workpiece. There are no physical cutting forces between the tool and the workpiece. EDM has proved valuable especially in the machining of super tough, hard and electrically conductive materials such as the new space age alloys. Negatively charged particles (electrons) loose from the cathode surface and move towards the anode under the influence of the electric field force. During this movement in inter-electrode gap, the electrons collide with the neutral molecules of dielectric. In this process, electrons are also detached from these neutral molecules of the dielectric resulting in more ionization [Abbas, 2007].

The ionization becomes so intense that a very narrow channel of continuous conductivity is established. In this channel, there is a continuous flow of considerable number of electrons
towards the anode and that of ions towards the cathode; their kinetic energy is converted into
heat energy, hence heating of anode due to the bombardment of electrons and heating of
cathode due to the bombardment of ions, takes place. Thus, it ends up a momentary current
impulse resulting in a discharge, which may be an arc or a spark. The spark energy raises the
localized temperature of the tool and work piece to such a high value that it results either in
melting, or melting as well as vaporization of a small amount of material from the surface of
both the electrode at the point of spark contact [Ozgidik, 2005]. In fact, due to evaporation of
dielectric, the pressure in the plasma channel rises to a very high value and it prevents the
evaporation of superheated metal. As soon as the pulse off time starts, the pressure drops
instantaneously allowing the superheated metal to evaporate. The amount of material eroded
from the work piece and the tool will depend upon the contribution (in the form of KE) of
electrons and ions respectively. A servomechanism controls the movement of the tool electrode
towards the work piece. The sparking takes place over the entire surface of the work piece
hence the replica of the tool produced on the work piece [Ozgidik, 2005].

2.6.3 EDM process parameters

The machining performances depend on various EDM parameters (variables). The EDM
parameters can be categorized into two groups:
1. Electrical Parameters
   (a) Polarity     (b) Peak current   (c) Pulse duration   (d) Power supply voltage
   (e) Pulse-on time  (f) Pulse-off time
2. Non-electrical parameters
   (a) Rotational speed of electrode   (b) Injection flushing pressure
Adjustable parameters are always considered as critical parameters.
Although the non-electrical parameters are less significant as compared to electrical
parameters, many researchers have focused on this area [Lee, 2003; Zhang, 2011]. The
dielectric flushing affected the EDM performance due to changing erosion rate, mirror like
finish achieved by multi divided electrode method investigated the effect of rotary tube
electrode on machining characteristics of Al2O3/6061Al composite. Improved jet flushing for
EDM was investigated. They found that the distribution phenomenon of debris had a good
correlation with the geometry of the work piece surface produced.
2.6.4 Performance characteristics of EDM process

As discussed in the above section, the choice of non-conventional process of machining is due to metallurgical difficulties in machining very hard material or the materials which undergo fracture / transformations upon mechanical stresses. Hence, the most important performance characteristic of EDM process is its ability to remove the material from its work piece. According to the principle of EDM process, it can easily be understood, that the electric discharge, which is used to remove the material causes craters on EDMed surface resulting in lower surface finish. The surface produced by the EDM process consists of a multitude of small craters randomly distributed all over the machined face. The quality of surface finish depends upon energy per spark. If energy content is higher, deeper crater will result leading to poor surface. The surface roughness has been found to be inversely proportional to frequency of discharge. The EDM process is generally designed either to maximizing MRR / surface finish or optimizing both. The work of many researchers [Abbas, 2007; Sundaram, 2008; Kao, 2010; Chiang, 2007; Lee, 2003] studied both of these characteristics with varying the input process parameters like polarity, peak current, voltage, pulse on time, pulse off time, nature of dielectric, flushing pressure etc.

2.6.5 Applications of EDM

The EDM process is most widely used by the mould-making tool and die industries, but is becoming a common method of making prototype and production parts, especially in the aerospace, automobile and electronics industries in which production quantities are relatively low. It is used to machine extremely hard materials that are difficult to machine like alloys, tool steels, tungsten carbides etc. Drilling of curved holes, internal thread cutting and helical gear cutting with higher tolerance limits can be obtained in EDM machining. Hence areas that require higher surface accuracy use the EDM machining process. Ceramic materials that are difficult to machine can be machined by the EDM machining process.

2.7 Wear study

Wear study is related to surface interactions and more specifically the removal of material from a surface of a material as a result of its interaction with the other surface in contact during relative motion between surfaces. Also the wear is defined as progressive damage resulting in material loss due to relative contact between adjacent components under relative motion. Wear is not a material property, however, it is a systems response. In order to extend it as an inherent property of a material, the researchers have standardized the condition of the other surface in
contact during the wear tests. The two surfaces on contact can be in the form of discs or as a pair of a pin and a disc. Researchers have used different pairs depending on the property under investigation. Researchers like Zimba, 2004 have used pair of discs when contact fatigue and subsurface transformations. Pin on disc has been used by many researchers when the wear property like material removal rate, characteristics of worn out surface and coefficient of friction between the two surfaces is the prime area of research. Another variable in the study of wear of material is the condition of wear. Most tribological pairs are supplied with a lubrication system so as to avoid the excessive wear and damage during investigation for relatively softer material and wear resistant materials are allowed to rub together in dry condition to attain measurable wear within the laboratory test duration. Although some wear is always expected during normal operation of equipment but excessive friction causes premature failures and this creates significant economic costs due to equipment failure, cost for replacement parts and downtime. Friction and wear also generate heat, which represents wasted energy that is not recoverable. In other words, wear is also responsible for overall loss in system efficiency.

2.7.1 Wear mechanisms

The simplest classification of surface interactions is into those involving either mild or severe wear. The generally observed differences between the two regimes of mild and severe wear in variety of sliding metallic systems are summarized in Table 2.2. [Bhushan, 1996].

Table 2.2: The distinguishing characteristics between mild wear and severe wear.

<table>
<thead>
<tr>
<th>Mild Wear</th>
<th>Severe Wear</th>
</tr>
</thead>
<tbody>
<tr>
<td>Produces extremely smooth surfaces – even sometimes smoother than the original.</td>
<td>Rough, deeply torn surfaces – much rough than the original.</td>
</tr>
<tr>
<td>Debris is extremely small, typically in the range of 100 nm (3.28 × 10 ft)-13in.</td>
<td>Large metallic wear debris, typically up to 0.01 mm (3.28 × 10 ft) in.</td>
</tr>
<tr>
<td>High electrical contact resistance, but little true metallic contact.</td>
<td>Low contact resistance, but true metallic junctions are formed.</td>
</tr>
</tbody>
</table>

From the contact interaction viewpoint, the wear can be divided into four fundamental modes: adhesive wear, abrasive wear, fatigue wear and corrosive wear. These are explained in details in following sections.
2.7.1.1 Adhesive wear

Adhesive wear occurs when two nominally flat solid bodies are in sliding contact or pressed into one another and fragments are pulled off one surface and adhere to the other, due to the strong adhesive forces between atoms [Bhushan, 1996]. It is the most common form of wear and is commonly encountered in conjunction with lubricant failures. It is commonly referred to as welding wear due to the exhibited surface characteristics. It usually occurs when lubrication is inadequate. It results in metal transfer usually called scoring. Metal/metal contact should be prevented to combat adhesive wear. This can be obtained by applying a thin layer of lubrication oil or by applying a protective metal layer like: nitride, oxides and carbides. The tendency of contacting surfaces to adhere arises from the attractive forces that exist between the surface atoms of the two materials. The type and mechanism of attraction varies between different materials. Most solids will adhere on contact to some extent, however, oxidation films and contaminants naturally occurring, generally suppress adhesion.

Fig. 2.30: Adhesive wear produced by formation and subsequent shearing of welded junctions between two sliding surfaces [Bhushan, 1996].

2.7.1.2 Abrasive wear

Abrasive wear is damage to a softer surface by one or more of the harder material. The mechanisms of abrasive wear are micro-cutting, micro-plowing and micro cracking. There are two types of abrasive wear:

- Two body abrasion: The first body is a softer material and the second one is a harder material, e.g. grinding, cutting.
- Three body abrasion: Here two bodies rub against each other with a third abrasive medium in between them, e.g. polishing of metallurgical samples on polishing papers.
There are a number of factors which influence abrasive wear and hence the manner of material removal. Several different mechanisms have been proposed to describe the manner in which the material is removed. Three commonly identified mechanisms of abrasive wear as plowing, cutting and fragmentation. Plowing occurs when material is displaced to the side, away from the wear particles, resulting in the formation of grooves that do not involve direct material removal. The displaced material forms ridges adjacent to grooves, which may be removed by subsequent passage of abrasive particles.

Fig. 2.31: Two body abrasive wear when one surface is harder than the other, cuts material away (a) and (b). Three body abrasion mechanism as the wear debris then acts as an abrasive between the two surfaces (c) and (d)[Bhushan, 1996].

Cutting occurs when material is separated from the surface in the form of primary debris, or microchips, with little or no material displaced to the sides of the grooves. This mechanism closely resembles conventional machining. Fragmentation occurs when material is separated from a surface by a cutting process and the indenting abrasive causes localized fracture of the wear material. These cracks then freely propagate locally around the wear groove, resulting in additional material removal by spalling.
2.7.1.3 Fatigue wear

Repeated cycles of contact are necessary in the generation of wear particles by fatigue wear. The repeated loading and unloading cycles to which materials are exposed may induce the formation of sub-surface or surface cracks, which eventually will result in the fracture of the surface with the formation of large fragments, leaving large pits in the surface. Cracks are nucleated at and below the surface as the deformation goes on. Further loading and deforming causes cracks and pre-existing voids to extend and propagate. After a critical number of contacts, an asperity fails due to fatigue, producing a wear particle.

2.7.1.4 Corrosion wear

Corrosion wear or chemical wear occurs when the sliding takes place in corrosive liquids or gases. In air atmosphere, the most dominant corrosive medium is oxygen. Therefore wear in the air atmosphere is usually called oxidative wear. When there is no sliding between surfaces, the chemical products of the corrosion may form a thin, protective film. The film tends to slow down the corrosion. The sliding action, however, wears the chemical film away and allows the chemical attacks to continue. The chemical wear requires both chemical reactions and rubbing. The rate of chemical reaction increases by the frictional heat produced at the contact point of sliding surfaces. This means that the reactions which normally occur only at high temperatures may occur at moderate or even at ambient temperatures during sliding.

2.7.2 Variables influencing wear

The wear process has a large number of variables and by understanding how they influence the friction and wear, the process can be controlled and evaluated. Variables influencing wear are combined to the following categories which are considered as significant: materials and the operational, geometric and environmental factors. Due to the large amount of variables, only the important ones from each category are presented below.

The compositions, properties and microstructure of the materials determine their wear rates under different operating conditions. The properties of particular significance are hardness, impact strength, toughness, modulus of elasticity, corrosion resistance and fatigue resistance. As mentioned previously, the solid solution of materials has also an influence on wear rate. In general, the wear of alloy tends to be lower than that from pure components. Based on
chemical composition and processing, a variety of microstructures and physical properties of DI can be obtained.
The operational variables which significantly influence the wear are normal load, sliding velocity, time and temperature. As the load is increased, the wear rate will increase regardless of amount of lubrication. An increase in normal load also increases frictional heat. Generally the increase in sliding velocity has the same effect as load. However a decreasing wear rate may occur at higher velocities, because there is less time available to develop a full wear particle. In the initial stage, the wear rate is high and decreases to a constant value when the surface contact area increases with time. The wear rate increases after operating for a definite period of time. The transition can be attributed to a number of causes such as change in type of wear and increase of surface temperature to a critical value or lubricant contamination.
Geometric variables are those defining the geometric contacts between the solids and between different materials. They are typically surface roughness, surface topography, particle size and shape. The change in contact characteristics which occurs during the wear process can be observed and explained on macro level, micro level and nano level [Garrison, 1982].
Environmental variables are typically the amount of lubrication, contamination, ambient temperature and atmosphere. For instance, change in temperature influences wear rate by modifying the surface properties of materials and properties of lubricants. Increasing humidity usually makes the wear process faster [Garrison, 1982].

2.8 Wear of ADI
Abrasion resistance of ADI has been studied by many researchers keeping in view of its application in automobile industry components and equipments for agricultural ploughing applications. Most studies undertaken are under dry conditions. As the primary cause of wear is abrasion, hence most of the studies by the researchers are under ADI offers the design engineer abrasion resistance that is superior to competitive materials over a wide range of hardness. Generally, ADI will outwear competitive materials at a given hardness level. For example from Fig. 2.32, an ADI component at 30 to 40 Rc will wear comparably to a quenched and tempered steel component at nearly 60 Rc in an abrasive wear environment. This property, as shown in Figs. 2.33 and 2.34, allows the designer to select the combination of strength, ductility and abrasion resistance, that will provide the best component performance in a particular application.
The superior abrasion resistance and the low sensitivity of abrasion resistance to bulk hardness are related to the strain-induced transformation of stabilized austenite, which occurs when the surface of an ADI component is subjected to deformation. The result of this transformation is a significant increase in surface hardness shown in Fig. 2.35. This increase in surface hardness and its relationship to microstructure, is responsible for the reduced sensitivity of abrasion resistance to hardness. As the bulk hardness of ADI is reduced by the austempering temperature, the amount of stabilized austenite increases. This increase in austenite content increases the hardness increment produced by surface deformation. As a result, a DI component austempered to produce a lower hardness displays an abrasion
resistance greater than that predicted by its bulk hardness, provided that the abrasion mechanism involves sufficient deformation to transform the surface layers to martensite. Through variations in austempering conditions, the designer can optimize the abrasion resistance and related mechanical properties of an ADI component. For a combination of high toughness and abrasion resistance, an austempering temperature in the range of 350 °C-375°C should be used. When a combination of high strength and abrasion resistance are required, an austempering temperature of 260°C will yield the best results.

![Fig. 2.34: Micro hardness scan of an abraded ADI sample [Garrison, 1982].](image)

![Fig. 2.35: Relationship between the austempering temperature and the amount of stabilized austenite surface of abraded and unabraded ADI samples [Garrison, 1982].](image)

2.8.1 Coefficient of friction

The Coefficient of friction is the property of the metal surfaces in contact, which defines its state of roughness / smoothness. Numerically it is defined as the ratio of the normal force acting on the surfaces of contact to its normal reaction. Coefficient of friction is an inherit property of two materials in contact and is constant. Under dynamic and static conditions in
engineering applications, static friction has no significance, hence the researchers [Ahmadabadi, et.al 1999] have calculated and discussed coefficient of friction under dynamics conditions. Experimentally the abrasion tests as undertaken by all the researchers have shown that the coefficient of friction is independent of the amount of normal load acting on the surface and depends on the initial state of roughness of samples. As abrasion wear ultimately smooth the surfaces under contact, hence the coefficient of friction starts from a lower value and then ultimately reaches a steady state. Some researchers have [Garrison, 1982] accounted for different values of coefficient of friction for different ADIs under investigation. Researchers have correlated the coefficient of friction with the microstructural phase of the surface of the pin. Martensite being brittle shows higher value of coefficient of friction, martensite shows $\mu = 0.7$ to $0.8$ and for microstructure of fine bainite, it is of the range of $0.4$ to $0.5$.

2.8.2 Worn out surfaces

Another characteristic of the wear surface is through the development of correlation between the material removal rate and study of worn out surfaces. Worn out surfaces also tell about the mechanics of wear. The presence of large and small pits removal of metal by abrasions and the presence of irregular pits is due to the mechanism of gauging. The presence of larger and smaller scars is due to the mechanism called plowing. Graphite being a dry lubricant tends to produce long streaks and deformation of graphite nodules.

2.9 Problem formulation

Detailed literature review carried out on ADI, comprehensively discusses the microstructure development, mechanical properties of ADI and effect of common alloying elements like molybdenum, manganese, nickel, copper etc. The research work of some authoritative author in this field like Darwish, Rouns, Elliot, O. Eric and Garrison developed the understanding that, the theory correlating the mechanical properties with composition, austempering parameters and microstructures has already been established. The initial attempts to improve mechanical properties were made by researchers, by designing the special composition of ductile iron for laboratory experimentation. In this attempt, low manganese ductile irons have been studied for producing ADI with better mechanical properties including tensile properties, toughness, impact strength etc. Though the manganese is the most potent austemperability agent but it segregates at eutectic cell boundaries and hence delays the acicular transformation. This makes processing window narrower. Even on air cooling after austempering, it results in...
formation of martensite. The production of low manganese alloyed ductile iron is extremely difficult from practical point of view. The commercially viable ductile irons in Indian scenario uses steel scraps as raw material for ductile iron production that invariably contain 0.2 - 0.5 Wt. % Mn which gives a high input of manganese to ductile iron. So, high manganese content, poses a challenge for ADI researchers, as it deteriorates the tensile properties. However it is felt that there is need to explore potential applications for ADIs developed from high manganese ductile iron, where tensile properties are not primarily important. Because, wear can be an area for applications of such ADIs, therefore, wear properties are needed to be investigated for such ADIs. To counter the effect of high manganese in ductile iron addition of copper, nickel has been recommended in literature. As copper and nickel do not segregate at cell boundaries, they promote the austemperability. The studies available in literature involve very high percentages of nickel additions in the range of 2 to 4.5%. Nickel being costly, its addition in ductile iron has to be optimized for economic considerations. Thus a need is felt to produce commercially viable iron with minimum possible nickel content. The effect of nickel addition on structure, austempering kinetics, processing window, mechanical properties, wear performance and machining of ADI should be investigated. In view of the above, the present study aims to develop four commercially viable ductile iron compositions with and without nickel addition. Nickel composition was kept in the range of 0.1 – 0.6%. Copper and nickel were chosen as these elements do not segregate leading to non-uniform austempered microstructure. The present work deals with commercial viable ductile iron with 0.2 – 0.25 Wt. % Mn in its ductile iron composition and the addition of nickel and copper was made judicious keeping in view its role for producing uniform ausferrite microstructure.

On basis of literature review, the ductile iron used for the present study may not be suitable for the application requiring good tensile properties; however, it can become a good choice for the wear applications for agricultural, mining and digging implements etc. Since ADI is a complex material, it is not easy for the industry to exploit its potential in different industrial applications, and only specialized parts have been developed using the trial and error method. Research till date is a lot less committed to the wear applications of ADI and development of basic theory of phase transformations, for the benefit of wear resistant ADI. There is need for close control on heat treatment parameters, time and temperature of austenitization and time and temperature of austempering to attain the required austempered structure having the desired wear performance. A close estimation on heat treatment parameters shall be of great commercial importance. Correlating structural parameters in austempered structure to the wear behavior would be of help in understanding the role of heat treatment parameters in order to
achieve their optimum combinations. However no attempt till date has been made in the
direction.

First step of ADI treatment is austenitization. ADI has a distinction here from that of steel
because of the presence of graphite nodule acting as reservoir of carbon. The austenitizing
temperature and time are important in attaining uniform carbon content which, contributes to
the uniform microstructure in ADI. Therefore optimization of austenitization time and
temperature for any ductile iron composition is necessary. Austempering follows the
austenitization and evolution of austempered microstructure during austempering process
depends on important variables like time and temperature of austenitization and austempering
apart from the composition of the base iron. The austempered structure would be examined
under optical microscope in order to understand the range of microstructure and its
constituents, which would be developed by the heat treatment of ductile iron used in present
study. The corresponding spectrum of mechanical properties in terms of hardness, 0.2% proof
stress, ultimate tensile stress, percentage elongation and impact strength were determined.
XRD studies were planned for quantitative studies of phase constituents and correlated these
with microstructure, mechanical properties and wear behavior.

Possibly, the most significant hurdle for the engineering community is to overcome the
difficulties for the successful machining of ADI, to fully realize the potential of ADI.
Generally when machining is conducted on components prior to the heat treatment, it never
offers significant difficulty but machining post heat treatment is always avoided. The phase
transformation of high carbon austenite to martensite leads to poor machinability
characteristics, which will be investigated systematically.

Now a days, industries are facing problems in manufacturing of components due to several
reasons such as the complexity of the job profile, surface requirements with higher accuracy
and surface finish. These problems become more critical for the materials involving high
strength. It is very difficult to find tool material, which is sufficiently hard and strong to cut
such materials. The solution lies in the fact that a process may be designed for machining of
the materials where the physical and metallurgical properties of the materials do not pose any
challenge to the machining process. Non-conventional machining methods such as electrical
discharge machining (EDM) offer solution to such problems, but involve a very high cost. The
present study involved determination of EDM process parameters for machining of ADIs so as
to establish data, which could be used by the industry.

In summary, the present study has been undertaken to focus attention on a limited segment of
desired alloying elements in commercially viable ductile iron developed in Indian scenario and
investigate the effect of heat treatment parameters on resultant ADI microstructure, mechanical properties, machinability and wear performance. The investigation aimed to determine EDM process parameters for machining of ADI.